Morphological evolution of epitaxial cobalt–semiconductor compound layers during growth in a scanning tunneling microscope

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(Received 13 November 2001; accepted 13 May 2002)

We investigate the mechanisms of CoSi₂ and CoGe₂ growth by carefully controlled e-beam evaporation of Co onto Si(001) and Ge/Si(001) substrates from the very initial submonolayer stage, monitored in situ by scanning tunneling microscopy, aided by reflection high-energy electron diffraction. In order to grow different epilayer morphologies, we use flat and vicinal surfaces and two different methods of synthesis: reactive deposition (where Co is deposited onto a hot substrate), and solid-phase reaction (where Co is deposited at lower, or room, temperature). We attempt to account for the observed morphological differences in the epilayers by correlating them with parametric differences in the deposition and growth processes. © 2002 American Vacuum Society. [DOI: 10.1116/1.1491555]

I. INTRODUCTION

Cobalt disilicide (CoSi₂, known mostly as a good self-aligned salicide) has a low mismatch with silicon. Yet, it does not grow as a monocrystalline twodimensional layer on the Si(001) surface, rather it forms misoriented threedimensional islands. CoGe₂ is another interesting metal–semiconductor compound that can be used, for example, as a contact to SiGe alloys and GaAs. CoGe₂ forms three-dimensional islands on Ge/Si(001) that are very similar to the CoSi₂ ones on Si(001).

In this work we try to shed light on the factors that affect surface morphology of ultrathin heteroepitaxial cobalt–semiconductor compound layers in relation to the kinetics of phase formation in those layers. We propose an explanation which relates the layer’s surface morphology to the layer–substrate mismatch as a function of a particular sequence of kinetically driven phase-formation processes, and accounts for our experimental observations.

II. EXPERIMENTAL METHODS

The experimental procedures and methods used in this work closely resemble those used in our previous work (e.g., see Refs. 5 and 7). The experiments were performed in a JEOL ultrahigh vacuum (UHV) elevated-temperature scanning tunneling microscope (STM), equipped with reflection high energy electron diffraction (RHEED) and a low energy electron diffraction (LEED)/Auger spectrometer, capable of operating up to 1250 °C by resistive heating. Si(001) wafers were chemically treated ex vacuo by a repeated etch-and-regrowth procedure to produce a clean homogeneous oxide. In UHV (base pressure 1×10⁻¹⁰ Pa), after thorough degassing this oxide was evaporated by repeated flashes at 1200 °C, and the clean Si surface was left to order during slow cooling to the desired temperature. During the sample flashes and anneals, the pressure was kept below 10⁻⁷ Pa. This kind of treatment has generally proved effective in producing a well-ordered (2×1) surface, as was indeed verified this time by LEED and, after transferring the sample into the STM chamber, by RHEED and by STM.

At this point Co from a water-cooled four-element e-beam source inclined 45° to the sample’s surface was deposited at about 10⁻⁷ Pa in two different sets of silicide growth experiments:

1. reactive deposition epitaxy (RDE) onto that Si(001) surface preheated to 500 °C, and
2. solid-phase epitaxy (SPE) with deposition at room temperature (RT), and a subsequent series of anneals to promote the silicide reaction.

Modifications to this surface due to exposure to Co atoms were monitored in situ, either by STM (in constant-current mode using electrochemically etched W tips, described elsewhere), or by RHEED in (100) and (110) azimuths.

For the cobalt germanide growth, a flat pseudomorphic Ge layer was grown first by gas-source molecular beam epitaxy (GSMBE) from GeH₄ to the maximal wetting layer thickness of about seven monolayers (7 ML), below the critical thickness of the Stranski–Krastanow transition. Then Co was deposited onto this Ge/Si(001) surface and monitored in a similar way to the silicide growth procedure, i.e.,
3. RDE onto the Ge/Si(001) surface preheated to 500 °C, and
4. SPE with deposition at 230 °C, with a subsequent series of anneals to promote the germanide reaction.

Submonolayer Co coverages (<1 ML) were deposited in the RDE and 2–3 ML in the SPE processes for the both silicide and germanide growth experiments.

