Formation of In$_x$Ga$_{1-x}$As nanocrystals in thin Si layers by ion implantation and flash lamp annealing

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Abstract

The integration of high-mobility III–V compound semiconductors emerges as a promising route for Si device technologies to overcome the limits of further down-scaling. In this paper, a non-conventional approach of the combination of ion beam implantation with short-time flash lamp annealing is employed to fabricate In$_x$Ga$_{1-x}$As nanocrystals and to study their crystallization process in thin Si layers. The implantation fluence ratio of Ga and In ions has been varied to tailor the final nanocrystal composition. Raman spectroscopy and x-ray diffraction analyses verify the formation of ternary III–V nanocrystals within the Si layer. Transmission electron microscopy reveals single-crystalline precipitates with a low number of defects. A liquid epitaxy mechanism is used to describe the formation process of III–V nanocrystals after melting of the implanted thin Si layer by flash lamp annealing. The fabricated In$_x$Ga$_{1-x}$As nanocrystals are mainly Ga-rich with respect to the implanted Ga/In ratio.

1. Introduction

The steady progress of microelectronic technology has come to a point where a further decrease of device dimensions approaches physical limits. The integration of III–V compound semiconductors into Si-based systems is a promising path to circumvent these limits. The high charge carrier mobilities in III–V materials in comparison to Si, especially for n-type material, is the outstanding feature which predestines them for the integration in future device technologies [1–3]. Another prominent characteristic of most III–V’s is their direct band gap in the visible or near-infrared range allowing them to be used for optical applications [4–6]. In ternary III–V compounds, the band gap can even be engineered by varying the ratio within the group-III or within the group-V elements [7]. Further applications in non-volatile memories [8, 9] and photovoltaic devices [10] as well as for dilute ferromagnetic semiconductors when combined with e.g. Mn [11] highlight the broad range of technologies which can be accessed with III–V compound semiconductors.

Due to this large potential, there is a great interest for the integration of III–V compound semiconductors into Si technology. This integration can be performed by several techniques, where molecular beam epitaxy [12–14], metal–organic vapor phase epitaxy [15–17] and wafer bonding [18, 19] are the most prominent ones. Another approach is ion beam synthesis of III–V crystallites within the Si host material [20, 21] using sequential ion beam implantation and thermal annealing [22, 23]. With the transition to ultra-short annealing times by flash lamp annealing (FLA), the control of nanocrystal (NC) size and quality could be enhanced [24]. A combination of this III–V integration method with additional reactive ion etching resulted in the formation of p–n-heterojunction diodes based on InAs NCs on top of Si nanocolumns [25]. The integration of III–V compound semiconductor NCs in Ge by this preparation technique [26] allows the combination of the high electron mobility of III–V compounds with the high hole mobility of Ge. Recently, ion beam synthesis of In$_x$Ga$_{1-x}$As NCs in bulk Si in combination with rapid thermal annealing has been demonstrated as well [27].
In this paper, the fabrication of ternary In$_{x}$Ga$_{1-x}$As NCs in thin Si layers by a combination of ion implantation and subsequent ms-range FLA is demonstrated. Using different fluence ratios of Ga and In ions, In$_{x}$Ga$_{1-x}$As NCs of various compositions were prepared. With the aim to characterize the NCs and to understand the formation mechanism, optical and microstructural properties were investigated.

2. Methods

For sample preparation, thermally oxidized SOI (Si-on-Insulator) substrates with a final layer stack of 65 nm SiO$_2$/60 nm Si/150 nm SiO$_2$/bulk Si were used. Initially, the Si device layer had a (100) orientation. These substrates were implanted sequentially with As, Ga and In ions. The used implantation energies have led to a projected range between 100 and 600 cm$^{-2}$. The resulting annealing temperature can be calculated by considering the optical constants, heat capacity and heat conductivity of treated samples. For this purpose the temperature profiles have been simulated by solving the one-dimensional heat equation with the commercial software COMSOL Multiphysics® [28] (see figures 4(c) and A1 in the appendix). Using the simulated temperature profiles the theoretical annealing temperatures range between 1000 °C and 1400 °C.

