Temperature-dependent electron microscopy study of Au thin films on Si (1 0 0) with and without a native oxide layer as barrier at the interface

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Abstract
Real-time electron microscopy observation on morphological changes in gold nanostructures deposited on Si (1 0 0) surfaces as a function of annealing temperatures has been reported. Two types of interfaces with silicon substrates were used prior to gold thin film deposition: (i) without native oxide and on ultra-clean reconstructed Si surfaces and (ii) with native oxide covered Si surfaces. For ≈2.0 nm thick Au films deposited on reconstructed Si (1 0 0) surfaces using the molecular beam epitaxy method under ultra-high vacuum conditions, aligned four-fold symmetric nanogold silicide structures formed at relatively lower temperatures (compared with the one with native oxide at the interface). For this system, 82% of the nanostructures were found to be nanorectangle-like structures with an average length of ≈27 nm and aspect ratio of 1.13 at ≈700 °C. For ≈5.0 nm thick Au films deposited on Si (1 0 0) surface with native oxide at the interface, the formation of a rectangular structure was observed at higher temperatures (≈850 °C). At these high temperatures, desorption of gold silicide followed the symmetry of the substrate. Native oxide at the interface was found to act like a barrier for the inter-diffusion phenomena. Structural characterization was carried out using advanced electron microscopy methods.

(Some figures in this article are in colour only in the electronic version)

1. Introduction
As the dimensions of electronic devices decrease with increasing packing density, the thickness of metal layers decreases continuously. An understanding of the growth of metal thin films on a semiconductor becomes important both from fundamental and practical points of view. The interaction of metal films with silicon has attracted considerable attention because of its importance in semiconductor technology. Metal nanoparticles on surfaces are being used effectively in nanofabrication, as catalysts for carbon nanotube growth [1, 2], catalysts for the vapour–solid–liquid (VLS) growth of inorganic nanowires [3–5] and for nanoelectronics applications [6–8]. The formation of nano-metal silicides, used as Schottky barriers and ohmic contacts in devices, is of particular interest in the emerging area of nanoscience and nanotechnology. The mechanism of these interactions, the range of temperatures required and the resulting composition of Au–Si alloy depend on whether these gold nanoislands interact directly with the substrate or have to deal with a barrier, e.g. a native oxide layer, before they could interact with the substrate. Either of the processes may give rise to various kinds of self-assembled nano/microstructures depending on the symmetry of the substrate. The role of nanosilicides is important in many ways and hence a deeper understanding at atomic scale of the silicon/metal interfaces is indispensable [9]. Recently, fabrication of micro/nanoscale pits on Si substrates with
long-range ordering and reliable shape control via a facile vapour transport method assisted by Au nanoparticles has been achieved [10]. The resulting pits, in the shape of triangles, squares and wires, were obtained on Si (111), (100) and (110) substrates, respectively [10]. How desorption takes place to form regular shaped pits would be interesting to study. This paper addresses a part of this problem as we focus on the desorption of gold at high temperatures from regular shaped gold silicide structures.

Self-assembly is an attractive nanofabrication technique because it provides the means to precisely engineer the structures on the nanometre scale over large sample areas. Self-organizing nanocrystal assemblies have already shown the degree of control necessary to address the challenges of building nanometre-scale technologies [11]. A systematic study of such processes has implications on modern day semiconductor technology. Studies of processes going on at the Si/metal interfaces at low temperatures have been well attempted by several groups [11–20]. Most of these studies were done under ultra-high vacuum (UHV) conditions and on systems where there was no oxide layer in between the metal film and the silicon substrate. Many of them have reported the formation of various metastable gold silicides due to the intermixing of silicon and metal. There have been numerous interesting studies on the mechanism of relaxation of strain in such epitaxial layers. Strained epitaxial layers are inherently unstable and have interesting properties. It has been reported that dislocation formation and shape transition are the important processes by which relaxation of strain occurs [21–24]. In recent years, it has been recognized that shape changes such as island formation constitute a major mechanism for strain relief [21, 23]. Tersoff and Tromp reported that a strain-induced shape transition may occur [21]. Below a critical size, islands have a compact symmetric shape. For larger sizes, they adopt a long thin shape that allows better elastic relaxation of the island’s stress [21].

