Heterogeneous nucleation of catalyst-free InAs nanowires on silicon

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Abstract
We report on the heterogeneous nucleation of catalyst-free InAs nanowires on Si(111) substrates by chemical beam epitaxy. We show that nanowire nucleation is enhanced by sputtering the silicon substrate with energetic particles. We argue that particle bombardment introduces lattice defects on the silicon surface that serve as preferential nucleation sites. The formation of these nucleation sites can be controlled by the sputtering parameters, allowing the control of nanowire density in a wide range. Nanowire nucleation is accompanied by unwanted parasitic islands, but careful choice of annealing and growth temperature allows us to strongly reduce the relative density of these islands and to realize samples with high nanowire yield.

Keywords: nanowire, InAs, catalyst free, silicon

1. Introduction

Monolithic integration of III–V nanowires on silicon substrates has gained considerable research interest because of its significant potential for future electronic and opto-electronic devices. Indeed, a broad range of devices such as light emitting diodes (LEDs) [1], lasers [2], field effect transistors (FETs) [3], and solar cells [4] have already been demonstrated based on III–V NWs grown on silicon substrates. Interesting device applications such as tunnel diodes and tunnel FETs [5] were also reported by utilizing III–V/silicon heterojunctions.

The common way to grow NWs is the VLS mechanism, in which foreign metal nanoparticles deposited on the silicon substrate catalyse the growth [6]. Catalyst-free growth techniques are however intensely explored, driven by the need to avoid metal contamination, and favour full compatibility with the present silicon technology [7, 8]. Several silicon substrate preparation approaches were employed for reproducible and controllable NW growth. Self-assembled III–V NWs were grown on oxide-masked silicon with nanometre-sized pin-holes [9], while site-controlled growth was demonstrated using selective area epitaxy (SAE) [10]. NW density and morphology were shown to be controlled by the oxide mask itself, but this in turn requires precise control over the chemical and physical properties of the oxide [11], thus making this approach rather time consuming and challenging, in contrast to the simple case when NWs are directly grown on silicon substrates [12]. Actually, direct growth of III–V NWs, in particular InAs on silicon, does have critical issues. InAs NW growth is usually accompanied by unwanted parasitic islands, and both continue to nucleate even after long growth times [13, 14]. These islands cover a substantial fraction of the silicon surface, which is further decreased by additional nucleation, leading to coalescence of NWs and islands. Not only does coalescence reduce the density of freestanding NWs, it also results in the deterioration of the NW morphological, structural, and optical properties [15]. For most practical purposes, it is desirable to suppress the coalescence process by reducing the initial nucleation density as well as inhibiting new nucleation. However, there are very few reports on density-control methodologies for InAs NWs on silicon [16, 17]. Therefore, there is an urgent need to
understand the nucleation mechanisms of NWs and islands deposited on silicon to achieve control of their density.

In this work, we present a new approach to prepare Si(111) surfaces for the controlled growth of InAs NWs by chemical beam epitaxy (CBE). We show that controlled sputtering of the Si(111) surface enhances the nucleation of InAs crystals (NWs and islands), while no nucleation occurs on non-sputtered Si(111) surfaces under identical growth conditions. We argue that this stems from the formation of surface defects associated with the sputtering process that serve as preferential physical nucleation sites. Furthermore, we show that the silicon surface can be modified by in situ growth and ex situ sputtering parameters, allowing us to control the density of InAs NWs over a wide range. We show that the yield of NWs with respect to islands can be maximized by choosing an optimum growth window, obtaining InAs NW densities in the range of about 1–30 NWs μm⁻² with a yield of about 50%.