In order to investigate the In$_x$Ga$_{1-x}$As samples μ-Raman spectroscopy, x-ray diffraction (XRD) and scanning (SEM) as well as transmission electron microscopy (TEM) analyses were performed. During Raman spectroscopy samples were exposed to a 532 nm Nd:YAG laser in backscattering geometry and the Raman shift was measured between 100 and 600 cm$^{-1}$. XRD was performed on an Empyrean Panalytical 4-circle θ-2θ diffractometer using Cu-Kα radiation. Cross-sectional TEM analysis was done with an image C$_t$-corrected FEI Titan 80–300 microscope at an accelerating voltage of 300 kV. Besides bright-field TEM and high-resolution TEM imaging, high-angle annular dark-field scanning TEM (HAADF-STEM) and energy-dispersive x-ray spectroscopy (EDXS) were performed to collect information about chemical composition, defects and interfaces of the III–V NCs.

3. Results and discussion

In figure 1(a) Raman spectra of SOI samples implanted with different In/Ga fluence ratios are compared. Additionally, the Raman spectrum of the Si-implanted reference sample (black) is given. All samples were annealed at 97.2 J cm$^{-2}$ (ca. 1370 °C). Comparing In$_x$Ga$_{1-x}$As samples with the reference, it can be clearly seen that the formation of a crystalline III–V component was successfully achieved by the combination of ion implantation and FLA. Figure 1(b) shows the shift of the peak positions of the InAs-like and GaAs-like TO and LO phonon modes as a function of the In content. The peak positions have been extracted by fitting the Raman spectra and follow the trend of the theoretical curves for a phonon mode shift with varying In$_x$Ga$_{1-x}$As composition [29]. For the binary
III–V compounds, namely InAs (x$^\text{nom}$ = 1) and GaAs (x$^\text{nom}$ = 0), the characteristic phonon modes are observed at 219 (TO) and 237 cm$^{-1}$ (LO) for InAs [30] and at 268 (TO) and 286 cm$^{-1}$ (LO) for GaAs [31], respectively. The spectra of the ternary phases show a two-mode phonon behavior [32]. With decreasing In content (decreasing x$^\text{nom}$) the observed phonon modes change from an InAs-like mode to a GaAs-like one. The InAs-like phonon modes lose intensity and converge while the GaAs-like phonon modes evolve. Due to this two-mode phonon behavior it is possible to draw conclusions about the composition of In$_x$Ga$_{1-x}$As. These calculated compositions (see table 1) differ from the nominal compositions in such a way that the fabricated In$_x$Ga$_{1-x}$As NCs are Ga-rich (x$^\text{Raman}$ < x$^\text{nom}$). It is assumed in a first approximation that the system is sufficiently relaxed and that the observed shifts of the peak positions are mainly due to composition. Nevertheless, the width of the peaks suggests a size or composition distribution of the ternary NCs or, more probable, a combination of both.

The Si 2TA phonon mode present at 301 cm$^{-1}$ for all spectra indicates the recrystallization of the thin Si layer. This is supported by the very strong TO + LO phonon mode peak at 520 cm$^{-1}$. The broad peak shifting from 360 to 390 cm$^{-1}$ with increasing Ga content represents the amphoteric doping of Si in III–V. The substitutional defects Si$_{\text{Ga}}$ (384 cm$^{-1}$) and Si$_{\text{As}}$ (399 cm$^{-1}$) and their donor–acceptor pair (393 cm$^{-1}$) in GaAs are Raman-active [33]. For InAs the Si$_{\text{In}}$ defect shows a Raman shift of 359 cm$^{-1}$ [34]. We suspect a mixture of these defect-related Raman modes to be responsible for this broad peak. At high In contents Si$_{\text{In}}$ defects are more dominant giving rise to a signal at lower wavenumbers, while at high Ga contents the peak shifts towards higher wavenumbers due to Si$_{\text{Ga}}$ and Si$_{\text{As}}$ defects effecting the Raman spectra more. In conclusion, Raman spectroscopy proves that In$_x$Ga$_{1-x}$As compound semiconductor NCs with variable composition can be successfully formed in SOI substrates via sequential ion implantation and subsequent FLA. Furthermore, these In$_x$Ga$_{1-x}$As NCs are heavily doped with Si.

