RPCVD growth of isolated Ge crystals and suspended layers on micrometric Si pillars

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ABSTRACT

In this work, we demonstrate the growth of Ge crystals and suspended continuous layers on Si(001) substrates deeply patterned in high aspect-ratio pillars. The material deposition was carried out in a commercial reduced-pressure chemical vapor deposition (RPCVD) reactor, thus extending the "vertical-heteroepitaxy" technique developed by using the peculiar Low-energy plasma-enhanced chemical vapor deposition reactor (LEPECVD), to widely available epitaxial tools. The growth process was thoroughly analyzed, from the formation of small initial seeds to the final coalescence into a continuous suspended layer, by means of Scanning- and Transmission electron microscopy, x-ray diffraction, and μ-Raman spectroscopy. The pre-oxidation of the Si pillars sidewalls and the addition of hydrochloric gas in the reactants proved to be key to achieve highly selective Ge growth on the pillars top only, which, in turn, is needed to promote the formation of a continuous Ge layer.

Thanks to continuum growth models we were able to single out the different roles played by thermodynamics and kinetics in the deposition dynamics.

We believe that our findings will open the way to the low-cost realization of tens of μm thick heteroepitaxial layer (e.g. Ge, SiC, GaAs) on Si having high crystal quality.

1. INTRODUCTION

The vertical heteroepitaxy (VHE) of Ge crystals on Si substrates patterned in μm-sized, high aspect ratio Si pillars, was recently demonstrated by using a Low-energy plasma-enhanced chemical vapor deposition (LEPECVD) technique[1, 2]. By exploiting strong out-of-equilibrium conditions, obtained through high deposition rates (4 nm/s) and relatively moderate temperatures (∼500°C), the VHE allows for the growth of high quality Ge crystals featuring several μm heights. These heterostructures can be of interest in view of possible applications because of some superior properties of Ge with respect to those obtained in deposition on bare Si(001) wafers. In fact, the high aspect ratio of the VHE Ge crystal allows for the almost complete relaxation of the thermal stress[3], thus avoiding wafer bending and cracking. Moreover, threading dislocations (TD) in the active (top) part of the Ge crystal can be completely eliminated either by lateral expulsion (as in aspect-ratio trapping[4]) or by bending produced by faceted growth[5]. Very recently, the possibility to eliminate dislocations by a suitable compositional grading of the VHE has also been demonstrated[6, 7].

These features, together with the "self-pixelation" induced by the ordered array of Si pillars, open interesting technological perspectives, e.g. in the development of next-generation infrared and x-ray detectors[8, 9]. On the other hand, other applications (e.g. in microelectronics) require continuous two-dimensional (2D) heterolayers. In order to respond also to these needs, it was recently shown that it is possible to trigger coalescence of adjacent VHE pillars into a continuum film by post-growth annealing[10], or by suitably modifying the growth conditions[11]. This leads to μm-thick Ge films, suspended on the Si pillar array underneath. If the film is divided in 100-300 μm-wide patches, rather than covering the whole wafer, a significant thermal strain relaxation of the Ge layer is provided by the tilting of the Si pillars[12].

So far, VHE of Ge on micron-sized pillars was achieved only using a LEPECVD reactor. The peculiarity of the technique has surely limited the popularity of the approach, slowing down research in this field.

In this work, we show that growth of isolated Ge crystals and of a suspended Ge layer on µm-sized Si pillars can also be achieved using a commercial reduced-pressure CVD (RPCVD) reactor, by exploiting a newly developed deposition procedure, involving pre-oxidation of the Si pillar sidewalls and the addition of HCl in the reactant mixture. These two factors promote the selective growth of Ge on the pillars top, a key feature for the subsequent development of µm-sized Ge faceted crystals, and, eventually, to the formation of a continuous, suspended Ge film. A close comparison between experimental results and theoretical simulations was carried out, allowing us to thoroughly describe the growth dynamics at each stage of the growth and to single out the different contributions of thermodynamics and kinetics.

The successful realization of the VHE using RPCVD, the workhorse of microelectronic industry, opens new routes to its exploitation in the field of substrate engineering, including the much sought after heterointegration of III-V materials on Si, as well as Ge integration on the mainstreamed Si platform for various applications like e.g. high-mobility CMOS transistors[13], memories[14], thermoelectric[15], solar cells[16], and monolithic integration of photonics with CMOS technology[17].