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III. EXPERIMENTAL RESULTS

A. Reactive deposition epitaxy of Co/Si(001) and CoGe/Si(001)

The evolution of the growing silicide and of germanide surfaces during the RDE process can be seen and compared in Figs. 1(a)–1(d) and 1(e)–1(l), respectively. Even the slightest exposure of the initially (2×1)-reconstructed Si(001) surface to Co atoms causes the formation of dimer vacancy lines (DVLs), which modifies the reconstruction to (2×N),3,12 as shown in Fig. 1(a). The pregrown Ge layer, shown in Fig. 1(f), with (M×N) quasiperiodicity9–11 on the initially (2×1)-reconstructed Si(001) surface [Fig. 1(e)], served as a substrate for germanide growth.7,12

Reaction of the evaporated Co adatoms with the Si(001) and Ge/Si(001) surfaces at 500 °C immediately resulted in three-dimensional (3D) nanocrystals, as can be seen in Figs. 1(b) and 1(c) and 1(g)–1(l), respectively [note the different scale bars in Figs. 1(e) and 1(k)–1(l)]. For the completeness of the discussion, it should be mentioned that a more two-dimensional (2D) RDE cobalt silicide growth, in the form of rectangular platelets shown in Fig. 1(d), is also possible, if the growth takes place on vicinal Si(001) surfaces.5 Also, small numbers of 2D silicide islands, e.g., two in the center of Fig. 1(b), can be noticed to form on the singular, i.e., flat, surface as well. However only very few of these are formed, and the vast majority of the silicide islands resulting from RDE are three-dimensional. Careful analysis of atomic-resolution STM micrographs [cf. Fig. 1(c)] and RHEED patterns in several azimuths (cf. Fig. 1 insets) has lead us to identify the hut-shaped Co–Si nanocrystals as [221]-oriented cobalt disilicide (fluorite-type CoSi2) islands,5 and elongated CoGe2 on top of pyramidal Ge islands.7

B. Solid phase epitaxy of Co/Si(001) and Co/Ge/Si(001)

In the next set of experiments a clean Si(001)-(2×1) surface [Fig. 2(a)] was covered with Co [Fig. 2(b)] at RT, and then annealed at 420 °C [Fig. 2(c)], 465 °C [Fig. 2(d)], 500 and 550 °C [Fig. 2(e)], and 700 °C [Figs. 2(f)–2(i)]. In the last set of experiments, the germanide/Si(001) surface
from the previous RDE experiment [Fig. 2(j)] was covered with Co [Fig. 2(k)] and annealed at various temperatures and times. Interesting morphological changes took place after 1 [Fig. 2(l)] and 15 h [Fig. 2(m)] anneals at 400 °C. The anneals were conducted until the formation of disilicide and digermanide phases.

The first indication of disilicide (CoSi$_2$) formation appeared after a 550 °C anneal [Fig. 2(e)], as seen by the appearance of (2×2) reconstruction of the (111)-CoSi$_2$ surface in STM images [magnified in the bottom-left inset of Fig. 2(e)]. However at this stage the diffraction pattern was not yet fully characteristic of disilicide: the very faint RHEED streak arching [top-right inset of Fig. 2(e)] related to disilicide formation$^5$ is indicative of only partial transformation. The disilicide fluorite lattice was unambiguously identified by STM observations of the well known $\sqrt{2}\times\sqrt{2}$-R45° reconstruction [cf. bottom-left insets of Fig. 2(f)] and RHEED patterns in Figs. 2(h) and 2(i) and top-right inset of Fig. 2(f), only after a 700 °C anneal. Hence while ordering of the initially disordered Co/Si(001) layer [Fig. 2(b)] from RT deposition onto the Si(001)-(2×1) surface [Fig. 2(a)] had already begun at 420 °C, as reflected by modification of the corresponding RHEED pattern [top-right inset of Fig. 2(c)] and STM detection of small flat ordered patches [bottom-left inset of Fig. 2(c)], CoSi$_2$ formation was evidently completed only at 700 °C.
Some ordering of the initially disordered Co/Ge/Si(001) layer [Fig. 2(k)] took place even below 400 °C, however no appreciable change was detected until the 400 °C anneal [Fig. 2(l)]. The Co–Ge reaction seems to have again produced two types of nanocrystal, Ge pyramids being one of them, like in the RDE of Co/Ge/Si(001) described in Sec. III A. However this time quite symmetrical, rather than elongated, equiaxed crystallites were located at the troughs (rather than at the crests) of the faceted Ge pyramids, as shown in Fig. 2(l) [cf. Figs. 1(g)–1(l)]. Further annealing caused yet another morphological change: only one type of elongated, hut-shaped nanocrystals could be detected, as shown in Fig. 2(m). The disappearance of the Ge pyramids was also observed in the RHEED patterns by the disappearance of the chevron-like features in Figs. 2(n) and 2(o). We identified the equiaxed nanocrystals at the Ge-pyramid troughs as Co$_5$Ge$_2$, and the elongated ones at the Ge-pyramid crests as CoGe$_2$ phases, respectively.