Figure 2(a) shows XRD 2θ scans in the 2θ range from 22° to 100° measured at a grazing incidence angle of 1°. For the Si self-implanted reference sample the XRD pattern provides only the Si reflections (blue bars), which proves polycrystalline recrystallization of the implanted thin Si layer during FLA. The samples implanted with In$^+$, Ga$^+$ and As$^+$ show additional Bragg peaks which are related to III–V compound formation (yellow bars). With decreasing In content, the 111, 220 and 311 reflections of the III–V compound NCs shift to higher angles indicating a decrease of the lattice constant and thus a shift from InAs to GaAs. Having a closer look at the reflections, a peak splitting is observed for the ternary phases, especially for x$^\text{nom}$ = 0.5. This splitting is due to the coexistence of at least two In$_x$Ga$_{1-x}$As phases, which are denoted as In-rich and Ga-rich In$_x$Ga$_{1-x}$As with respect to the nominal composition.

Figures 2(b) and (c) show crystallite sizes and μ-strain of the III–V and Si crystallites deduced from a Williamson–Hall analysis. For both III–V phases, the crystallite sizes are around 30 nm. In contrast, the Si crystallites within the thin reference sample are about 10 nm in size. Despite the grazing incidence, the crystallite size measured by XRD is a statistical average value and represents almost exclusively the vertical dimensions of the III–V NCs. As shown in figure 2(c), the III/V crystallites have lower μ-strain than the Si crystallites in the reference sample. The μ-strain itself displays a variation of the lattice parameter within the crystallites. In extreme cases, this variation can either be accounted to lattice imperfections or to a variation in the composition of the crystallites. However, in the present case the origin of the μ-strain is probably a mixture of both. The μ-strain of both ternary III–V phases shows a slight trend to higher values with decreasing In content (x$^\text{nom}$), but absolute values around 0.1% are small compared to the lattice mismatch expected for In$_x$Ga$_{1-x}$As to Si which is between 11.5% (InAs) and

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**Figure 1.** (a) Raman spectra of SOI samples implanted with different In$^+$/Ga$^+$ fluence ratios and a Si$^+$ implanted reference sample (black). All samples were annealed at 97.2 °C (1370 °C). With decreasing In-content (decreasing x$^\text{nom}$ from 1 to 0) the TO and LO phonon modes shift from InAs-like to GaAs-like positions. The positions of InAs TO and GaAs LO phonon modes are inserted as a guide to the eye. (b) Observed phonon mode positions compared with theoretical peak positions [29] for different ternary In$_x$Ga$_{1-x}$As composition.
4.1% (GaAs). Hence, the $\mu$-strain within the III–V crystallites is negligible. The variation of the $\mu$-strain denoted by the error bars is smaller in the binary samples ($x_{\text{nom}}=1$ and 0) than in the ternary samples. Therefore, we account the $\mu$-strain mainly to a variation in composition of the ternary III–VN C s.

In $x_{\text{Ga}}1-x_{\text{As}}$ has a cubic crystal structure and as the $\mu$-strain is negligible, the lattice parameter of the different ternary III–V phases can be deduced from the XRD peak positions. Furthermore, the composition of the In$_x$Ga$_{1-x}$As NCs can be calculated according to Vegard’s law [35] which says that the lattice parameter of a solid solution changes linearly with the variation of its composition. The determined values ($x_{\text{XRD}}$) are given in table 1.