2. Experimental details

The substrate preparation protocol is schematized in figure 1. Commercially available Si(111) substrates covered with a native or 20 nm thick thermal-oxide layer were used (figure 1(a)). The oxide was completely removed by a 2 min buffered oxide etch (BOE) with an etch rate of about 90 nm min⁻¹. The substrates were then rinsed in deionized water and blown dry with nitrogen to obtain hydrogen-terminated Si(111) surfaces. These substrates will be termed ‘non-sputtered’ silicon substrates in the following (figure 1(b)). No NW growth occurred on non-sputtered silicon substrates (figure 1(c)).

Subsequent surface treatment involved sputtering the silicon surface with controlled argon or SiO₂ energetic particles (figure 1(d)). Energetic argon particles were obtained by an inductively coupled plasma (ICP) sputtering system (chamber pressure = 1.5 × 10⁻⁴ Torr, argon flow = 15 sccm), while SiO₂ particles are obtained by a radio-frequency (RF) magnetron sputtering system (chamber pressure = 1 × 10⁻⁴ Torr). After sputtering and prior to NW growth, the sputtered substrates were again etched for 2 min in BOE to remove residuals and deposited oxide. The substrates were rinsed in deionized water and blown dry with nitrogen. These substrates will be termed ‘sputtered’ silicon substrates (figure 1(e)) and were used for NW growth (figure 1(f)).

Immediately after etching, all substrates were indium-bonded on a molybdenum plate in air by keeping them on a hot plate at 250 °C for less than a minute, transferred to the load lock of the CBE chamber and pumped to 10⁻⁷ Torr in a few minutes. Auger electron spectroscopy of ‘as mounted’ substrates reveals that some silicon oxides are present on the surface (see supplementary information S1, figure S1), as a consequence of the indium bonding procedure. Although the amount of oxide is very small, this indicates that substrate handling before mounting requires particular care.

The process employed for growing InAs NWs involved a high temperature annealing at a temperature Tann under As flux corresponding to a TBAs line pressure of 1.0 Torr for 15 min. After this step, the temperature was ramped down to the growth temperature Tgrowth, and InAs growth was initiated by introducing As and In fluxes at metal–organic line pressures P_{TBAs} = 3.0 Torr and P_{TMIn} = 0.3 Torr respectively, for a growth time tgrowth. The NW growth was terminated by cooling the samples under As flux. All temperatures were measured using a pyrometer with an accuracy of ±10 °C.

Field-emission scanning electron microscopy (SEM) images of the NWs were acquired with a Zeiss Merlin SEM operated at 5 keV. For each sample, the density was measured from plan-view SEM micrographs taken from at least ten random areas of the sample. The minimum measurable InAs crystal size is about 3 nm (SEM image resolution). The statistical data on the density were obtained using open-source ImageJ software by automatic particle counting after thresholding the SEM images. High-resolution transmission electron micrographs (HRTEM) were acquired using a Libra 200 TEM operating at an accelerating voltage of 200 kV. The incident electron beam was in the vicinity of the (110) silicon zone axis. The samples were sliced into thin sections by focused ion beam (FIB).

3. Nucleation and growth mechanism

Figure 2 shows plan-view SEM micrographs of InAs NWs grown on non-sputtered Si(111) (figure 2(a)) and on sputtered Si(111) substrates employing two different sputtering procedures: SiO₂ sputtering using an RF magnetron sputtering system (figure 2(b)) and argon sputtering using an ICP system (figure 2(c)). The sputtered surface using SiO₂ sputtering was obtained using a sputter bias (V_{sputter}) of 520 V and a sputter time (t_{sputter}) of 60 s. V_{sputter} and t_{sputter} were 35 V and 600 s, respectively, for ICP argon sputtering. The NW growth process involved annealing at T_{ann} = 790 °C and the growth at T_{growth} = 400 °C for t_{growth} = 15 min.