Metal silicide thin films on Si (100) have been studied in great detail for more than two decades due to the possibility of obtaining self-aligned epitaxial metal–semiconductor interfaces [19, 22]. Sekar et al [23, 24] reported the formation of silicide when a 100 nm Au film deposited on Br-passivated Si (111) was thermally annealed. The annealing treatment of the as-deposited Au/Si at eutectic temperature (~363 °C) gave rise to a silicide of composition close to Au4Si as inferred from Rutherford backscattering spectrometry (RBS) studies. Mundschau et al reported the formation of triangular islands when sub-monolayer gold was deposited epitaxially on Si (111) substrates under UHV conditions [25]. Vacuum annealing of thick and continuous gold films deposited on passivated Si (110) surface showed the formation of high-aspect ratio gold silicide (Au4Si) wire-like structures around 363 °C [26]. Earlier, we showed the formation of well-aligned low-aspect ratio gold silicide nanostructures that were grown under UHV conditions at temperatures in the range 600–700 °C in the absence of native oxide at the interface [27]. Using real-time in situ high-temperature transmission electron microscopy (with 40 ms time resolution), we showed the formation of high-aspect-ratio (~15.0) aligned gold silicide nanorods in the presence of native oxide at the interface during in situ annealing of gold thin films on Si (110) substrates [28]. This was explained using an oxide mediated liquid–solid growth mechanism. For other substrate surfaces, such as Ni (100) substrates, homo- and hetero-epitaxial growth has been studied, in which Kopatzki et al have reported the formation of nickel nanosquares/rectangles of average size 2 nm [29]. Similarly, Frank et al have reported the formation of rectangles/squares in the homo-epitaxial system of Ag on Ag (100) substrate [30]. For gold on Si (100) substrates, a detailed study on the preferred orientation growth of gold under N2 atmospheric conditions was carried out and found that structures reorient themselves at higher temperatures [31]. In this paper, no details on possible silicide formation and desorption were reported. The knowledge of real-time shape variations and structural phase transitions in Au/Si system would help to properly understand silicide growth at nanoscale and would form a basis for appropriate applications that are based on the use of gold silicide islands as catalysts. As the Si (100) substrate is used for many electronic applications in the semiconductor industry, it will be very useful to understand nucleation, growth of gold silicide nanostructures and desorption as a function of temperature.

In this paper, we present the experimental observation of the growth of aligned gold silicide nanostructures during in situ thermal treatment in a transmission electron microscopy (TEM) heating stage for gold deposited on Si (100) surfaces with and without an oxide layer at the interface. Followed by this, ex situ scanning transmission electron microscopy (STEM) measurements and high-resolution electron microscopy measurements confirm the presence of gold in these rectangular-shaped silicide structures and also show depletion of gold due to desorption of gold silicide at high temperatures.

