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Growth and structure of Cr ultra-thin films deposited on Co(0001)

P. Ohresser [a], F. Scheurer [a,*], B. Carrière [a], J.P. Deville [a], A. Dobroiu [b]

[a] *Institut de Physique et Chimie des Matériaux de Strasbourg, UMR 046, 23 rue du Loess F-67037 Strasbourg Cedex, France*
[b] *Institutul de Fizica Atomica BP MG 6, 76900 Bucuresti-Magurele, Romania*

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Abstract

The growth and structure of ultra-thin Cr films deposited on a Co(0001) single crystal are investigated for substrate temperatures ranging from 300 to 500 K by means of Auger electron spectroscopy, low energy electron diffraction and photoemission. For room temperature growth, the interface is sharp and the Cr layers arrange in a Nishiyama–Wassermann-type epitaxy. At elevated temperatures the interface is diffuse and a Kurdjumov–Sachs orientation is observed.

Keywords: Chromium; Cobalt; Epitaxy; Growth; Magnetic films; Surface structure

1. Introduction

The investigation of metallic multilayers has become an important field in magnetism since some of them show interesting properties, namely perpendicular anisotropy, giant magnetoresistance and interlayer coupling. Among others, Co–Cr multilayers have already been realized several years ago for magnetic recording purposes, as well in the (001) orientation as in the close-packed one [1]. It is now admitted that the magnetic properties of these multilayers are strongly dependent on the sharpness of the interface, the growth mode and the structure. It is thus important to build molecular beam epitaxy (MBE) quality multilayers or thin films in order to correlate their magnetic properties and their morphology. In two recent papers on MBE-grown Co–Cr multilayers it has been shown that a structural transition occurs between 2 and 6 Å from a pseudomorphic growth of Cr towards both Nishiyama–Wassermann (NW) and Kurdjumov–Sachs (KS) orientations [2,3]. On the other hand, a pseudomorphic growth has been observed up to nearly 18 Å for thin Cr films deposited on Co(0001) oriented layers [4].

In the present paper, Cr ultra-thin films are deposited on a Co(0001) single-crystal for various temperatures between 300 and 500 K. The growth mode and the structure of the interface is analyzed by Auger electron spectroscopy (AES), low energy electron diffraction (LEED) and X-ray photoemission spectroscopy (XPS).

2. Experimental procedure

The experiments have been carried out in two different ultra-high vacuum chambers (pressure in the low 10^{-10} mb range). One is equipped with a

* Corresponding author. Fax: +33 8810 7248; e-mail: scheurer@morgane.u-strasbg.fr.

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MAC2 Riber Auger analyzer and an Omicron reverse-view LEED optics. The set-up allows a monitoring of the Auger spectra during the Cr evaporation. The second chamber is equipped with a monochromatized Al X-ray source and a VSW analyzer is used for the XPS measurements.

The Co single-crystal is cleaned by argon ion sputtering during annealing at 300°C followed by a final annealing at 350°C. The latter temperature must not be overrun in order to avoid the martensitic transition of Co [5]. The Cr is evaporated from a tungsten basket containing high-purity polycrystalline chromium. The cleanliness of the substrate and Cr films is checked by AES or XPS and the contamination level is kept below the detection limit. The calibration is done by measuring the attenuation of the Auger LMM transition on the room temperature grown film showing perfect exponential-shaped growth kinetics, as described in Ref. [6]. The evaporation rate is estimated to about 0.10 ± 0.03 Å/min. The inelastic mean free path taken is approximately 9 Å for the LMM transitions, compatible with Ref. [7].