SEM micrographs in figures 2(a)–(c) show that InAs crystals nucleate only on the sputtered Si(111) surfaces. Extensive investigation of non-sputtered samples grown with
The InAs crystals shown in figures 2(b) and (c) consist of NWs (red circles) and islands (yellow circles). Figure 2(d) shows a magnified view of InAs NWs and islands and highlights their morphological differences. NWs (red circles) have a hexagonal cross-section delimited by six \{112\} facets normal to the Si(111) surface. On the other hand, islands (yellow circles) display irregular cross-sections. For a given sample, the yield \(Y\) of NWs is determined as the ratio of NW density to the total crystal (NWs and islands) density. It is \(\sim 0.5\) for the samples shown in figures 2(b) and (c). As will be shown later, \(Y\) can be controlled by the annealing and growth conditions. Figure 2(e) shows a cross-sectional HRTEM image of a NW taken from the sample shown in figure 2(b). The micrograph is taken along the Si(110) zone axis with the NW growing along the vertical [111] axis. At this early stage, for the chosen growth parameters, the NWs have aspect ratio less than unity and display vertical [112] sidewalls. NWs are growing epitaxially on the substrate, with horizontal InAs (111) planes and a flat top facet. Already at this early stage of growth, the crystal structure has a high stacking fault density, showing intermixing of wurzite and zincblende layers. Overall, the morphology is equivalent to that of longer InAs NWs previously reported [7]. Figure 2(f) is a 90° tilted SEM micrograph of the sample of figure 2(b) with the electron beam impinging normally on a Si(110) cleavage plane, showing again the (111) oriented growth of the NWs. As in the other panels, the red and yellow circles highlight one NW and one island, respectively.

In order to identify defect formation on the sputtered silicon surface, we investigated the characteristic features after sputtering. It is well known that sputtering can transform a crystalline material into an amorphous state when the density of defects becomes very high [22]. However, the NW growth with the (111) axis normal to the Si surface (see figure 2(f)) suggests that the crystalline order of the silicon substrate is not destroyed by sputtering and that the surface is not amorphized. This is confirmed by HRTEM analysis, which shows defect-free crystalline order of the sputtered Si substrates after growth. An example of a micrograph acquired along the Si(110) zone axis is displayed in figure 3(a). The darker region at the bottom of the image is the Si substrate, where the crystal lattice is evident. The darker areas near the NW base are due to strain contrast derived from the large lattice mismatch between Si and InAs. Higher energy sputtering \(V_{\text{sputter}} > 40\) V in ICP argon sputtering and \(V_{\text{sputter}} > 520\) V in RF magnetron SiO2 sputtering (see supplementary information S2 and S3) results in a decrease in the InAs crystal density (see figures S2 and S3 in supplementary information). Hence, we argue that the nucleation sites that mediate epitaxial growth may be surface

Figure 2. Plan-view SEM micrographs of InAs crystals grown under identical conditions on (a) non-sputtered, (b) SiO2 sputtered, and (c) ICP argon sputtered silicon substrates. (d) Plan-view SEM micrograph of InAs NWs and islands from the sample shown in (b). The coloured markers are used to identify islands (yellow) and NWs (red). (e) High-resolution TEM image of an individual InAs NW from the sample shown in (b) taken along the Si(110) zone axis. (f) 90° tilted SEM micrograph of the same sample as in (b).

a wide variety of parameters never revealed NWs or islands except near borders or recognizable defected areas of the substrate. The enhancement of nucleation on sputtered Si(111) surfaces in comparison to that of non-sputtered Si(111) surfaces is therefore likely due to the formation of surface defects by sputtering [18]. Although a large number of processes occur during sputtering, it is well established that bombarding a surface with energetic particles leads to the formation of surface defects, which here serve as preferential nucleation sites [18, 19]. In fact, the role of surface defects as nucleation sites in catalayt-free NW growth is already established: it was found that defects ease the formation of critical nuclei and enhance the nucleation rate [20, 21].

