Fabrication of 1D metal nanostructures on a vicinal Au(111) surface *
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The growth of Mn and Gd on a vicinal Au(111) surface is observed by scanning tunneling microscopy at room temperature. One-dimensional nanostructures have been fabricated with Mn adatoms. Dosed Mn atoms are adsorbed at lower corners of {111} steps, and Mn adlayers grow in a step-flow growth mode on (111) terraces. On the other hand, there is no tendency for Gd to form one-dimensional structures. Dosed Gd atoms nucleate not only at lower corners of the steps but also on the terraces. As the coverage is increased, Gd adislands isotropically grow in size around the nucleation sites on the terraces. The mechanism of these different growth manners is qualitatively discussed. [DOI: 10.1380/ejssnt.2003.142]

Keywords: Manganese; Gadolinium; Gold; Vicinal surface; One-dimensional; Growth; Scanning tunneling microscopy

I. INTRODUCTION

The fabrication of low-dimensional metal nanostructures has received a great deal of attention in recent years. Theoretical works have predicted the appearance of strongly enhanced magnetic moment for one monolayer (ML) films of 3d transition metals on noble metal substrates [1, 2]. A large number of experimental investigations have been carried out on the growth of transition metal thin films. Manganese is expected to have large magnetic moment in its low-dimensional structures such as thin films (two-dimensional (2D)) [3-6] and substitutional impurities (zero-dimensional) [7, 8] although bulk Mn is not ferromagnetic. On the other hand, gadolinium is one of the four ferromagnetic pure metals (Fe, Co, Ni and Gd) and has been studied intensively by experiment and theory. Due to the small surface energy of Gd, its adlayers in the submonolayer range tend to exhibit various superstructures with low atomic density [9-13]. The relation between magnetic order and atomic geometry in the 2D Gd adlayers is the subject of this recent study.

One-dimensional (1D) monatomic chains and 1D nanostructures are also good candidates that can exhibit enhanced magnetism peculiar to their low-dimensionality [14]. It is well known that vicinal surfaces are ideal templates for fabrication of 1D nanostructures [12-17]. Since self-organized 1D nanostructures can be established with long-range order, the vicinal substrates are regarded as suitable substrates for investigations of the 1D system by using macroscopic techniques such as photoemission spectroscopy and electron diffraction. From this point of view, we have been using vicinal Au(111) surfaces to build 1D or 2D nanostructures. In our recent works, it was confirmed by scanning tunneling microscopy (STM) that dosed Fe atoms are adsorbed at Au steps and form 1D nanostructures [18]. Furthermore, angle-resolved photoemission spectroscopy over a wide range of Fe coverages revealed that these Fe nanostructures have 1D nature not only about the geometric feature but also about the electronic states, which seems to be reinforced by the 1D electronic structure of underlying vicinal Au(111) substrates [19].

In this study, we investigate the growth of Mn and Gd on Au(111) and its vicinal surface by STM at room temperature (RT). We dose a small amount of Mn on the Au(111) surface and observe the nucleation and growth of Mn adislands. In addition, we present STM images of 1D Mn chains and 1D Mn adlayers formed along {111} steps on the Au(788) surface. Furthermore, we attempt to fabricate 1D Gd nanostructures on Au(788) in the same way. By making comparison of STM images taken after deposition of Mn and Gd, we discuss the growth mechanism within the submonolayer range focusing on adsorption energy, surface energy and diffusivity.

II. EXPERIMENTAL

We employed Au(111) and Au(788) crystals as substrates for the growth of Mn and Gd. The Au(788) surface is misoriented by the angle of 3.5° towards the [111] azimuth with respect to the (111) plane. Its terrace width is 39 Å in average, each of which consists of sixteen atomic rows [20]. The Au substrates were cut to the specific orientations and shaped by spark erosion. They were chemically etched and mechanically polished to obtain good metal surfaces (final polishing step = 0.03 µm). The accuracy of the misorientation of the vicinal substrates was less than 0.1°. Once introduced into an ultrahigh vacuum (UHV) chamber, the samples were treated by extensive sputtering (Ar+, 400 eV, 15 nA) and annealing (400 °C) cycles. After cooling the samples to RT, Mn (Gd) was dosed with an electron beam evaporator. The flux was measured with a water-cooled quartz-crystal film-thickness monitor. A typical deposition rate was about 0.05 ML/min. STM experiments were performed using a Beetle type STM (Besocke Delta Phi GmbH [21]) with an electrochemically etched W tip. All STM images were

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III. RESULTS AND DISCUSSIONS

Figure 1 shows an STM image of the Au(111) surface taken after Mn deposition of 0.1 ML. The Au(111) surface reconstructs to relieve the surface stress, and exhibits the herringbone structure where discommensuration lines run in the $[\overline{2}11]$ directions and separate the surface into regular fcc and faulted hcp stacking regions [22, 23]. As can be seen in the STM image, dosed Mn atoms aggregate and form Mn adislands at elbows of the herringbone zigzag geometry. This leads to a periodic array of the Mn adislands, as regularly spaced stripes of Mn [6]. Furthermore, Fig. 1 shows a number of depressions at the elbows instead of the Mn adislands. Meyer et al. found in their STM study on Ni/Au(111) that such depressions can be considered as Ni atoms inserted in a surface layer, and act as preferential nucleation sites for the island growth [24]. It can be analogically concluded that the dosed Mn atoms are substituted in the Au surface layer at the elbows prior to the formation of the Mn adislands. The same behavior is observed for the cases of Fe [25, 26], Co [27, 28], Ni [24, 29, 30], Rh [31] and Pd [32], whereas Al [33], Ag [34, 35] and Au [30, 36] do not show such preferential nucleation. It is believed that the preferential nucleation is due to the large mismatch of the substituted atom with the substrate for Fe, Co, Ni, Rh or Pd [37], which rather relaxes the lattice strained at the elbows. Since Mn has an atomic radius smaller than Au and comparable to Pd, Mn would be expected to exhibit the preferential nucleation.