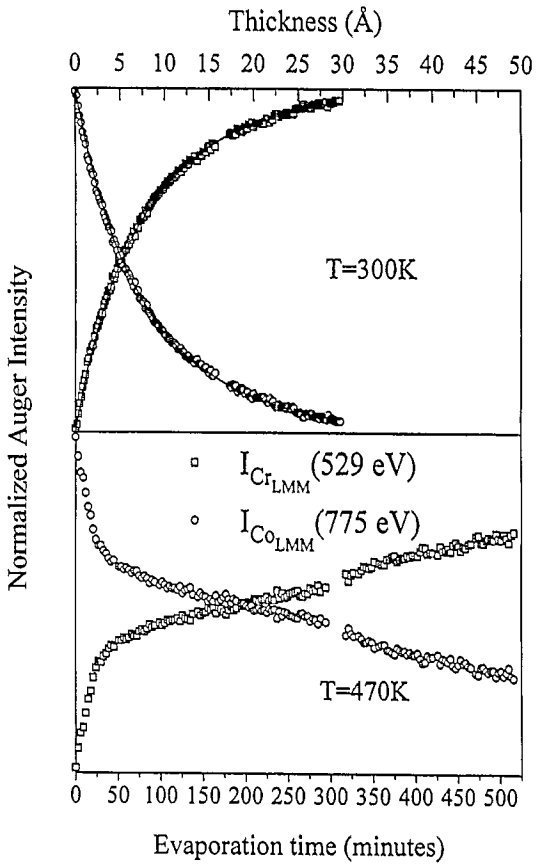

Fig. 1. Cr and Co LMM Auger intensities as a function of the evaporation time at 300 K (a) and 470 K (b).

3. Results

Auger growth kinetics are recorded for different substrate temperatures between 300 and 500 K. For temperatures below approximately 400 K (see Fig. 1a), the intensity of the Co Auger transitions decays exponentially with the deposited Cr amount and the high energy transitions are no longer observable after about 30 Å deposition (one monolayer having a thickness of approximately 2 Å). The Cr Auger transition intensity increases conversely. Strong interdiffusion can thus be ruled out. The observed behaviour is characteristic of a simultaneous layer growth mode i.e. when new layers start growing while the previous ones are not completed [8,9]. Below 400 K the LEED pattern shows well defined additional spots around the main spots, up to at least 12 Å (Fig. 2a). This pattern is typically obtained for a bcc (110) plane epitaxy on a close-packed surface in the NW orientation as demonstrated in Fig. 2b: the simulated NW pattern results from the superimposition of three equivalent bcc (110) domains, the bcc $\langle 001 \rangle$ direction being parallel to a dense substrate row (of the $\langle \bar{1}2\bar{1}0 \rangle$ type). Generally three satellite spots are observed instead of two for low coverage. In the case of Cr on Co(0001) however, the third spot is hidden by the main substrate spot because of the very close nearest neighbour distances of Co and Cr (2.507 and 2.498 Å respectively). On further Cr deposition, the diagram symmetry is conserved but the LEED spots become more elongated and the background brighter.

For growth temperatures above 430 K, the growth mode changes drastically: the exponential decay is no longer observed in the growth kinetics (Fig. 1b). After a 2 Å equivalent coverage the intensity of the LMM Co (Cr) Auger transition decreases (increases) much more slowly. At 50 Å equivalent coverage, the low energy Auger transitions are still visible. This behaviour strongly supports the growth of a diffuse interface or a Volmer–Weber growth mode. There

are no significant energy shifts observed, neither for the Auger transitions nor for the 2p and 3p XPS peaks of Co and Cr. No further indication of interdiffusion is gained here. However, strong interdiffusion has been evidenced by nuclear magnetic resonance for 400 K grown multilayers [3]. We think thus that a surface alloy forms on top of which Cr island could grow as suggested by the following LEED experiments. Above 430 K the Co(0001) LEED pattern remains unchanged up to about 2 Å equivalent cov-