2. EXPERIMENTAL SECTION

Sample preparation. Si(001) 200mm wafers were covered with a hardmask, consisting of 10 nm thermal silicon oxide (SiO_2) and 150 nm silicon nitride (Si_3N_4) deposited by a low-pressure chemical vapor deposition (LPCVD) batch reactor. Lithographical patterning by an i-liner process, reactive ion-etching (RIE) and pulsed RIE etching (Bosch process) were used to etch deep into the Si wafer.

We realized different "block" arrays (e.g. 200×200 μm²) of 1×1 μm² (SUB1) and 2×2 μm² (SUB2) Si pillars, 8 μm tall, with 1 μm wide spacing. After resist ashing and Piranha-cleaning, the sidewalls of the created Si pillars were oxidized by 100 nm thermal SiO_2 using LPCVD. Finally, the hardmask on top of the Si pillars was removed by wet chemistry using hydrofluoric acid and phosphoric acid as etchants. After wet chemistry, the sidewalls of the Si pillars were still covered with 80 nm-thick SiO_2-layer while the top facet was left oxide free. Typical results are displayed in Fig. 1(a, b).

The selective Ge growth was carried out in an ASM Epsilon 2000 RPCVD single wafer system. Prior to the growth, the samples were cleaned in standard RCA solution combined with a HF dip last clean. Afterwards, the samples were immediately loaded into the RPCVD growth chamber and baked out at 1000°C for 30 sec in order to remove remaining native oxide and moisture.

Experimental characterization. For analysis of the surface topography, we used a Zeiss Merlin scanning electron microscope (SEM) operating at 1.5 kV and a Bruker Innova Atomic force microscope (AFM), working in tapping mode.

The structural quality was investigated by Transmission electron microscopy (TEM) using a FEI Tecnai Osiris operating at 200 kV. TEM lamellas were prepared by milling and undercutting processes using a Zeiss NVision 40 Focused ion beam operating at up to 30 kV.

X-ray diffraction (XRD) measurements were performed using a SmartLab diffractometer from Rigaku equipped with a 9 kW rotating anode Cu source (Cu-K_α = 0.1541 nm). μ-Raman measurements were carried out using a Renishaw inVia microscope in backscattering geometry with 633 nm Helium-Neon red laser, 1800 lines/mm grating and x50 objective with numerical aperture of 0.75, which results in a laser spot size of ≈ 520 nm in diameter and a Ge penetration

depth of ≈ 32 nm. Each scan was calibrated by the position of the Si-Si vibration of Si(001) bulk crystal references set to 520.7 cm^{-1}.

Phase-field simulations. A phase-field approach is exploited[18], allowing for a natural description of the complex dynamics of coalescence observed between the pristine islands in the early stages of the growth. This is possible thanks to the implicit description of the crystal surfaces, based on the phase-field function φ, assuming value 1 within the solid and 0 in the vacuum region. A smooth transition at the interfaces between these two phases, provides a diffused description of the free surface, nominally localized at the φ=0.5 isoline. The evolution of the growth front is described within a simple evaporation/condensation model by solving the Allen-Cahn equation $\partial\varphi/\partial t = -\delta F/\delta\varphi$. The free energy F takes the form of a Ginzburg-Landau functional[18]

$$F = \int_V \gamma(\hat{\boldsymbol{n}}) \left[\frac{\epsilon}{2}|\nabla\varphi|^2 + \frac{1}{\epsilon}W(\varphi)\right] + \frac{1}{\epsilon}\mathrm{S}(\varphi)\, d\boldsymbol{x}$$

with ϵ a measure of the width of the diffused interface and γ the surface energy density, eventually dependent on the local profile orientation $\hat{\boldsymbol{n}}$. A double-well potential $W(\varphi) = 18\varphi^2(1-\varphi)^2$ is present to enforce the stability of the solid and vacuum phases, i.e. the minima of W. The difference in energy between these two bulk phases $\Delta\mu^0 = \mu_v^0 - \mu_s^0$ is accounted by the last term $S(\varphi) = \Delta\mu^0[1-\varphi^3(6\varphi^2 - 15\varphi + 10)]$. Here it is assumed that only surface energy contributes to the chemical potential. Indeed, the elastic contribution due to the Ge/Si lattice misfit is neglected by considering as initial state islands large enough to be plastically relaxed ***(see Results and Discussion part below)***. As bulk diffusion is not active at the present growth temperature[19], we also neglected Si/Ge intermixing, expected only at the very early stages of the island formation, possibly proceeding our initial state. Surface anisotropy is