Figures 2(a) and 2(b) show the growth of Mn on the Au(788) surface, which were taken after Mn deposition of 0.08 and 0.3 ML, respectively. In contrast to the Au(111) surface, vicinal Au(111) surfaces exhibit regularly arranged monatomic height steps with a parallel configuration, and the discommensuration lines run perpendicular to the steps [38]. Consequently, the herringbone structure can be no longer observed on the surfaces. As is shown in Fig. 2(a), dosed Mn atoms are adsorbed at lower corners of the Au steps and form 1D Mn wires along the steps. It is observed that the second Mn rows begin to grow after the complete formation of the first ones. How-
FIG. 3: Growth of Gd on the Au(111) surface. The Gd coverage is 0.1 ML.

However, the succeeding Mn growth is not perfect row-by-row growth mode as shown in Fig. 2(b), hence this corresponds to the initial stage of a 2D Stranski-Krastanov (SK) growth mode. As the coverage is increased, Mn adlayers grow on the (111) terraces in a step-flow growth mode. In Fig. 2(b), it can be clearly seen that straight interfaces between the Au(111) terraces and the adjacent Mn adlayers form parallel lines with regular spacing. This suggests that the straight steps observed on the clean Au(788) surface are not affected by the Mn attachment but conserved as they were without the formation of a Mn-Au alloy.

Next, we show the growth of Gd on the Au(111) and its vicinal surfaces. Figure 3 shows an STM image taken after deposition of 0.1 ML Gd on the Au(111) surface. As is often the case of Mn and other transition metals, dosed Gd atoms are substituted in a surface layer prior to the formation of the Gd adislands, which are observed as depressions in the STM image. However, it is worth noting that the depressions can be seen not only at the elbows of the herringbone reconstruction but also at other locations. Generally, the location of nucleation sites gives valuable information about the adsorption energy. Our experimental results suggest that Mn has the smallest adsorption energy at the elbows, whereas there is no significant difference in adsorption energy for a Gd adatom between at the elbows and at other sites. Goyhenex and Bulou calculated the adsorption energies of a Co adatom at various sites on the reconstructed Au(111) surface and found that the adsorption energy is lower at the discommensuration lines than in the commensuration regions, which gives support to the preferential nucleation at the discommensuration lines [39]. However, since nucleation of Gd is observed evenly at various sites, this is not the case of Gd on Au(111).

Figures 4(a) and 4(b) show the growth of Gd on the Au(788) surface. It can be clearly seen from Fig. 4(a) that dosed Gd atoms are adsorbed at lower corners of the steps. This indicates that the step sites give the minimum adsorption energy to the Gd adatoms similarly to other cases. However, at the same time, some depressions are observed on the (111) terraces. This may be possibly due to the small difference of the adsorption energy between at the step sites and on the terraces. If the effect of the periodic potential caused by the regular array of the {111} steps of the Au substrate is negligible, the...
behavior of the nucleation must be similar to that on the flat substrate. Another possible origin is small mobility of the Gd adatoms on the Au(111) plane. This is probably the case that the Gd adatoms have small diffusivity and high miscibility, which urge the Gd atoms to nucleate on the terraces before they reach the steps. As the coverage is increased, Gd adislands are formed both at the steps and on the terraces, as shown in Fig. 4(b). Contrary to the growth of Mn on the vicinal Au(111) surface, no tendency of the step edge decoration and 1D wire formation is observed.

The growth modes in lower dimensions can be analogously compared to those in three dimensions [40] such as layer-by-layer, SK and island growth. Since most of transition metals exhibit relatively high surface energy, it is generally difficult to fabricate their smooth thin films in a layer-by-layer growth mode. Instead, it is known that generally difficult to fabricate their smooth thin films in a 2D island growth mode on vicinal substrates. Instead, it is known that they tend to grow in an island growth mode on flat substrates and in a 2D island growth mode on vicinal substrates, respectively. In contrast, it was found by Fonin et al. that Mn thin films grow in the layer-by-layer growth mode up to 3 ML on the Au(111) surface [6]. In this study, we have found that the Mn adlayers grow in the 2D SK growth mode on the vicinal Au(111) surfaces. This is probably because the surface energy of Mn is relatively small compared to those of other transition metals, but the surface energy of Mn is relatively close to that of Au [41-46].

On the other hand, the surface energy of Gd is characteristically small compared to those of Mn and Au [41-46]. Nevertheless, Gd does not exhibit such a smooth 1D formation on the vicinal Au(111) surface. As we observed, the dosed Gd atoms nucleate both at the step sites and on the terraces and the Gd adislands grow isotropically, hence the 1D wire formation is eventually hindered. Raising the substrate temperature during the deposition may be effective to fabricate smooth 1D Gd wires [13], although it must suffer from the intermixing with substrate.

On the other hand, Gd does not show any preferential formation on the Au(111) and its vicinal surfaces. On the Au(111) surface, the dosed Gd atoms are adsorbed not only at the elbows but also at other surface locations. These adatoms are substituted in the surface layer, and act as nucleation sites for the random formation of the Gd adstructures. We found that the step edges preferentially adsorb Mn adatoms, and 1D adlayers grow in a step-flow (2D SK) mode on the (111) terraces. The smooth formation of the 1D Mn nanostructures is encouraged by the minimum adsorption energy at the step sites, and the low surface energy of Mn comparable to Au.

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