2. Experimental

A thin Au film of thickness ≈2 nm was deposited on n-type Si (100) of resistivity 10–20 Ωcm, by the molecular beam epitaxy (MBE) method under UHV (base pressure ≈2 × 10–10 mbar) conditions [32]. The Si (100) substrates were loaded into the MBE chamber and degassed at ≈600 °C for about 12 h inside the chamber, followed by flashing for about 3 min by direct heating at a temperature of ≈1200 °C. In this process, native oxide was removed and a clean Si (100) surface was obtained. On such ultra-clean surfaces, ≈2.0 nm thick gold films were grown epitaxially by evaporating Au from a Knudsen cell. The deposition rate was kept constant at ≈0.14 nm min–1. During the growth, the chamber vacuum was ≈6.2 × 10–10 mbar. The thickness monitor was calibrated with RBS measurements. After the deposition, the sample was taken out of the MBE chamber. We have also deposited Au films of ≈5.0 nm thickness on n-type Si (100) by the thermal evaporation method under high vacuum (≈4 × 10–6 mbar) conditions. For this case, native oxide was kept intact and the thickness was monitored using a quartz crystal monitor. Planar TEM specimens were prepared from the above samples. Discs of ≈3 mm diameter were cut using an ultrasonic disc
cutter followed by lapping until it reached $\approx 100 \mu m$ thickness. Using a dimple grinder, the samples were further thinned down to $\approx 25 \mu m$ at the centre. Finally, electron transparency was achieved by low-energy Ar$^+$ ion milling. Temperature-dependent TEM measurements were done with 200 keV electrons (2010, JEOL HRTEM) at the Institute of Physics, Bhubaneswar (IOPB), India. A single-tilt heating holder (GATAN 628UHR, which has a tantalum furnace and can heat the specimen up to 1000°C) was used. The temperature was measured by a Pt/Pt–Rh thermocouple and was accurate within a couple of degrees. The holder has a water cooling system to avoid overheating of the sample surroundings and the specimen chamber, while keeping only the sample at a specified temperature. Real-time measurements were carried out using a CCD camera (GATAN 832) in which real-time movies can also be recorded. For the $\approx 5$ nm Au on Si (100) system, temperature-dependent TEM measurements (in situ, real time) were carried out using a 200 keV system (TEM at IOPB) and ex situ, room temperature (RT) and high-resolution measurements were carried out using STEM and high-resolution TEM (HRTEM) measurements using 300 keV electrons in a C$_s$-corrected FEI Titan 80/300 system at the University of Bremen, Germany.