As a peak splitting is observed in the XRD patterns for the ternary compounds, two $x_{\text{XRD}}$ values are displayed for a specific implantation fluence ratio. Since there are no hints for texture, comparing the intensities of the individual III–V XRD peaks points to a major and minor phase contribution. The NC compositions of the major contribution obtained by XRD verify the statement from Raman spectroscopy that the fabricated In$_x$Ga$_{1-x}$As NCs are mainly Ga-rich with respect to the nominal composition. This fact is supported by the appearance of the In 101 and 110 Bragg reflections at 32.9° and 39.1° (red bars) which proves the presence of a metallic In phase in that samples. With decreasing In content, the In reflections disappear.

In order to get more information about the morphology of the NCs, SEM and TEM analyses were performed. Top-view SEM images in figures 3(a)–(e) depict arbitrarily shaped, bright areas representing particles with sizes ranging from few tens of nanometers to about 400 nm for all fluence ratios. Comparing samples implanted with different fluence ratios, it is observed that the mean particle size has the tendency to decrease with decreasing In content ($x_{\text{nom}} \rightarrow 0$). However, the number of particles does not change significantly which leads to a decrease in the overall volume fraction of the precipitates. In the corresponding HAADF-STEM micrographs of figures 3(f)–(k), cross-sections of these samples are depicted. From top to bottom, the SiO$_2$ capping layer, the thin
Si device layer, the SiO$_2$ buried oxide layer and the bulk Si can be distinguished. In each sample, the thin Si device layer comprises several bright areas which represent the III–V precipitates. These particles mainly appear block-like and are limited in height by the surrounding SiO$_2$ layers. Additionally, smaller precipitates which are located directly at the interfaces are observed and some of them are connected by filaments. Regarding the reduction of particle size with decreasing In content, as observed in the SEM images, the number of III–V precipitates visible in the HAADF–STEM micrographs is not high enough to give an adequate statistical evaluation. Besides the III–V NCs located within the thin Si layer, a dotted line of bright spots is observed in the surrounding SiO$_2$ layers close to their interfaces to the thin Si layer, indicating the presence of material with a high atomic mass.

A statistical particle size analysis of the SEM micrographs has been performed to verify the tendency of the mean particle size to decrease with decreasing In content using the open source software ImageJ [36]. Figure 4 shows the results of this statistical evaluation depicting the particle size histograms of the different SOI samples (a) and the change of the mean particle size with decreasing In content ($x_{\text{nom}}$) (b). Most of the particles formed after FLA have sizes below 100 nm independent of the nominal composition. However, for the samples with higher In content, there are more particles with larger sizes than for the samples with lower In content. This can also be seen for the mean particle size of the ternary III–V NCs, which decreases from 99 nm for the InAs sample towards 81 nm for the GaAs sample. The error bars represent the standard deviation of the mean particle size and decreases with decreasing In content resulting in a narrower particle size distribution for samples with lower In content. The rather large variation in lateral particle size is responsible for the peak broadening of the III–V phonon modes observed during Raman spectroscopy. Figure 4(c) displays the time profile of the surface temperature during FLA annealing with energy density of 97.2 J cm$^{-2}$. After a fast cooling phase caused by heat conduction, further cooling due to thermal radiation and heat convection takes place on a timescale of several s. Whereas GaAs already solidifies after less than 100 $\mu$s, InAs and In stay much longer in the liquid phase.

In figure 5(a) a more detailed TEM analysis of these particles is exemplarily given for the sample with $x_{\text{nom}} = 0.5$ treated with FLA at 97.2 J cm$^{-2}$ (ca. 1370 °C). The cross-sectional bright-field TEM micrograph in figure 5(a) depicts the implanted layer stack with the thin SiO$_2$ capping layer, the recrystallized Si device layer with ternary