IV. DISCUSSION

The observed morphological differences between the surfaces of Co–Si and Co–Ge reaction products can be considered in light of the silicide and germanide phase-formation kinetics as a function of the deposition and growth processes and elasticity theory. The layer morphology is driven by a kinetics as a function of the deposition and growth processes considered in light of the silicide and germanide phase-formation.

In addition to five stable phases, tetragonal Co$_3$Si ($a = 8.42$ Å, $c = 5.81$ Å), orthorhombic $\alpha$-Co$_5$Si ($a = 4.92$ Å, $b = 3.74$ Å, $c = 7.11$ Å) and tetragonal $\beta$-Co$_5$Si, cubic CoSi ($a = 4.44$ Å) and cubic CoSi$_2$ with a CaF$_2$ structure ($a = 5.37$ Å) in the equilibrium Co–Si phase diagram, four metastable phases have also been observed, hexagonal Co$_3$Si ($a = 4.98$ Å, $c = 4.07$ Å), Co$_5$Si, orthorhombic $\gamma$-Co$_5$Si, and tetragonal Co$_5$Si$_{13,14}$ However, with the exception of a new metastable cubic CoSi$_2$ phase with a CsCl structure that was found by von Känel et al. in epitaxially grown layers, only $\alpha$-Co$_5$Si, cubic CoSi, and cubic CoSi$_2$ (CaF$_2$) have ever been detected in thin films. In the equilibrium Co–Ge bulk phase diagram contains seven phases, cubic Co$_3$Ge, hexagonal Co$_5$Ge$_2$ ($a = 3.93$ Å, $c = 5.01$ Å), hexagonal $\beta$-Co$_5$Ge$_2$, and orthorhombic $\alpha$-Co$_5$Ge$_2$, monoclinic ($a = 11.65$ Å, $b = 3.81$ Å, $c = 4.95$ Å) and cubic ($a = 4.64$ Å) CoGe, tetragonal Co$_5$Ge$_2$ ($a = 7.64$ Å, $c = 5.81$ Å), and orthorhombic Co$_5$Ge$_2$ ($a = b = 5.68$ Å, $c = 10.82$ Å); Pearson’s crystallographic data also contains orthorhombic Co$_3$Ge ($a = 5.02$ Å, $b = 3.82$ Å, $c = 7.26$ Å). However in thin films the only phases to have ever been observed to form were CoGe, Co$_3$Ge$_7$, and CoGe$_2$.  

A. Reactive deposition epitaxy of CoSi(001) and Co/Ge/Si(001)

Slow metal deposition onto a substrate at high temperature ensures a low effective metal concentration at the growth interface, i.e., every metal atom arriving at the surface reacts immediately, and hence there is no metal accumulation. Consequently, the first phase formed is the one that is poorest in metal. In metal–silicon systems, this is usually metal disilicide, MeSi$_2$, which is the last phase in the equilibrium diagram, while the intermediate metal-rich phases, such as Me$_2$Si and MeSi, are skipped. Immediate and direct formation of disilicides in RDE was experimentally confirmed for Co–Si, Ni–Si and Fe–Si systems, in accord with our identification of the hut-shaped nanocrystals in Figs. 1(b) and 1(c) as CoSi$_2$. However there is strain in the CoSi$_2$ layer due to its mismatch with the underlying Si lattice. Linear mismatch strain is defined as

$$e = \frac{a_{\text{layer}} - a_{\text{subs}}}{a_{\text{subs}}},$$

where $a$ is the surface lattice constant. Accordingly, linear mismatch between CoSi$_2$ ($a_{\text{layer}} = 5.37$ Å) and Si ($a_{\text{subs}} = 5.43$ Å) is $e \approx -1.10\%$, i.e., CoSi$_2$ is under 1.10% tension. Epitaxial systems that deviate from cubic symmetry are more appropriately described by the areal lattice mismatch:

$$e_A = \frac{a_{1}^{\text{layer}} - a_{1}^{\text{subs}}}{a_{1}^{\text{layer}}} - \frac{a_{2}^{\text{layer}} - a_{2}^{\text{subs}}}{a_{2}^{\text{layer}}},$$

where $a_1$ are the surface unit-cell base vectors. According to this definition CoSi$_2$ experiences $-2.20\%$ tensile biaxial strain. Using finite element analysis, we have shown that pyramidal and hut shapes are effective in relieving that strain, thus accounting for the shapes of three-dimensional CoSi$_2$ nanocrystals observed.