3. Results and discussion

In situ heating experiments were carried out at a ramp rate of 7 °C min$^{-1}$. At each set of temperatures, the specimen was kept for about 30 min to achieve a stable temperature. We discuss the 2 nm Au on Si (100) system without the native oxide barrier at the interface. As shown in figure 1(a), the gold islands have irregular shaped isolated structures in the as-deposited system. The selected area diffraction (SAD) pattern shown in figure 1(b) confirms the polycrystalline nature of gold films. A close look at the diffraction pattern suggests the existence of some texturing, as more reflections from the {220} plane can be observed. A high-resolution TEM taken on nanostructures shows lattice spacing corresponding to elemental gold (figure 1(c)) from one of the isolated island structures. Figure 2 depicts the bright field (BF) images at various temperatures (200, 300, 325, 350, 363, 400, 510, 600 and 700°C) and figure 3 shows the SAD patterns taken at various temperatures (325, 350, 363, 600 and 700°C). At 200°C, as shown in figure 2(a), an almost similar morphology was found as in the as-deposited case. This shows that the system is quite stable without any inter-diffusion among the gold nanostructures below 300°C (figures 2(a) and (b)). Figure 2(c) depicts a BF image taken after the system was kept stable at 325°C. Interestingly, the shape of the nanostructure appears to start having definite symmetries (like four-fold) at these low temperatures. During the observation, we noted that some rectangular structures started forming at many places. The real-time SAD pattern at 325°C (i.e. while the system was kept at this temperature) is shown in figure 3(a). The SAD pattern confirms the presence of silicon and unreacted gold in the system. This shows that silicide formation is not a necessary condition to have well-defined and symmetric structures. In-plane diffusion depending on the substrate orientation appears to play a major role in various shape formations of these gold nanostructures. It is well known that bulk Au–Si system has a eutectic phase at 363°C. The temperature was increased by smaller intervals (<25°C) around the bulk eutectic temperature. Figure 2(d) depicts a BF image taken at 350°C. At this temperature, a larger number of nanostructures having rectangular shapes were formed. As the temperature was increased to 363°C and then to 400°C (figures 2(e) and (f)), the coverage area of the rectangular-shaped nanostructures increased. Also some of the structures were of square shape. The SAD pattern taken at 363°C (figure 3(c)) showed signature (compared with the case at 325°C of unreacted gold at this temperature. We did not take SAD patterns after cooling the specimen to RT at this stage and hence it is not possible to comment on eutectic phase formation. It is important to note the studies by Ijima et al. on the dynamic behaviour of ultra-fine gold particles (~2 nm) at atomic resolution using a HRTEM equipped with a real-time video facility [33]. They observed that particles continually changed their shapes, orientations and internal atomic arrangements within a time interval of one-tenth of a second. The in situ diffraction patterns were found to be diffused, giving an impression that the particles were in liquid phase. But, after cooling down to RT, high-resolution TEM showed a regular arrangement of atoms in individual particles [33]. In our studies, at lower temperatures (<363°C), gold nanostructures were relatively stable at these temperatures.
Real-time in situ heating measurements were also carried out at 400 °C (BF image in figure 2(f)), 510 °C (figure 2(g)), 600 °C (figure 2(h)) and 700 °C (figure 2(i)). More aligned nanorectangles/squares were formed at these higher temperatures. The average ratio of length to breadth, known as ‘aspect ratio’, was determined using many BF images taken at each temperature: 325, 350, 363, 400, 510, 600 and 700 °C. In determining the aspect ratios, the longer side of the rectangles was taken as the length. The average length, average aspect ratio and the percentage of nanostructures having square/rectangle shapes at various temperatures are tabulated in table 1. At 325 °C, the aspect ratio was found to be 1.30 ± 0.02 and the average length was 19.5 ± 1.1 nm. Interestingly, even at this stage, about 36% of the nanostructures were having a rectangular shape. As the temperature was increased, the average length and percentage of nanorectangles increased but the average aspect ratio decreased. At 363 °C, the average aspect ratio and length were found to be 1.18 ± 0.01 and 24.3 ± 1.0 nm. At 400 °C, the average length was increased slightly and the aspect ratio was almost the same (as at 363 °C). This indicated that there was a small change in one of the sides. The percentage of nanorectangles was found to be 69% at this temperature. After 510 °C onwards, the average length, average aspect ratio and percentage of particles having rectangle/square shape were almost constant. At 700 °C, the average aspect ratio and length were found to be 1.13 ± 0.02 and 25.9 ± 1.3 nm, respectively, and the percentage of particles having rectangle/square shape was 82%. This is an interesting point of observation. The aspect ratio larger than 1.0 is attributed to some kind of shape transformation. Such shape transformation occurs as a part of release of strain at the interfaces or within the island structures. During the silicide
growth at high temperatures, higher aspect ratio silicide could form so as to reduce the strain energy, which is often explained by a coincidence site lattice matching mechanism. In the present case, the variation in the aspect ratio is about 10–20% over the temperature range, and cannot be attributed to the coincidence site lattice matching mechanism. In the present case, the variation in the aspect ratio is about 10–20% over the temperature range, and cannot be attributed to the coincidence site lattice matching mechanism. From figure 2, one can also observe that initially the gold nanostructures were having very rough edges and corners. With the increase in temperature, smoothening of these edges and corners was observed. Ressel et al. have reported on the behaviour of liquid Au–Si alloys on Si surfaces when ≈2.0 nm of gold was deposited on Si (1 1 1) and Si (1 0 0) substrates at 400 °C under UHV conditions to form droplets [34]. Detailed and elaborate real-time SEM measurements by this group discussed many aspects of wetting phenomena of Au–Si liquid alloys but as the characterization was done using SEM, crystalline quality details of these structures were not discussed [34]. Shape of the droplets changed from circular to hexagonal above 750 °C for the Au/Si (1 1 1). But, in the case of Au/Si (1 0 0) droplets were octagonal when solid and on melting became rounded at higher temperatures. The authors attributed this to the anisotropy in the line tension between solid/liquid/gas lines. With increasing temperature, the entropic effects reduced the free energy difference between straight and curved steps leading to the rounding of the shape [34].