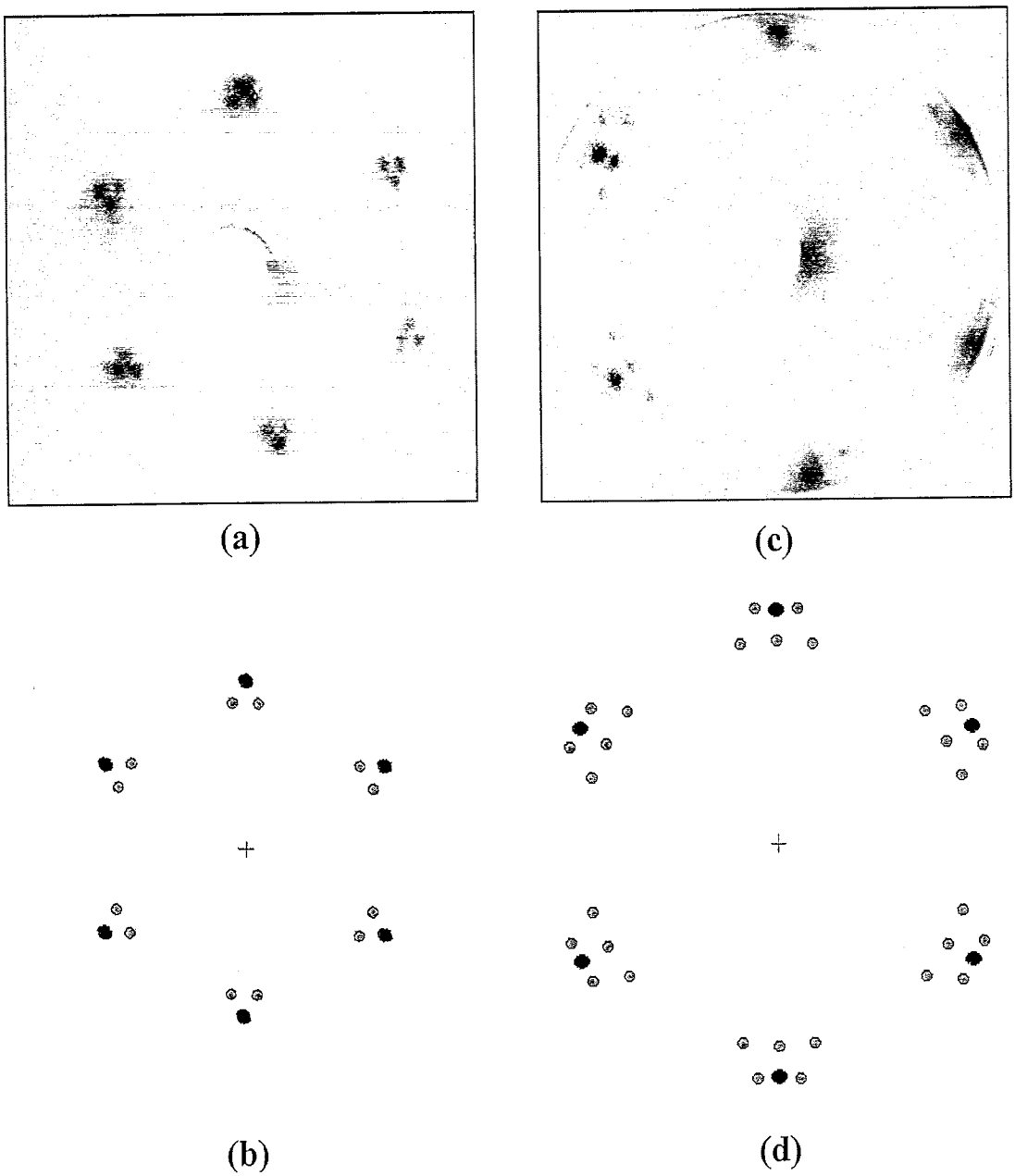

Fig. 2. (a) LEED pattern of a 6 Å Cr layer grown at 300 K (119 eV), (b) simulation of the NW orientation, (c) LEED pattern for a 50 Å-equivalent Cr layer at 470 K (68 eV), (d) simulation of the KS orientation. The simulated LEED patterns are deduced from the arrangements in Fig. 3. The LEED patterns are recorded in a slightly off-normal geometry.