The different InAs crystal density on the two types of sputtered silicon surface shown in figures 2(b) and (c) is likely due to the different sputtering parameters \(V_{\text{sputter}}\) and \(t_{\text{sputter}}\) and techniques employed. The details of the influence of sputtering parameters on the InAs crystal density \(N\), which is the sum of InAs NW and island density, for the two sputtering techniques are discussed in the supplementary information (see figures S2 and S3). It is shown that \(N\) increases with \(V_{\text{sputter}}\) and \(t_{\text{sputter}}\) as long as surface defects favourable for nucleation are created by the sputtering particles. Although \(N\) depends strongly on \(V_{\text{sputter}}\) and \(t_{\text{sputter}}\) (see supplementary information S2 and S3), a direct comparison of the two techniques is not straightforward. This is because the impact of the particles impinging on the surface depends not only on \(V_{\text{sputter}}\) but also on the mass of the sputtering species, chamber pressure and geometry, which are different in ICP argon and RF magnetron SiO2 sputtering systems. The chamber pressure and geometry can influence the mean free path of the sputtering particles and thus their energies. In addition, an oxide layer is deposited during SiO2 sputtering that can inhibit the creation of new defects when the thickness of the deposited layer is greater than the penetration depth of the sputtering particles.

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The roughness has been calculated as the average roughness across resolutions and scan sizes. After a surface flattening to remove acquisition artefacts, the roughness has been calculated as the average roughness (Rₐ) using WSXM software [25]. Depending on the individual scan, Rₐ values show some variability. The mean Rₐ from all the scans (±standard deviation) is 0.38 ± 0.19 nm and 0.59 ± 0.29 nm for the non-sputtered and sputtered (SiO₂ sputtering, Vₛputter = 520 V, tₛputter = 60 s) surfaces, respectively. Though a slight difference is observable, it is well within the uncertainties of the measured roughness. Representative AFM topographs are reported in figures 3(b) and (c) for a non-sputtered and a sputtered sample, respectively.

The final confirmation of the role of surface defects as InAs nucleation centres is shown in figure 4. Figures 4(a) and (b) show N and Y, respectively, as a function of Tₐₐₙ for InAs crystals grown on SiO₂ sputtered silicon substrates (Vₛputter = 520 V, tₛputter = 60 s). Figures 4(c)–(e) are selected SEM micrographs of samples grown after annealing at different Tₐₐₙ. Annealing time was kept fixed at 15 min and the process was performed in an As flux corresponding to Pₐₐₙ of 1.0 Torr. Growth was carried out at Tₕₒₛₜₑₑₑ = 400 °C for 15 min.

Figure 4(a) demonstrates an appreciable reduction of InAs crystal density (N) with an increase in the annealing temperature. At the lowest Tₐₐₙ of 460 °C N is highest (~210 crystals μm⁻²), while at the highest Tₐₐₙ of 830 °C InAs crystals are completely absent. This reduction in N observed due to annealing suggests that more surface defects annihilate as Tₐₐₙ is increased, and at the highest Tₐₐₙ of 830 °C the sputtered silicon surface regenerates into an ordered silicon surface, comparable to a non-sputtered one (see figure 2(a)).

Figure 4(b) shows that the yield of NWs (Y) increases with Tₐₐₙ, and at Tₐₐₙ = 790 °C Y is maximum (~0.5). At Tₐₐₙ > 790 °C there is no InAs crystal nucleation (see figure 4(a)) and hence Y is undefined.