In the following, we discuss the SAD and HRTEM results obtained for a system at different temperatures and at RT (after cooling from 700 °C). Real-time SAD patterns showed the diffraction spots of single crystalline silicon and rings of gold. Apart from this, a few extra spots and rings were also found which match neither with silicon nor gold. Interestingly, from 325 to 350 °C, no extra spots other than the pure silicon spot and gold ring were seen for real-time SAD (figures 3(a) and (b)). After 363 °C, diffuse rings of d-spacing 0.235 nm were observed along with silicon spots (figure 3(c)). This d-spacing value (0.235 nm) matched with both gold and gold silicide phases. At 510 °C, the SAD pattern showed single crystalline spots of silicon and diffuse rings of d-spacing 0.235 nm (figure 3(d)). A more careful analysis revealed that the edges of the nanorectangles were aligned along the $\langle 110 \rangle$ and $\langle 1 \bar{1} 0 \rangle$ direction with respect to the silicon substrate. At 600 °C, few extra reflections other than silicon and gold spots were observed (figure 3(e)) which were found to match with at least three reported phases of gold silicides. The spot nos 1, 2, 3 and 4 correspond to d-spacings of 0.253 nm, 0.453 nm, and 0.235 nm.
0.356 nm and 0.469 nm, respectively. When compared with the reported phases these lattice spacings matched with several phases such as Au$_5$Si$_2$, Au$_5$Si, Au$_7$Si, Au$_2$Si and Au$_3$Si. Similarly at 700 °C, spot nos 1 and 2 and ring no 3 corresponded to a $d$-spacing of about 0.147 nm, 0.261 nm and 0.294 nm, respectively (figure 3(f)). They matched with the phases of Au$_5$Si$_2$ and Au$_5$Si.

Figure 4(b) gives the SAD pattern obtained at RT after cooling the sample from 700 °C. Apart from Si and Au reflections we found a few extra rings which were found to match with at least four reported phases of gold silicide. Ring nos 1, 2 and 3 correspond to a $d$-spacing of about 0.133 nm, 0.156 nm and 0.261 nm, respectively. On comparing them with the reported phases, these spacings match with several phases such as Au$_5$Si, Au$_3$Si, Au$_7$Si and Au$_5$Si$_2$. Many diffraction patterns of gold silicides have been presented over the years [35–40]. Hultman et al rightly pointed out that when comparing the observed data with reported phases of gold silicides, surprisingly there is good agreement between the $d$ values for at least 8–12 strongest reflections of all the alloys [41]. Looking at the constancy of lattice spacing of the strongest lines, they doubt whether these diffraction patterns arise from different phases or they represent various degrees of ordering of superlattices of one fundamental structure, because the fundamental lines are similar for all reported phases except Au$_5$Si$_2$. As in our case, lack of symmetry further adds to the difficulty of assigning one definite phase to the obtained diffraction pattern. It should be noted that the minimum size of
SAD (SAD aperture) available with our system is bigger than a single silicide/island. Hence, the SAD pattern shows these features (representing various reflections). Figures 4(a) and (c) show the BF image of the nanostructures and the high-resolution image of a single nanorectangle. From the BF image one can observe that some Moiré fringes are present on the nanorectangles and that the corners of the rectangles are rounded. Due to the different bonding strengths along two facets the migration barrier experienced by the gold atoms along the length of the rectangle and around the corners differs, which causes the rounding of the corners [42]. In summary, for the 2 nm Au on clean Si (1 0 0) surfaces, ordered and symmetric structures are formed at higher temperatures. The diffraction data suggest the formation of nanosilicide structures at higher temperatures.