included by considering a continuous surface energy density function $\gamma(\hat{\boldsymbol{n}})$ as in Ref[20], with minima along the orientations of the typical facets of Ge, i.e. {001}, {113}, {111}, {110}, and the values are taken from the literature[21] following the procedure reported in Ref.[22]. The partial differential equations of the model are numerically solved by Finite Element Method (FEM), exploiting the AMDiS toolbox[23, 24]. In particular, a semi-implicit integration scheme and space-adaptive meshing are considered for efficiency. To reduce the computational cost further, we simulated a single pillar only with mirror boundary conditions. The initial profile consists of a distribution of 50 half-ellipsoidal islands, with random positions and axis lengths, laying on the top surface of the Si pillar. The latter is represented by a simple parallelepiped shape, carved away from the simulation cell, with Neumann boundary conditions at its sides. The cell size is set to match the experimental pattern.

3. RESULTS AND DISCUSSION

Since SEM images (not shown) evidenced that the selective deposition of Ge on both SUB1 and SUB2 substrates follow a very similar behaviour, we first focused here on the growth dynamics on SUB2. A detailed comparison between the two substrates will be resumed in the following of the manuscript when discussing the formation of the continuous Ge layer. In Fig. 1(c-e), we display the SEM images of samples in which a 60 min deposition was performed with different recipes. **The first growth attempt was carried out using a "standard" two-temperature (300°C/550°C) RPCVD process using GeH_4 as reactant gas with H_2, which lead to high quality Ge/Si(001) relaxed heterolayers[25] and was close to the process temperature conditions of the reported LEPECVD process (~ 500°C)[1,2]**. Figure 1(c) shows that, beside the formation of high quality, faceted Ge µ-structures on top of the Si pillars, the

growth process resulted in the nucleation of several randomly distributed and sized Ge cluster on the SiO_2 pillar sidewalls (arrow). One possible explanation for this additional Ge growth on the SiO_2-covered sidewalls of the pillar, which is not observed on "flat" SiO_2 hardmask layer[21], could be the presence of local imperfections in the SiO_2 layer, like for instance the presence of residual Si and carbon particles after the Bosch lithographic process. **To remove potential local imperfections in the SiO_2 layer and to increase the selectivity of the Ge growth, we were forced to modify the process condition by eliminating the low temperature deposition and increasing the subsequent deposition temperature from 500°C, i.e. one used in LEPECVD, to 650°C, with the addition of HCl gas to the H_2-GeH_4 gas mixture[26].** The role of the HCl is to etch away low-quality material, thus favoring the selective growth of the material on the top pillar opening, as can be seen in Fig. 1(d). Here we point out that deposition carried out on pillars featuring bare Si sidewalls (i.e. not covered by SiO_2), the growth resulted to be non-selective even when HCl was employed, as can be seen in Fig. 1(e). Therefore, *both* sidewall oxidation and a gas phase etchant are needed to achieve VHE in a RPCVD process.

Growth is here below analyzed in detail, including intermediate stages (Fig. 2(a)) and comparisons with simulations (Fig. 2(b)), for the SUB2 case, offering the possibility to directly compare results with typical LEPECVD growth on 2×2 μm² Si pillars[2].

In Fig. 2(a), we display the Ge growth dynamics by showing the SEM images of samples obtained for different deposition times t_{dep}. After the first nucleation of isolated islands ($t_{dep} \leq 1$ min), a pronounced coarsening is observed, leading to a single, non-faceted μ-crystal (t_{dep} = 15 min). Further evolution (t_{dep} = 60 min) leads to the appearance of well-defined {001}-, {113}-, {111}-, and {110}-oriented facets. For increasing deposition time (120 min) the {001} almost disappears, and the Ge μ-structure acquires a more rounded shape, hosting large {113} facets as

expected from Ge equilibrium crystal shape calculations[27]. In Fig. 3, we display a comparison of the top morphology between the two latter cases as obtained by AFM.

The growth process has been modelled using Phase-field simulations based on an evaporation/condensation dynamics, i.e. the motion of the growth front results from the difference between the local chemical potential $\mu(x)$ at the surface and its reference value in the gas. Further details are given in the Experimental Section.