The nucleation mechanism of islands and NWs depends on the nature of the Si(111) surface. Annealing and cooling cycles under UHV conditions determine not only the defect density but also the surface reconstruction [26, 27]. It is known that different surface reconstructions have different surface energies that determine the polarity and morphology of InAs crystals [28–30]. It is likely that the surface reconstructions formed below Tₐₐₙ = 790 °C do not favour NW formation, and therefore the yield is low. On the other hand, we expect that the annealing process at Tₐₐₙ = 790 °C for 15 min and the subsequent cooldown to the growth temperature of 400 °C in As background transforms a large portion of the surface from Si(1 × 1)–H to Si(1 × 1)–As [27, 31]. At annealing temperature up to approximately 500 °C hydrogen atoms desorb from the hydrogen-passivated silicon surface and it becomes As-passivated in As ambient [29]. An As-passivated silicon surface was found to have a very low surface energy due to the existence of As lone-pair states inducing Volmer–Weber (VW) growth of InAs islands [32]. Moreover, the As-passivated silicon surface was shown to resemble an InAs(111)B oriented surface, and nucleation of VW islands on such a surface resulted in a high yield of vertically oriented catalyst-free NWs [27, 29, 33]. Although the As-passivated silicon surface should be very stable [32], the passivation of the surface may still be incomplete due to...
The activation energy is determined. ICP argon sputtering with sputtering parameters

$V_{\text{sp}} = 520 \text{ V}$ (red squares) and $V_{\text{sp}} = 345 \text{ V}$ (red squares) and $V_{\text{sp}} = 35 \text{ V}$ (blue triangles). The inset shows an Arrhenius plot of the same set of data for the SiO$_2$-sputtered samples, from which the activation energy is determined. Figure 5(b) shows the variation of $Y$ as a function of $T_{\text{growth}}$ on the SiO$_2$-sputtered substrates.

Figure 5(a) reveals that $N$ strongly decreases with increasing growth temperature for all types of sputtered substrates. The crystal density $N$ for the substrate sputtered with SiO$_2$ with parameters $t_{\text{sp}} = 3 \text{ s}$ and $V_{\text{sp}} = 345 \text{ V}$ (red squares) is reduced to zero at 420 °C, while at the same temperature $N = 25 \, \mu \text{m}^{-2}$ for the substrate with SiO$_2$ sputtering parameters $t_{\text{sp}} = 600 \text{ s}$ and $V_{\text{sp}} = 520 \text{ V}$ (green circles). On the other hand, the ICP argon-sputtered substrates have a very narrow temperature window with $N = 0.7 \, \mu \text{m}^{-2}$ at 365 °C (blue triangles), while no nucleation takes place above 400 °C for a fixed growth time of 15 min.

This simple observation can be rationalized by the island nucleation theory [19, 37, 38]. According to the theory, the crystal density, $N$, is given as [38]

$$N \propto N_D \exp(-\Delta E_b / kT_{\text{growth}}) \exp(\Delta E_f / kT_{\text{growth}}),$$  

(1)

where $N_D$ is the density of surface defects and $\Delta E_b$ is the nucleation energy barrier associated with defect-mediated heterogeneous nucleation. $\Delta E_f$ is the sum of a number of energy terms for thermally activated processes accounting for surface diffusion, desorption, dissociation of InAs critical nuclei, etc [37, 39]. Generally, a high density of defects effectively reduces $\Delta E_b$ and enhances the nucleation rate.

According to equation (1), the first exponential term will lead to an increase and the second to a decrease of the nucleation density with $T_{\text{growth}}$. Therefore, the decrease in $N$ with increasing $T_{\text{growth}}$, as shown in figure 5(a), can be linked to the increase in thermally activated processes such as surface diffusion and desorption. The total activation energy, $\Delta E$, is the sum of $\Delta E_b$ and $\Delta E_f$. $\Delta E$ is derived from the Arrhenius plot (see inset of figure 5(a)), and is $0.23 \pm 0.05 \text{ eV}$ for SiO$_2$-sputtered silicon with parameters $t_{\text{sp}} = 60 \text{ s}$ at 520 °C and $1.43 \pm 0.30 \text{ eV}$ for sputtered silicon with parameters $t_{\text{sp}} = 3 \text{ s}$ at 345 °C. These results confirm that increased sputtering time and voltage increase the surface defect density and lower the total nucleation energy barrier when grown under identical conditions. As $N$ is zero for ICP argon-sputtered silicon at temperature greater than 400°C, no Arrhenius plot could be obtained.