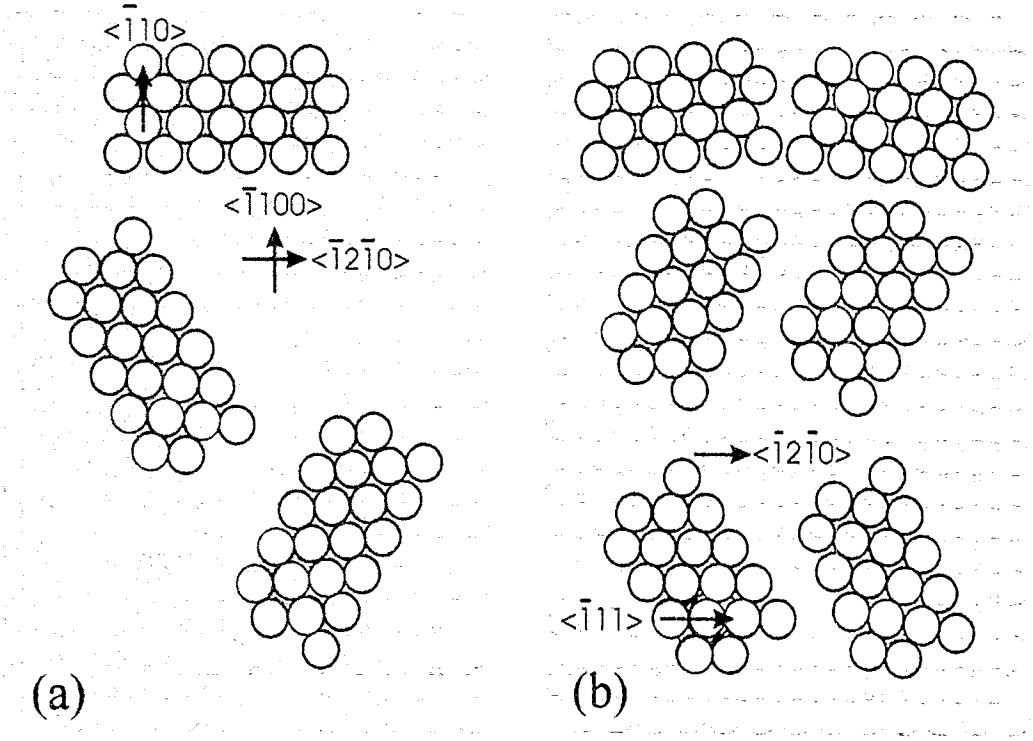

Fig. 3. Real space representation of the equivalent domains for the NW orientation (a) and the KS orientation (b). The lattice constants are those for bulk Cr and Co.

erage. Then it turns to a typical KS pattern as can be seen in Fig. 2c. Even for higher coverages, the substrate spots are still pronounced, in agreement with the AES interpretation of interdiffusion. It is however not possible to make a precise statement about the composition of the alloy or about its precise location (subsurface alloy, surface alloy or buried Cr layer). The simulated pattern (Fig. 2d) is obtained by the superimposition of 6 bcc (110) domains on a hexagonal surface, the dense rows of the adsorbate and substrate being parallel. The lattice constants used in the simulation are those of bulk Cr and Co, since no changes are evidenced in the experimental diagram.

The temperature range between 400 and 430 K has been analyzed by LEED in more details. It seems that the transition from the NW to the KS orientation occurs in an abrupt way around 420 K within a temperature interval of about 10 K. It is important to notice that, contrary to previous work [2–4], we never observe by LEED either simultaneous NW and KS patterns or pseudomorphic growth. A (1×1) LEED pattern is only obtained at higher temperatures (above 430 K) for a very low coverage. It has also been observed up to 10 Å for C-contaminated Cr films deposited at room temperature. We checked if it is possible to induce both NW and KS structures at the same time, as it has been observed for 370 K-grown multilayers [2,3], by the following procedure: first, the minimum amount of Cr (about 7 Å) is deposited at 500 K in order to observe the KS orientation; then the sample is cooled down to room temperature for further Cr deposition. The NW orientation is never observed in this experiment, and moreover the Cr goes on growing in the KS orientation, contrary to normal room temperature deposition.

4. Discussion

In the case of bcc/hcp (0001) or fcc (111) interfaces, there are mainly two types of orientational relationships, NW or KS, as described above (Figs.