Applying the same reasoning to RDE of Co onto the Ge/Si(001) layer supports our conclusion of the immediate formation of cobalt digermanide, CoGe$_2$, however it does not explain the presence of two nanocrystal types [Figs. 1(g)–1(l)], or their striking relative disposition (i.e., one on top of another). This phenomenon can be accounted for, on the other hand, by using arguments based on elasticity, i.e., mismatch strains due to differences in respective layer–substrate lattice constants, and their relaxation. Essentially, in a (001) CoGe$_2$[001] Ge and [100] CoGe$_2$[100] Ge orientation, the CoGe$_2$ layer on Ge experiences 0.35% biaxial compression or, in effect, even higher (up to 9.42%) because the Ge layer itself is fully strained to the Si lattice constant ($a_{\text{Si}} = 5.43$ Å) at least at the Ge/Si(001) interface. This strain is sufficient to drive the nucleation of 3D CoGe$_2$ nanocrystals. However this also means a strain gradient across the Ge layer of between nominally 0.35%–9.42% biaxial tension (or even higher if the $c$-variant (100) CoGe$_2$[001] Ge and [001] CoGe$_2$[100] with $c_{\text{CoGe}_2} = 10.82$ Å stretched over two Ge unit cells is present in the film) at the upper CoGe$_2$/Ge interface, and 9.42% biaxial compression at the lower Ge/Si(001) interface. This strain gradient may well suffice to drive the 2D-to-3D transition of the initially flat, pseudomorphic Ge layer, and allow elastic strain relaxation at the 3D Ge island crests. The relative location of the elongated CoGe$_2$ and pyramidal Ge nanocrystals can also be understood within the framework of strain distribution and relaxation in the layers. As CoGe$_2$ possesses the larger unit cell of...
the two, the most “comfortable” location for it would be at the maximally relaxed regions in the vicinity of the pyramid or hut crests. Hence it seems that, at first, strained CoGe₂ nanocrystals form, and in doing this apply additional strain on the Ge wetting layer, lift its pseudomorphism and “pull out” Ge nanocrystals.

B. Solid phase epitaxy of Co/Si(001) and Co/Ge/Si(001)

In the case of solid-phase reaction, where the effective metal concentration at the growth interface is large and/or the substrate temperature is not high enough for an instantaneous reaction, the first phases formed will be richer in metal. In the Co–Si system the sequence of phase formation under SPE conditions is Co₂Si, CoSi, and only then CoSi₂. At least one of the metal-rich phases has to precede CoSi₂, however not necessarily both Co₂Si and CoSi appear in the reaction sequence. As was pointed out by Horsfield and Fujitani, all three phases have very similar formation energies, and hence the order in which they appear must depend entirely on the reaction kinetics and on other factors, such as the reduction of interface energy by a phase that is better matched with silicon. These authors have also proposed a model to explain why at times only one of the two intermediate phases forms. According to their model, first a metastable Co–Si structure is formed, and it can transform into Co₂Si by absorbing more Co atoms, or collapse into CoSi by shear, if there are no more Co atoms available.