It can be seen in figure 5(b) that there is an optimum temperature window in which the yield of NWs is maximum. It is seen that, within the temperature window of 385–400 °C, $Y \sim 0.5$, i.e. 50% of the InAs crystals nucleate as NWs. In order to understand this, the island nucleation theory must be coupled with the self-induced nucleation mechanism. According to the self-induced nucleation mechanism, NWs are formed from three-dimensional VW islands after undergoing a series of shape transformations [8, 40]. Such a process is thermally activated and an optimum temperature is required to overcome the energy barrier for shape transformations [41–43]. At this optimum temperature, a large fraction of three-dimensional VW islands can transform into NWs, thereby increasing the yield as shown in figure 5(b).

Figure 5. Growth temperature dependence of InAs crystal density ($N$) for three different substrate preparation protocols using SiO$_2$ sputtering with sputtering parameters $t_{\text{sp}} = 60 \text{ s}$, $V_{\text{sp}} = 520 \text{ V}$ (green circles) and $t_{\text{sp}} = 3 \text{ s}$, $V_{\text{sp}} = 345 \text{ V}$ (red squares) and ICP argon sputtering with sputtering parameters $t_{\text{sp}} = 600 \text{ s}$, $V_{\text{sp}} = 35 \text{ V}$ (blue triangles). The inset shows an Arrhenius plot of the same data for the SiO$_2$-sputtered samples, from which the activation energy is determined. (b) Yield of InAs NWs as a function of $T_{\text{growth}}$ for the SiO$_2$-sputtered samples shown in figure 5(a).

The temperature dependence of adsorption and desorption of As atoms on silicon [31]. The silicon surface may therefore contain at the same time some portions of unpassivated surface [34] and surfaces that resemble the InAs(111)A surface [27, 35]. These surfaces have different surface energies in comparison to the As-passivated silicon surface. The VW islands that nucleate on unpassivated surfaces and surfaces that resemble the InAs(111)A surface may not develop the top (111) facets and therefore continue to grow isotropically, as for the islands as reported in [29, 35, 36]. This simple argument also applies to our case, where 100% yield of NWs was not achieved. Overall, these results indicate that a spectrum of defect sites exists on the sputtered substrate, enabling InAs crystal nucleation whose density and morphology depends on the annealing conditions.

In order to investigate the nucleation process further, we analyse the InAs crystal density ($N$) as a function of growth temperature. Figure 5(a) shows the variation of $N$ as a function of $T_{\text{growth}}$ for SiO$_2$-sputtered silicon substrates employing two different sputtering parameters, and for an ICP argon-sputtered silicon surface. NW growth time was set at 15 min and was preceded by annealing at $T_{\text{ann}} = 790 \text{ °C}$ for 15 min. The inset shows an Arrhenius plot of the same set of data for SiO$_2$-sputtered silicon substrates, from which the activation energy is derived. Figure 5(b) shows the variation of $Y$ as a function of $T_{\text{growth}}$ on the SiO$_2$-sputtered substrates.

According to equation (1), the first exponential term will lead to an increase and the second to a decrease of the nucleation density with $T_{\text{growth}}$. Therefore, the decrease in $N$ with increasing $T_{\text{growth}}$, as shown in figure 5(a), can be linked to the increase in thermally activated processes such as surface diffusion and desorption. The total activation energy, $\Delta E$, is the sum of $\Delta E_b$ and $\Delta E_f$. $\Delta E$ is derived from the Arrhenius plot (see inset of figure 5(a)), and is $0.23 \pm 0.05 \text{ eV}$ for SiO$_2$-sputtered silicon with parameters $t_{\text{sp}} = 60 \text{ s}$ at 520 °C and $1.43 \pm 0.30 \text{ eV}$ for sputtered silicon with parameters $t_{\text{sp}} = 3 \text{ s}$ at 345 °C. These results confirm that increased sputtering time and voltage increase the surface defect density and lower the total nucleation energy barrier when grown under identical conditions. As $N$ is zero for ICP argon-sputtered silicon at temperature greater than 400°C, no Arrhenius plot could be obtained.