The simulation results for Ge μ-structures, having similar volumes of the experimental ones, are reported in Fig. 2(b). The simulated growth dynamics highlights both the decrease in the total free-surface area and, with the help of a dedicated colour map, the progressive exposure of facets characterized by low surface-energy density values.

The comparison between experiments and simulations is particularly interesting. Up to the appearance of well-defined facets (t_{dep} = 60 min) observed and predicted evolutions are very similar. **As the model directly promotes the exposure of lower-energy facets, this confirms that, at variance with previous LEPECVD experiments at lower temperature and higher deposition rates[1], at least during the initial stages, the growth is mainly driven by thermodynamics.** The model, however, does not seem to quantitatively capture the late evolution, overestimating the size of the (001) top facet, i.e. underestimating the corresponding growth velocity with respect to other facets. It appears that the incorporation-time, orientation-independent in the model, of facets such as {113} is slower than on (001), in agreement with the conclusions drawn in Refs.[1, 2].

The ratio between facet growth velocities can be estimated from the comparison between the morphology of the t_{dep} = 60 min and t_{dep} = 120 min samples, as determined by SEM (Fig. 2(a)). Ideally, the resulting values should be included in our phase-field growth model. This is however

highly non-trivial, and it is left for future work. Nonetheless, a much simpler qualitative analysis of the effect of growth velocities can be carried out. In Fig. 4(a), a tentative prediction of later stages of growth (after t_{dep} = 180, 240, and 300 min) is drawn by simply assuming self-similar growth at constant growth velocities.

This corresponds to assuming growth to be fully determined by facets velocities, with inter-facet diffusion being kinetically frozen, thereby mimicking typical LEPECVD conditions[1]. Merging of adjacent pillars is predicted, leading to the formation of a fully faceted continuous film (Fig 4(b)). This regime contrasts to some extent with the predictions of the evaporation/condensation model. Indeed, if the latter simulations are carried on for longer times (90-150 min) a continuous film is still predicted, but the resulting morphology appears much smoother, as evidenced in Fig. 4(c). Moreover, the lateral enlargement is overestimated leading to merging after only ~120min.

Regardless the inherent discrepancy between the two theoretical predictions, both approaches strongly suggested to perform deposition for longer time, in the hope of observing partial/full merging of the Ge crystals and to clarify both kinetic and thermodynamic contributions. Following the theoretical estimate, a new sample was therefore deposited for 300 min, leading to the continuous, suspended layer morphology displayed in Fig. 5. Even if faceted features still persist in correspondence of the µ-crystal tops, recalling the prediction of the kinetic model (Fig 4(b)), we can observe strong smoothening and accumulation of material below the film (black arrow in Fig. 5(b)), as predicted by the evaporation/condensation model. A further surface smoothening could be achieved by exploiting the enhancement of the surface diffusion[10, 22] through extended growth-interruption/ annealing experiment which will be the subject of a follow-up study.

In summary, the comparison between theory and experiments reveals the important roles played by both thermodynamics (shape determined by surface-energy minimization) and kinetics (orientation-dependent incorporation times) in determining the growth of isolated Ge crystals and their eventual merging into a continuous film by RPCVD.

A more direct comparison with low-temperature/high rate LEPECVD is now possible. As we have shown above, the RPCVD-VHE of Ge on Si pillars requires oxidation of the lateral sidewalls in order to prevent Ge growth in other places except the top region. In LEPECVD such "localization" is instead ensured by the very short diffusion lengths[1], so that oxidation is not required. Even in the presence of oxide on the sidewalls, the Ge crystals grown by RPCVD follow a different morphological evolution. They quickly enlarge laterally, so that merging is achieved after a few microns of vertical growth only. This is in marked contrast with low-temperature LEPECVD (~500°C), where adjacent crystals were shown to develop vertical facets and to grow parallel without ever touching each other for over 50 microns[1]. Longer diffusion lengths in RPCVD, allowing for material redistribution are likely to cause this difference.

Indeed, in the Fig. 5 of Ref.[11], results similar to the present ones were obtained by forcing unusually slow deposition rates and high deposition temperatures (and, thus, longer diffusion length) in LEPECVD, leading to lateral Ge-crystal merging after 2 µm of Ge deposition only.