As noted in Sec. III B, the silicide-forming reaction had already started at about 420 °C, with CoSi₂ beginning to form only at about 550 °C and transformation undoubtedly completed at 700 °C. We believe that Co₂Si was the first phase to form at 420 °C, in excellent agreement with Ref. 17. Consumption of Si atoms from the substrate for the reaction resulted in deep holes in the substrate, e.g., these can be seen in the upper-left corner of Fig. 2. The consumption of Si atoms from the substrate for the reaction resulted in deep holes in the substrate, e.g., these can be seen in the inset of Fig. 2 (a). Nonetheless, this phase can grow coherently on Si, e.g., in a (100) Co₂Si(001) Si and [010] Co₂Si[110] Si orientation, with −9.83% biaxial strain in the Co₂Si film. Since the CoSi mismatch with Si is far too great for coherent epitaxial growth on Si, a Co₂Si→CoSi transformation is likely to introduce misfit dislocations at the interface and even cracks. Hence the formation of deep cracks that extend throughout the film, as can be seen in the inset of Fig. 2 (d) after annealing at a higher temperature of 465 °C, may indicate the formation of CoSi. Since the silicide layer is now at least partially relaxed, the 2D morphology of the sequentially formed Co₆Si₂ [Figs. 2(e)–2(g)] after higher-temperature anneals is not surprising. Another consequence of that relaxation is the formation of misoriented grains, e.g., (111) shown in Fig. 2(e), (001) in Fig. 2(f), and possibly (110) and other orientations, as well as massive faceting (not shown). In-plane misorientation can also be seen, for example, at a (001) Co₆Si₂ surface, an example of which is seen in the bottom insets of Fig. 2(f): clearly in coherent epitaxial growth, a pseudomorphic Co₆Si₂ layer grows on top of Si in cube-on-cube fashion, with its (100) directions parallel to those of Si, whereas the insets exhibit two variants of the 2 × 2 - R45° reconstruction with their unit-cell sides parallel to both (100) (bottom-right inset) and (110) (bottom-left inset) Si directions. Figure 2(g) shows an interestingly reconstructed surface that has never been reported in the literature so far. The exact nature of this surface is beyond the scope of this article, but there are two interpenetrating surface structures, possibly a result of phase separation, that can be described as a mixture of the Co₆Si₂ (001) c(2×4) [different from the c(2×4) reconstruction described in our previous work] and (2 × 3). On the other hand, because of the complex nature of reconstructing sur-

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**Fig. 3.** (a) Intensity profiles along [100] and (b) [110] RHEED patterns as a function of the annealing temperature. In each case the topmost profile and the one beneath are from the epitaxial SiC and the Si substrate, respectively, used for measurement calibration, and the next ones (going from top to bottom) are in increasing order of the annealing temperature from RT Co/ Si(001) up to a 700 °C anneal and corresponding CoSi₂ formation.
faces, and the fact that the distances measured are integral multiples of lattice vectors, other reconstructions on differently oriented surfaces cannot be excluded. Figures 2(h) and 2(i) show RHEED patterns in [100] and [110] azimuths, respectively, which help to identify the surface after 700 °C annealing as CoSi2.

The explanation we propose for the SPE of Co–Ge is very similar to that for RDE in Sec. IV A. When the Ge/Si(001) layer [Fig. 2(j)] reacts with the Co adatoms deposited [Fig. 2(k)] and produces Co5Ge7 at a 400 °C anneal [Fig. 2(l)], the resulting strain distribution in the system can explain the striking morphology seen in Fig. 2(l). Epitaxial Co5Ge7 with its (001) surface parallel to the Ge(001) surface and [100] direction parallel to one of Ge ⟨110⟩ directions (periodicity along the [100] Co5Ge7 crystallographic direction doubles the Ge periodicity along its ⟨110⟩ directions) would exert 9.25% biaxial compression (or even more if the Ge lattice constant is strained to a value close to that of Si) on the underlying Ge layer. Again, this would result in the formation of Co5Ge7 and Ge nanocrystals in a manner similar to the formation of germanium and germanide nanocrystals in RDE, i.e., compression exerted by the Co5Ge7 nanocrystals on the flat Ge layer pulls out Ge pyramids and huts. However, because the Co5Ge7 lattice constants are smaller than those of Ge, this time the most “comfortable” location would correspond to the more compressed regions at the Ge pyramid and hut troughs, creating a “raisin-cake” appearance, as shown in Fig. 2(l). Further annealing caused the Co5Ge7 to transform into the final, CoGe2, phase, in the shape of elongated 3D nanocrystals, as shown in Fig. 2(m). Ge pyramids were no longer seen anywhere, most probably because all the Ge was consumed by that time and provided...
enough Ge for the Ge-rich CoGe$_2$. Their disappearance was reflected well in the corresponding RHEED patterns [Figs. 2(n) and 2(o)], especially along the $\langle 110 \rangle$ azimuth$^7$ [Fig. 2(o)].