![Figure 3. Top-view SEM (a)–(e) and corresponding cross-sectional HAADF-STEM (f)–(k) micrographs of SOI samples with different $x_{\text{nom}}$ annealed with FLA at 97.2 J cm$^{-2}$ (1370 °C). The average lateral size of the precipitates decreases with decreasing In content ($x_{\text{nom}}$). In cross-sectional view, block-like precipitates are observed with a height limited by the surrounding SiO$_2$ layers.](image-url)
III–V NCs, the SiO$_2$ box layer and the Si substrate at the bottom. Due to combined mass–thickness and diffraction contrast in the bright-field image, a clear distinction between Si and III–V crystallites in the device layer is not possible. To obtain micrographs showing almost exclusively atomic number contrast, HAADF-STEM imaging was performed. In figure 5(b) taken from the area marked by the yellow box in figure 5(a), one can see that there are two types of NCs: (1) block-like precipitates penetrating the whole Si layer and (2) triangular-shaped crystallites located at the interfaces. When having a closer look at type (2) NCs, it appears that thin filaments, which also pierce through the Si layer, are attached to them. According to the fast Fourier transform (FFT) given in figure 5(d), the block-like precipitate in figure 5(c) is single-crystalline. Based on the cubic zincblende structure and the measured interplanar distances, the FFT in figure 5(d) corresponds to a [112] zone axis pattern of In$_x$Ga$_{1-x}$As. EDXS analyses (not shown here) performed at various III–V precipitates show variations in relative Ga/In ratio, supporting the existence of different In$_x$Ga$_{1-x}$As phases within one sample.

Similar to the HAADF-STEM micrographs in figure 3, additional small clusters in the surrounding SiO$_2$ layers close to the SiO$_2$/Si interfaces can be observed in the HAADF-STEM image of figure 5(b). These small clusters have sizes ranging from 2 to 10 nm and appear to be crystalline. Since the tails of the implantation profiles extend into the SiO$_2$ layers, it is likely that these III–V NCs have been formed by a solid phase crystallization process as the SiO$_2$ layers remain solid during FLA. They grow due to a process similar to Ostwald ripening, but their size is limited due to the low amount of available group-III- and -V atoms and the lower diffusion constant in the solid phase of SiO$_2$.

Due to the present results a liquid phase epitaxy (LPE) mechanism is proposed as responsible formation mechanism for III–V NCs in the Si device layer. The schematic growth mode is depicted in figure 6. After ion implantation, the Si device layer is amorphous and the implanted group-III and group-V ions have a Gaussian-like depth distribution (figure 6(a)). The a-Si layer has a melting point which is about 200 K lower than that of c-Si$^{37}$ and the impurities decrease it even further, which is why the applied 20 ms FLA pulses are sufficient to melt the entire a-Si layer already at lower energy densities. The fast diffusion of group-III and -V atoms in molten Si results in a homogeneous distribution over the whole Si layer (figure 6(b)). The diffusion coefficients $D$ of these elements in liquid Si are several orders of magnitude higher than in c-Si close to the melting point $^{38,39}$. Immediately after the FLA pulse, crystalline Si seeds start to evolve when the temperature falls below the actual melting point of c-Si and an undercooled melt is formed (figure 6(c)). These c-Si seeds grow with further cooling and, as the group-III and -V elements have segregation coefficients $k$ below one $^{39}$, the molten phase is enriched with these elements (figure 6(d)).

During further cooling the different c-Si grains come into contact with each other and merge to bigger polycrystalline regions with grain boundaries decorated with group-III and -V ions (figure 6(e)). The Si grain

![Image](image-url)