We now present the experimental results on the second case where a thin native oxide layer is present at the interface of Si (1 0 0) substrate and 5 nm Au film. Figures 5(a) and (b) show the BF images of the thermally grown as-deposited gold nanostructures for this system and the corresponding SAD pattern. Due to non-wetting nature of gold on native oxide covered silicon substrates, isolated nanostructures of gold are formed on the surface (as shown in figure 5(a)). The average size of these gold nanostructures is 34±1.02 nm and have a coverage area of 24%. The coverage area remained the same throughout the temperature range (up to a temperature of 700 °C) showing no loss of gold from the surface. The nanostructures were also found to be isolated (i.e. non-connected network) in the case of native oxide at the interface [27]. While carrying out the real-time imaging during the annealing procedures on this system (5 nm Au/SiO₂/Si (1 0 0)), no appreciable change in morphology was observed until 750 °C. The native oxide at the interface hinders the inter-diffusion and hence the reaction rate for interacting with bulk silicon is reduced [27]. During MBE growth on clean surfaces, at relatively lower temperatures, the Au–Si interface allows inter-diffusion and in-plane diffusion in the system. When the specimen (the system with an oxide layer) was annealed at 800 °C, agglomeration was observed at some locations. At 850 °C, square/rectangular structures having average size 110 ± 1.3 nm were found along with the unreacted gold particles (figure 5(c)). Contrast between the rectangular structure and some unreacted gold particles can be clearly seen. This indicates the formation of some Au–Si compositions. SAD shows the single crystalline silicon background and some weak polycrystalline rings of Au (figure 5(d)). Desorption of gold from the rectangular gold silicide structure can also be seen at one corner in figure 5(c). A similar structure is shown with a high-resolution STEM in figure 6. STEM images were used to determine the nature of these restructuring structures. The STEM–high-angle annular dark field (HAADF) image shown in figure 6(a) depicts a rectangular gold silicide structure. This shows the expected Z-contrast, so that Au appears brighter than the Si substrate. Interestingly, at one corner we observe a decrease in the contrast value where a rectangular hole-like structure in larger island is observed. This suggests loss of gold atoms at this location. To analyse the local distribution of Au in these structures, STEM–x-ray energy dispersive spectrometry (XEDS) elemental mapping was carried out for the whole structure as shown in figure 6(a). Figure 6(b) shows the Au elemental distribution and figure 6(c) depicts the Si elemental distribution. These maps not only confirm that the island consists of Au, but also that Au depletion takes place at the island edges. The apparent Au depletion in figure 6 can be attributed to the melting induced desorption of gold from these silicide island structures. Figure 7(a) depicts a STEM–HAADF image of one gold silicide structure which developed hole-like structures and (b) is a TEM BF image of the same structure shown in figure 7(a). One of the main issues in this study is to understand where and how the gold silicide desorbs and forms a pit-like structure. As reported by Wang et al [10], the pits can be used to grow different kinds of structures.
Figure 7. (a) STEM–HAADF image on one of the larger gold silicide structures. This shows the formation of many hole-like structures inside (b) the TEM BF image of the same structure shown in (a), and (c) the lattice image taken from the area shown with circles in (b).

Figure 7(c) depicts a high-resolution lattice image taken at 300 kV using a Cs-corrected microscope. It is to be noted that we could not succeed in taking lattice images from the same structure using 200 keV electrons (figures 1–5 are carried out with 200 keV systems) as the structure appeared to be very thick. This demonstrates the importance of a high-energy electron microscope. The lattice spacing values suggest the presence of decomposed (or unreacted) gold (with a lattice spacing of 0.235 nm) and a possible metastable phase of gold silicide (a value of 0.261 nm corresponding to Au5Si2 (1 2 3) plane). Although the lattice image from the centre of the hole-like structure was not clear, fast Fourier transform of the area at the centre showed some crystalline nature. More detailed studies are necessary to understand the evaluation of the hole-like structures.

4. Conclusions

We showed that it is possible to grow symmetric (four-fold) gold silicide nanostructures when grown under ultra-high vacuum conditions and on clean Si (1 0 0) surfaces, and annealed ex situ in a heating stage in TEM even at relatively low temperatures. Real-time high-temperature studies showed evaluation of these structures and the area coverage value and the aspect ratio of these structures were found to increase at higher temperatures. For the system with oxide at the interface, no symmetric structures were found until ≈750 °C. At these higher temperatures, due to desorption of native oxide present at the interface, rectangular structures were formed. We observed desorption of gold and/or gold silicide structures and formation of symmetric hole-like structures in some cases.