To support our analysis, which is largely based on the presence or absence of strain in the films, we have attempted to estimate strain by measuring the interstreak distance in the RHEED patterns of the evolving phases. In the both RDE and SPE syntheses of Co–Ge the interstreak distance never changed at any stage of the process, indicating that all the germanide phases that formed in both processes remained fully strained and coherent with the substrate. On the other hand, while according to RHEED the RDE-formed CoSi$_2$ was coherent with the Si substrate,$^5$ the relaxation assumed to be due to the presence of cracks in the film prior to the formation of CoSi$_2$ in SPE was sufficiently tangible to show up in RHEED measurements. Figure 3 shows the evolution of RHEED patterns acquired in $\langle 100 \rangle$ and $\langle 110 \rangle$ azimuths [Figs. 3(a) and 3(b), respectively] with the progression of anneals. The two uppermost profiles are from epitaxial SiC and clean Si, used for calibration, and the rest are in order of increasing annealing temperature, with the lower corresponding to a prolonged anneal at 700$^\circ$C. The increasing interstreak distance in the $\langle 100 \rangle$ pattern indicates a decrease in interatomic spacing in the real lattice, i.e., relaxation of tensile strain in the CoSi$_2$/Si(001) layer. The relaxation calculated amounted to about 2%, indicating even some degree of overrelaxation. However, the value measured is too close to the measurement error to draw more far-reaching conclusions. It is also interesting to note that no relaxation was observed in the $\langle 110 \rangle$ azimuth [Fig. 3(b)], and perhaps this is evidence of some sort of anisotropic relaxation along the preferred $\langle 100 \rangle$ azimuth only.

V. SUMMARY

In this work we have attempted to identify the kinetic pathways to phase formation in Co–Si and Co–Ge ultrathin layer systems as a function of the deposition method (reactive deposition and solid-phase epitaxy) and to correlate those pathways with the morphologies observed. Figure 4 gives a summary of our conclusions on the phase-formation sequence.

The reactive deposition epitaxy resulted in the immediate formation of cobalt disilicide [CoSi$_2$, RDE arrow in Fig. 4(a)] and cobalt digermanide [CoGe$_2$, RDE arrow in Fig. 4(b)]. Both the disilicide and the digermanide layers retained full coherency with the substrate, and if some degree of relaxation took place it was beyond the sensitivity of our RHEED measurements, which indicated that both layers were fully strained. The tendency to relax mismatch strain in the layer caused three-dimensional disilicide and digermanide nanocrystals to form. The Co–Ge reaction represents a special case, because the Co was not reacted with bulk Ge, but rather with a thin intermediate Ge/Si(001) layer, itself strained at the Ge/Si interface. The additional strain exerted by the growing CoGe$_2$ on the initially two-dimensional pseudomorphic Ge layer lifted the two-dimensional metastability and caused three-dimensional Ge pyramids and huts to form instead underneath each emerging CoGe$_2$ nanocrystal. The Ge-pyramid crests represent elastically comfortable sites for CoGe$_2$, because the Ge lattice dilatation there helps to accommodate the CoGe$_2$ phase, which has a lattice constant that is slightly larger than Ge (0.35% biaxial mismatch). No intermediate phases were detected. These results are in good agreement with the elasticity theory that accounts for strain relaxation at the crests of three-dimensional islands.

On the other hand, intermediate phases richer in cobalt did form during solid-phase reaction. Co$_2$Si and CoSi could have been expected to form in the Co–Si reaction [SPE pathway in Fig. 4(a)] and Co$_2$Ge, CoGe, and Co$_5$Ge$_7$ phases in the Co–Ge reaction [SPE pathways, shown by broken arrows in Fig. 4(b)]. We argue that in the Co–Ge reaction the first phase to form was Co$_2$Ge$_7$, indicated by the solid arrow in Fig. 4(b), with CoGe and Co$_2$Ge skipped. This is supported by the fact that Co$_5$Ge$_7$ and even CoGe$_2$ (into which Co$_2$Ge$_7$ had eventually transformed) remained fully coherent with the substrate and strained, at least to the detection limit of RHEED, which would not have been the case had Co$_2$Ge with $\sim$13.77% biaxial mismatch, or unmatched CoGe, been formed beforehand. Again, nucleating Co$_5$Ge$_7$ nanocrystals exerted additional strain on the already strained Ge intermediate layer, and pulled out Ge pyramids and huts. However, this time elasticity considerations favored the more compressed regions at the troughs associated with Ge pyramids and huts as preferred sites for accommodating Co$_5$Ge$_7$ nanocrystal, since the Co$_5$Ge$_7$ phase exhibits a lattice constant smaller than Ge. In the Co–Si reaction both Co$_2$Si and CoSi formation, whether sequential or simultaneous, are possible. We believe that the silicide layer could have maintained the epitaxial coherency strains had CoSi not been formed, since Co$_2$Si exerts only about $\sim$9.83% biaxial strain. The CoSi lattice cannot commensurately match the Si one, and its formation can account for the relaxation measured quantitatively at least along $\langle 100 \rangle$ RHEED azimuths, and indirectly observed by STM. Again, similar to the RDE experiments, the SPE results conform to the elasticity arguments.