It can be seen in figure 5(b) that there is an optimum temperature window in which the yield of NWs is maximum. It is seen that, within the temperature window of 385–400 °C, $Y \sim 0.5$, i.e. 50% of the InAs crystals nucleate as NWs. In order to understand this, the island nucleation theory must be coupled with the self-induced nucleation mechanism. According to the self-induced nucleation mechanism, NWs are formed from three-dimensional VW islands after undergoing a series of shape transformations [8, 40]. Such a process is thermally activated and an optimum temperature is required to overcome the energy barrier for shape transformations [41–43]. At this optimum temperature, a large fraction of three-dimensional VW islands can transform into NWs, thereby increasing the yield as shown in figure 5(b).
Furthermore, the yield of NWs is independent of the sputtering parameters ($t_{\text{spatter}}$ and $V_{\text{spatter}}$) and is about 0.5 for all samples annealed at $T_{\text{ann}} = 790 \degree C$ for 15 min and grown for 15 min at 400°C (see figure S2(c) in supplementary information).

The nucleation mechanism was also investigated as a function of growth time, $t_{\text{growth}}$. Figures 6(a) and (b) report $N$ as a function of $t_{\text{growth}}$ for samples grown on ICP argon-sputtered substrates with sputtering parameters $t_{\text{spatter}} = 600 \text{ s}$, $V_{\text{spatter}} = 35 \text{ V}$, and on SiO$_2$-sputtered substrates with sputtering parameters $t_{\text{spatter}} = 60 \text{ s}$, $V_{\text{spatter}} = 520 \text{ V}$, respectively. Figure 6(c) shows an SEM micrograph of a sample grown for 135 min on an ICP argon-sputtered substrate. Figures 6(d) and (e) show SEM micrographs of samples grown for 15 and 60 min, respectively, on SiO$_2$-sputtered substrates. Growth is conducted at $T_{\text{growth}} = 400 \degree C$.

For the ICP argon-sputtered substrates, as shown in figure 6(a), $N$ increases with increasing $t_{\text{growth}}$ due to continuous new nucleation. Such a process causes a large NW diameter distribution, as shown for a sample grown for 135 min in figure 6(c).

A large distribution of NW diameters is also evident in SiO$_2$-sputtered substrates, as seen in figure 6(d), owing to new nucleation of InAs crystals. However, no significant increase in $N$ with $t_{\text{growth}}$ is seen in SiO$_2$-sputtered substrates (figure 6(b)). This is because the variation of $N$ with $t_{\text{growth}}$ is strongly affected by (i) surface coverage and (ii) coalescence effects. The high density of InAs crystals nucleated at the early stages of growth covers a large fraction of the silicon surface. As the growth time is increased, the size of the crystals increases, which further increase the surface coverage. High surface coverage reduces new nucleation events so that $N$ does not change appreciably [19, 44]. Furthermore, the coalescence of InAs crystals reduces the overall crystal density, and hence the occurrence of new nucleation cannot be reliably inferred from figure 6(b). The coalescence can be seen in a plan-view SEM micrograph in figure 6(e) for a sample grown for 60 min.

4. Conclusions

In conclusion, we have shown that InAs crystals nucleate on sputtered Si(111) surfaces while no nucleation occurs on non-sputtered Si(111) surfaces. We discuss the role of surface defects as preferential nucleation sites formed by sputtering under controlled parameters using different sputtering techniques. The InAs crystals nucleating on the sputtered silicon surfaces consist of InAs NWs and parasitic islands. Although the nucleation of parasitic islands could not be completely inhibited, the yield of NWs could be increased to about 0.5 by proper choice of growth and annealing temperatures. We show that a range of InAs NW densities can be obtained by modifying in situ growth and ex situ sputtering parameters.

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