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References

[1] Cheung C L, Kurtz A, Park H and Lieber C M 2002 J. Phys. Chem. B 106 2429
[2] Helveg S, Lopez-Cartes C, Sehested J, Hansen P L, Clausen B S, Rostrup-Nielsen J R, Abild-Pedersen F and Norskov J K 2004 Nature 427 426
[3] Dick K A, Deppert K, Karlsson L S, Wallenberg L R, Samuelson L and Seifert W 2005 Adv. Funct. Mater. 15 1603
[4] Hannon J B, Kodambaka S, Ross F M and Tromp R M 2006 Nature 440 69
[5] Moyen E et al 2009 Appl. Phys. Lett. 94 233101
[6] Bezryadin A, Dekker C and Schmid G 1999 Appl. Phys. Lett. 71 1273
[7] Morita T and Lindsay S 2007 J. Am. Chem. Soc. 129 7262
[8] Mativetsky J M, Burke S A, Fostner S and Grutter P 2007 Adv. Funct. Mater. 15 1603
[9] Li B, Luo Z Q, Shi L, Zhou J P, Rabenberg L, Ho P S, Allen R A and Cresswell M W 2009 Nanotechnology 20 085304
[10] Wang H, Zhang Z, Wong L M, Wang S, Wei Z, Li G P, Xing G, Guo D, Wang D and Wu T 2010 ACS Nano 4 2901
[11] Whitesides G M and Grzybowski B 2002 Science 295 2418–21
[12] Hiraki A, Nicolet M A and Mayer J W 1971 Appl. Phys. Lett. 18 178
[13] Narusawa T, Komiya S and Hiraki A 1973 Appl. Phys. Lett. 22 389
[14] Hiraki A, Luguijo E and Mayer J W 1972 J. Appl. Phys. 43 3643
[15] Hiraki A, Shuto K, Kim S, Kammura W and Iwami M 1977 Appl. Phys. Lett. 31 611
[16] Iwami M, Kim S, Kammura W and Hiraki A 1980 Japan. J. Appl. Phys. supplement 19 1489
[17] Mayer J W and Tu K N 1974 J. Vac. Sci. Technol. 11 86
[18] Green A K and Bauer E 1974 J. Appl. Phys. 47 1286
[19] Le Lay G 1983 Surf. Sci. 132 169
[20] Baumann F H and Schreoter W 1988 Phil. Mag. Lett. 57 75
[21] Tersoff J and Tromp R M 1993 Phys. Rev. Lett. 70 2782
[22] Bronzgrsermann S H et al 1998 Phys. Rev. Lett. 80 3795–8
[23] Sekar K, Kuri G, Satyam P V, Sundaravel B, Mahapatra D P and Dev B N 1995 Surf. Sci. 339 96
[24] Sekar K, Kuri G, Satyam P V, Sundaravel B, Mahapatra D P and Dev B N 1995 Phys. Rev. B 51 14330
[25] Mundschau M, Bauer E and Telieps W 1989 Surf. Sci. 213 381–92
[26] Rout B, Sundaravel B, Das A K, Ghose S K, Sekar K and Mahapatra D P 2000 J. Vac. Sci. Technol. B 18 1847
[27] Bhatta U M, Dash J K, Roy A, Rath A and Satyam P V 2009 J. Phys.: Condens. Matter 21 205403
[28] Bhatta U M, Dash J K, Rath A and Satyam P V 2009 Appl. Surf. Sci. 256 567
[29] Bhatta U M, Rath A, Dash J K, Ghatak J, Yi-Feng L, Liu C P and Satyam P V 2009 Nanotechnology 20 465601
[30] Frank S, Wedler H and Behm R J 2002 Phys. Rev. B 66 155435
[31] Vasishth S and Shirokoff J 2010 Appl. Surf. Sci. 256 4915–23
[32] Goswami D K, Satpati B, Satyam P V and Dev B N 2003 Curr. Sci. 84 903
[33] Ijima S and Ichihashi T 1986 Phys. Rev. Lett. 56 616
[34] Ressel B, Prince K C, Homma Y and Heun S 2003 J. Appl. Phys. 93 3886
[35] Suryanarayana C and Anantharaman T R 1974 Mater. Sci. Eng. 13 73
[36] Anantharaman T R, Luo H L and Klement W 1966 Nature 210 1040
[37] Gaigher H L and Vander Berg N G 1980 Thin Solid Films 68 373
[38] Green A K and Bauer E 1976 J. Appl. Phys. 47 1284
[39] Tsaur B Y and Mayer J W 1981 Phil. Mag. A 43 345
[40] Au5Si (JCPDS 26-0723), Au5Si (JCPDS 26-0725), Au5Si2 (JCPDS 36-0938), Au5Si (JCPDS 40-1140), Au5Si (JCPDS 24-0463)
[41] Hultman L, Robertson A, Hentzell T G, Enstroem I and Psaras P A 1987 J. Appl. Phys. 62 3647
[42] McCarthy D N and Brown S A 2008 J. Phys.: Conf. Ser. 100 072007