**Figure 4.** Change of the mean particle size of the ternary III–V NCs depending on the In content (a). With decreasing In content, the mean particle size slightly decreases (b). Temperature profile of the SOI wafer surface including the cooling phase after the FLA pulse using a FLA energy density of 97.2 J cm$^{-2}$ (c). Dashed lines represent the melting points of c-Si, GaAs, a-Si, InAs and metallic In (from top to bottom).
growth is also limited by the SiO$_2$/Si interfaces which in turn may be also decorated with the implanted species. Finally, when the temperature decreases below the melting point of the III–V compound, III–V NCs form between the polycrystalline Si grains (figure 6). This III–V formation initially utilizes the recrystallized Si as a template for epitaxial growth but then continues on its own leading to two types of interface regions: those where the crystal orientation of the Si grain matches that of the III–V NC and those where it does not. The Si grains limit the lateral size of the III–V NCs while their height is governed by the thickness of the Si layer, leading to a uniform height distribution over the whole sample. By recrystallization from the melt, the strain in the III–V crystal originating from the lattice mismatch between III and V compounds and Si is reduced. This leads to a lower number of defects, e.g. dislocations, and higher crystalline quality than in case of III–V integration via the solid phase.

The formation of mainly Ga-rich In$_x$Ga$_{1-x}$As NCs is attributed to the melting point difference between InAs and GaAs. For the ternary compound, the melting point decreases with increasing In content [40]. Therefore, a Ga-rich In$_x$Ga$_{1-x}$As phase has a higher melting point and starts to recrystallize earlier during cooling of the molten layer (figure 4(c)). The excess In not consumed in III–V NC formation will form metallic In precipitates when the temperature is low enough. However, there are other effects influencing the final stoichiometry of the NCs like the available space where group-III and -V elements can be drawn from and the probability of more than one NC forming simultaneously. In the end, a distribution of In$_x$Ga$_{1-x}$As compositions is possible although the majority will be Ga-rich with respect to the nominal value ($x_{\text{nom}}$). The decrease in the overall III–V volume fraction with decreasing $x_{\text{nom}}$ can be attributed to the different segregation coefficients of In and Ga. The segregation coefficient of Ga is closer to 1 than that of In which lowers the amount of group-III atoms available during III–V NC formation in samples with high Ga content resulting in a decreased amount of III–V material relative to the Si matrix.

Figure 5. Cross-sectional bright-field TEM (a) and HAADF-STEM micrographs (b) of a SOI sample with $x_{\text{nom}} = 0.5$ annealed at 97.2 J cm$^{-2}$ (1370 °C). A HRTEM image (c) of the big block-like precipitate marked in (b) with corresponding fast Fourier transform (FFT) (d) obtained from the quadratic area marked in (c) reveals an In$_x$Ga$_{1-x}$As single crystal. Based on the cubic zincblende structure and the interplanar distances, the FFT in (d) corresponds to a [112] zone axis pattern.

New J. Phys. 19 (2017) 063019 R Wutzler et al
4. Conclusion

We have demonstrated a process for the integration of III–V NCs in thin Si films. Ternary In$_x$Ga$_{1-x}$As NCs have been fabricated by a combination of high-dose ion beam implantation and millisecond-range FLA. Investigations with Raman spectroscopy and XRD prove the formation of these NCs by the characteristic bulk phonon modes and typical diffraction peaks, respectively. TEM analyses present single-crystalline III–V compound semiconductor NCs in a thin polycrystalline Si layer after implantation and FLA. The formation mechanism is described by a LPE mechanism, which is influenced by diffusion and segregation of the group-III and -V ions in molten Si as well as by the melting points of the particular species. Due to recrystallization via the liquid phase, the strain due to the lattice mismatch between III and V compounds and Si can be lowered and the number of resulting dislocations can be reduced.

In the ternary samples, more than one In$_x$Ga$_{1-x}$As stoichiometry can be present. Depending on the nominally implanted In/Ga/As ratio, the final ternary composition can be adjusted within a certain range. In our experiments, $x$ values ranging from 0 to 1 have been achieved although the In$_x$Ga$_{1-x}$As precipitates are Ga-rich compared to $x_{nom}$, which is accounted to the melting point difference between InAs and GaAs. The presence of mainly Ga-rich ternary NCs and the low segregation coefficient of In leads to the formation of metallic In cluster. The III–V particles are constrained in height by the Si/SiO$_2$ interfaces, leading to control of the size distribution in one dimension, although they have a rather broad lateral particle size distribution. Further control of the lateral size distribution as well as the III–V particle shape can be achieved by introducing a patterned capping layer as implantation mask.