After the growth procedure was successfully tested and analyzed on SUB2, we repeated the same deposition protocol on SUB1. As shown in Fig. 6(a), beside some undulations at the top free surface, a continuum film similar to that obtained on SUB2 (b) is achieved.

Let us finally supply a thorough characterization of the deposited material quality using TEM, XRD, and µ-Raman measurements. Close-up TEM images of the Ge region located above the Si pillars (Fig. 6(c, d)) reveal highly defective areas limited to the Si/Ge interface regions (blue arrows).

Only few defects, on {111} glide planes, are located at other positions in the Ge layer or even reach the surface (white arrows). It is evident that, beside misfit dislocations, the defective areas at the Si/Ge interfaces are mainly populated by stacking faults and microtwins, nucleating mostly during the first 15 min of Ge growth by coalescence of initial 3D Ge islands (see early growth state TEM images in Fig. 7). Finally, defect etching showed a threading dislocation density of ~5×10^7 cm^{-2}, which compares favorably against what obtained in Ge deposition on unpatterned Si wafers if, as in the present case, cycling annealing is not used (~8×10^8 cm^{-2} from Ref.[25]).

To determine the in-plane strain and the epitaxial relationship of the continuous Ge layer, we have performed XRD reciprocal space mappings (RSM) of the asymmetric Si and Ge($\bar{2}\bar{2}4$) reflections for Ge layer grown for t_{dep} = 300 min on both SUB1 and SUB2. In Fig. 8(a, b), we can see for both samples a sharp, high intensity Si($\bar{2}\bar{2}4$) signal surrounded by a broader, low intensity corona of ellipsoidal shape spreading along the [110] direction. As evidenced by Si(004) XRD ω scans in [110] direction (Fig. 8(c, d)), the ellipsoidal-shaped corona extends for $\Delta\omega = \pm1.3°$ in the SUB1 (Fig. 8(c)) and $\Delta\omega = \pm0.5°$ in the SUB2 cases (Fig. 8(d)), indicating an additional lattice curvature (tilt) in the Si μm-sized pillars. As we have discussed recently[12], the thermal mismatch strain between a grown continuous Ge layer "patch" and their underlying "block" array of Si pillars is partially relaxed by an inward bending of the pillars themselves; excepting stronger bending with thinner pillars (SUB1). In conclusion, the XRD observed Si tilting here orginates also from the Si pillars bending, which is larger in SUB1, and results in the shown thermal strain relaxation.

The in-plane strain of the coalesced Ge layer was measured to be $\varepsilon = +0.02 \pm0.05\%$, for SUB1 (Fig. 8(a)) and $\varepsilon = +0.10 \pm0.05\%$ for SUB2 (Fig. 8(b)), to be compared with the $\varepsilon = +0.18\%$ strain expected for a Ge film grown at 650°C on an unpatterned substrate[28].

The investigation of the lateral distribution of the strain relaxation over the Ge "patches" on Si pillar "block" arrays carried out by μ-Raman spectroscopy confirms the previous analysis and shows the rather good homogeneity of the measured strain over the entire patch (see Fig. 9), a very important feature in view of potential applications. In the Ge layer grown on SUB1, the patch boundaries are slightly more relaxed due to the higher tilt of the underlying pillars there (Fig. 9(a)), while, the lesser pillar tilt allows for a decreased strain relaxation only in the more "rigid" SUB2 sample (Fig. 9(b)). The average strain measured on two independent SUB1(SUB2) samples is ε = +0.05% (+0.14%), in good agreement with the XRD determination, with a full-width at half maximum of the strain distribution over the whole patch equal to FWHM= 0.09% (0.045%), as can be derived from Fig. 9 (c) and (d), respectively.

4. CONCLUSIONS

Isolated Ge crystals and suspended layers on micron-sized Si pillars display superior properties with respect to standard 2D layers, as they allow for fine control on thermal stress and defect distribution . Such structures were so far achieved by LEPECVD only, a rather peculiar growth technique. In this work, we have demonstrated that under suitable conditions, similar results can be obtained by exploiting a widely commercial available epitaxial tool such as RPCVD. In particular, vertical heteroepitaxial growth by RPCVD was forced by pre-oxidizing the lateral sidewalls of the pillars, and full selectivity was achieved by exploiting chlorine as etchant.