2 and 3). From geometrical considerations, justified by energy calculations, it has been shown that the perfect NW orientation occurs when the fcc to bcc nearest neighbour ratio r is $r_x = 0.9428$ (NW_x) or $r_y = 1.1547$ (NW_y). The perfect KS orientation is obtained for $r_k = 1.0887$ [9,10]. In the case of the Cr/Co interface, $r = 1.004$. This value is just between r_x and r_k, placing that system in a limit case, with a very slight tendency to a NW orientation. Indeed, the NW structure is observed for room temperature growth and the Cr is immediately in the bcc phase. At higher temperatures however, we observe the KS orientation above 2 Å. A rather similar behaviour has been observed on Pd/Mo(110) for which $r = 1.009$ [11]. A pseudomorphic growth followed by a NW orientation is obtained at room temperature. At 700 K there is a Pd–Mo alloy in a KS orientation on top of Mo(110). In the case of Cr/Co a 1.3% modification of the surface parameter could in principle be sufficient to produce a switch from NW to KS, according to geometrical considerations. Indeed, alloying, observed above 430 K, can induce such a parameter change. Though, a significant lattice constant variation could not be evidenced by X-rays for Co–Cr alloys [12]. Alternatively, alloy formation can modify the adsorbate–substrate interactions, which in turn could favour the switching from NW to KS, particularly in the Cr/Co limit case. This can be understood by considering the phase diagram calculated by Bauer and van der Merwe (Fig. 4 in Ref. [9]). Near $r = 1$ there is a region where pseudomorphism, NW or KS orientation are equally probable. However, in order to study the influence of alloying, we deposited about 6 Å Cr at room temperature and then annealed the sample at 500 K. This produces a surface alloy (evidenced by Auger) keeping a (1×1) surface structure (note that annealing room temperature-grown Cr layers at 500 K does not produce the KS structure, contrary to 500 K-deposited films). On further Cr deposition at room temperature on this alloy we still observe the NW orientation. There is thus no satisfying explanation for the switching of the NW orientation at room temperature to the KS one. It rather seems from our experiments that if once a NW or KS is initiated, the structure is kept for further Cr deposition. The explanation of the different structures obtained must probably be sought in kinetically-driven effects. The island nucleation process, and thus the strain release, is different depending on the growth temperature, on the density and nature of surface defects. As it was demonstrated recently, surface steps play an important role in the misfit accommodation process occurring during the NW or KS epitaxy [13,14]. This could also explain the differences in the results obtained on flat substrates used for multilayer growth and those on single-crystals, having generally more defects.

In summary, the growth of Cr on the Co(0001) surface strongly depends on the substrate temperature. At room temperature we observe a simultaneous layer growth with a NW orientation while above 430 K the interface is diffuse and there are epitaxial islands in a KS orientation.

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References

[1] For a list of references see for instance: N. Metoki, W. Donner and H. Zabel, Phys. Rev. B 49 (1994) 17351.
[2] W. Vavra, D. Barlett, S. Elagoz, C. Uher, R. Clarke, Phys. Rev. B 47 (1993) 5500.
[3] Y. Henry, C. Meny, A. Dinia and P. Panissod, Phys. Rev. B 47 (1993) 15037.
[4] J. Kohlhepp, H. Fritzsche, H.-J. Elmers, U. Gradmann, J. Magn. Magn. Mater. 148 (1995) 95.
[5] A. Barbier, Thesis Université Louis Pasteur (1993) unpublished.
[6] D. Tian, F. Jona, P.M. Marcus, Phys. Rev. B 45 (1992) 11216.
[7] M.P. Seah and W.A. Dench, Surf. Interface Anal. 1 (1979) 2.
[8] C. Argile and G.E. Rhead, Surf. Sci. Rep. 10 (1989) 280.
[9] E. Bauer and J.H. van der Merwe, Phys. Rev. B 33 (1986) 3657, and references therein.
[10] M. Kato, Mater. Sci. Eng. A 146 (1991) 205.
[11] Ch. Park, E. Bauer and H. Poppa, Surf. Sci. 154 (1985) 371.
[12] M.C. Cadeville, private communication.
[13] G.J. Shiflet and J.H. van der Merwe, Metall. Mater. Trans. 25 A (1994) 1895.
[14] G.J. Shiflet and J.H. van der Merwe, Acta Metall. Mater. 42 (1994) 1189 and 1199.