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Appendix

![Temperature profile graph](image)

**Figure A1.** Temperature profiles of the SOI wafer (thickness = 725 μm) at various depths during FLA exposure.

References

[1] del Alamo J A 2011 Nanometre-scale electronics with III–V compound semiconductors *Nature* **479** 317–23
[2] Radosavljević M et al 2009 Advanced high-K gate dielectric for high-performance short-channel In_{0.7}Ga_{0.3}As quantum well field effect transistors on silicon substrate for low power logic applications 2009 IEEE Int. Electron Devices Meeting (IEDM) pp 1–4
[3] Dewey G, Kothyar R, Pillariセットie M, Rakshit T, Then H and Chau R 2009 Logic performance evaluation and transport physics of Schottky-gate III–V compound semiconductor quantum well field effect transistors for power supply voltages (VDD) ranging from 0.5 V to 1.0 V 2009 IEEE Int. Electron Devices Meeting (IEDM) pp 1–4
[4] Ponc F A and Bour D P 1997 Nitride-based semiconductors for blue and green light-emitting devices *Nature* **386** 351–9
[5] Van Campenhout J, Rojo Romeo P, Regreny P, Seassal C, Van Thourhout D, Versluijs S, Di Cioccio L, Fedeli I M, Lagae C and Baets R 2007 Electrically pumped InP-based microdisk lasers integrated with a nanophotonic silicon-on-insulator waveguide circuit *Opt. Express* **15** 6744
[6] Cao Y L et al 2015 Hybrid III–V/silicon laser with laterally coupled Bragg grating *Opt. Express* **23** 8809–17
[7] Yin Z and Tang X 2007 A review of energy bandgap engineering in III–V semiconductor alloys for mid-infrared laser applications *Solid State Electron.* **51** 6–15
[8] Marent A, Geller M, Schliwa A, Feise D, Pötschke K, Bimberg D, Akçay N and Öncan N 2007 106 years extrapolated hole storage time in GaSb/AlAs quantum dots *Appl. Phys. Lett.* **91** 242109
[9] Maier P, Hartmann F, Emmerling M, Schneider C, Hofling S, Kamp M and Worschech L 2014 Charging dynamics of a floating gate transistor with site-controlled quantum dots *Appl. Phys. Lett.* **105** 033502
[10] Dahal R, Pantha B, Li J, Lin J Y and Jiang H X 2009 InGaN/GaN multiple quantum well solar cells with long operating wavelengths *Appl. Phys. Lett.* **94** 063505
[11] Zhou S 2013 Dilute ferromagnetic semiconductors prepared by the combination of ion implantation with pulse laser melting *J. Phys. D: Appl. Phys.* **48** 263001
[12] Yang T, Hertenberger S, Morkötter S, Abstreiter G and Kohlmüller G 2012 Size, composition, and doping effects on In(Ga)As nanowire/Si tunnel diodes probed by conductive atomic force microscopy *Appl. Phys. Lett.* **101** 233102
[13] Benyoucef M and Reithmaier J P 2013 Direct growth of III–V quantum dots on silicon substrates: structural and optical properties *Semicond. Sci. Technol.* **28** 094004–12
[14] Dimakis E, Jahn U, Ramsteinke M, Tahraoui A, Grandal J, Kong X, Marsuard O, Trampert A, Riechert H and Geelhaar L 2014 Coaxial multishell (In, Ga)As/GaAs nanowires for near-infrared emission on Si substrates *Nano Lett.* **14** 2604–9
[15] Honda Y, Kuroiwa Y, Yamaguchi M and Sawaki N 2002 Growth of GaN free from cracks on a substrate *J. Phys. D: Appl. Phys.* **35** 222–4