It was shown that a thermodynamic growth model nicely captures the first stages of growth and some general features of the grown structures, while kinetic contributions, here described by a faceted model, become more important when considering mature isolated Ge crystals, eventually leading to merging. **Despite some differences with respect to the typical morphology of the**

Ge crystals grown by LEPECVD at lower temperatures and higher deposition rates, our results prove that also RPCVD-VHE is effective in relaxing the thermal strain via the compliance of the underlying patterned Si, if sufficiently thin pillars are used and thus can potentially be employed to deposit arbitrarily thick, crack-free heteroepitaxial layers.

ASSOCIATED CONTENT

Figure captions

Figure 1: (a) SEM image of prepared 2 × 2 μm² Si pillars, 8 μm tall, with 1 μm wide spacing. The Si pillar in the foreground was mechanically cleaved in order to display the 80 nm thick SiO_2 sidewall coating. The wave-like shape of the SiO_2 sidewalls is typical for the Bosch process. (b) SEM image of a typical pillar-patterned substrate. (c-e) SEM images of 2 × 2 μm² Si pillars after 60 min Ge deposition by RPCVD (c) using a standard two-step process at 300°C/550°C without cyclic annealing[25], (d) using a one-step process at 650°C with HCl gas as etchant[26], and (e) using a one-step process at 650°C with HCl gas but without SiO_2-covered sidewalls.

Figure 2: (a) SEM images of the crystal growth on SUB2. Different durations of the deposition process are considered. (b) Selected stages of a phase-field growth simulations reproducing the trend observed in experiments. The initial profile is set by a random distribution of ellipsoids, mimicking the first islands. {001}, {113}, {111}, and {110} facets are considered as highlighted by the colour map, showing the variation of the surface energy density γ along the profile.

Figure 3: AFM analysis of the top of a typical Ge crystal grown on SUB2 for (a) t_{dep} = 60 min and (b) t_{dep} = 120 min. Both 3D perspective views and gradient images are shown.

Figure 4: Simulations of profile evolution toward crystal coalescence in a suspended film. (a) Prediction of the facet evolution in a fully kinetic regime for an isolated crystal with growth rates extrapolated from the experimental profiles at 60 min and 120 min long deposition, overlapped as in the figure (in red and blue respectively). In particular growth rates relative to (001) are v_{113}≈76%, v_{111}≈68%, v_{110}≈48%. (b) Prediction of the conditions of crystal coalescence when the crystals are 1 μm far apart, based on the calculation in panel a. (c) Extension of the phase field simulation shown in Fig. 2(b) to longer times in order to observe crystals coalescence.

Figure 5: (a) SEM view of the suspended Ge film obtained by 300 min growth on SUB2. (b) Evolution sequence reconstructed by overlapping SEM images for the crystals grown after 60 min (red coloured) and 120 min (blue) with a TEM cross-section of the suspended film obtained after 300 min deposition on SUB2 (green).

Figure 6: Cross-section TEM images of coalesced, single crystalline ~1.8 μm thick Ge layers after t_{dep} = 300 min Ge growth on (a) SUB1 and (b) SUB2. Black arrows indicate material growth below the film as in Fig. 5(a). Corresponding close-up images are displayed in panels (c) and (d), respectively. All images are projected along the <1$\bar{1}$0> azimuth.

Figure 7: ***(a)*** Cross-section TEM image of defective single crystalline Ge islands after t_{dep} = 1 min Ge growth on 2 × 2 μm² Si pillars (SUB2) as well as (b) corresponding close-up image of one selected Ge island. Images are projected along the <1$\bar{1}$0> azimuth.

Figure 8: RSM of asymmetric ($\bar{2}\bar{2}$4) reflections of Si and Ge measured on a t_{dep} = 300 min Ge layer grown on (a) SUB1 and (b) SUB2. Q_z-axis is parallel to (400) net plane normal and Q_x-axis is perpendicular to Q_z in the diffraction plane. The red arrows indicate the direction of full

relaxation in reciprocal space. Corresponding ω scans measurements for (c) SUB1 and (d) SUB2 around the Si(004) reflection in [110] direction.

Figure 9: μ-Raman lateral strain mappings and their corresponding lateral strain histograms (average lateral strain value fitted by Gaussian function (black)) of a 200 × 200 μm^2 block array of Si pillars overgrown by coalesced Ge layer after 300 min growth on (a, c) SUB1 and (b, d) SUB2, respectively.

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