[16] Tomioka K, Tanaka T, Hara S, Hiruma K and Fukui T 2011 III–V nanowires on Si substrate: selective-area growth and device applications *IEEE J. Sel. Top. Quantum* **17** 1112–29
[17] Watanabe S, Watanabe K, Higo A, Sugiyama M and Nakano Y 2012 Electrical conduction property at InAs/Si(111) interface by selective-area MOVPE Int. Conf. on Indium Phosphide and Related Materials (IPRM) 2012 pp 133–6
[18] Moutanabbir O and Gösele U 2010 Heterogeneous integration of compound semiconductors *Annu. Rev. Mater. Res.* **40** 469–500
[19] Roelkens G, Liu L, Liang D, Jones R, Fang A, Koch B and Bowers J 2010 III–V/silicon photonics for on-chip and intra-chip optical interconnects *Laser Photonics Rev.* **4** 751–79
[20] Budai J D, White C W, Withrow S P, Zuev R A and Zhu J G 1996 Synthesis, optical properties, and microstructure of semiconductor nanocrystals formed by ion implantation *MRS Online Proc. Libr.* **452** 89–98
[21] White C W et al 1996 GaAs nanocrystals formed by sequential ion implantation J. Appl. Phys. 79 1876–80
[22] Komarov F, Vlasukova L, Milchanin O, Komarov A, Wesch W and Yotogamaya A K 2009 Effect of implantation and annealing
regimes on ion-beam synthesis of InAs nanocrystals Lithuanian J. Phys. 49 105–10
[23] Komarov F, Vlasukova L, Wesch W, Kamarou A, Milchanin O, Grechnyi S, Mudry A and Ivanikovich A 2008 Formation of InAs
nanocrystals in Si by high-fluence ion implantation Nucl. Instrum. Methods B 266 557–64
[24] Prucnal S, Turek M, Drozdziel A, Pyszniak K, Zhou S Q, Kanjilal A, Skorupa W and Zuk J 2010 Formation of InAs quantum dots in
silicon by sequential ion implantation and flash lamp annealing J. Appl. Phys. 101 3156–9
[25] Prucnal S et al 2011 n-InAs nanoprisms fully integrated into Silicon Nano Lett. 11 2814–8
[26] Wutzler R, Rebohle L, Prucnal S, Hübner R, Facsko S, Böttger R, Helm M and Skorupa W 2016 III–V nanocrystal formation in ion-
implanted Ge and Si via liquid phase epitaxy during short-time flash lamp annealing Mater. Sci. Semicond. Process. 42 166–9
[27] Khelifi R, Frégaux M, Le Gall Y, Muller D, Schmerber G and Mathiot D 2015 Ion beam synthesis of embedded III–As nanocrystals in
silicon substrate Phys. Status Solidi c 12 35–9
[28] COMSOL Multiphysics(R) 2016 (http://comsol.com/comsol-multiphysics), (Burlington, MA, USA: COMSOL Inc.)
[29] Yamazaki S, Ushirokawa A and Katoda T 1980 Effect of clusters on long-wavelength optical phonons in Ga1−xInxAs J. Appl. Phys.
51 3722
[30] Aoki K, Anastassakis E and Cardona M 1984 Dependence of raman frequencies and scattering intensities on pressure in GaSb, InAs,
and InSb semiconductors Phys. Rev. B 30 681–7
[31] Waugh J L T and Dolling G 1963 Crystal dynamics of gallium arsenide Phys. Rev. 132 2410–2
[32] Brodsky M H and Lusovisky G 1968 Infrared reflection spectra of Ga1−xInxAs: a new type of mixed-crystal behavior Phys. Rev. Lett. 21
990–3
[33] Murray R, Newman R C, Sangster M J L, Beall R B, Harris J J, Wright P J, Wagner J and Ramsteiner M 1989 The calibration of the
strength of the localized vibrational modes of silicon impurities in epitaxial GaAs revealed by infrared absorption and Raman scattering
J. Appl. Phys. 66 2589–96
[34] Addinall R, Murray R, Newman R C, Wagner J, Parker S D, Williams R L, Droopad R, DeOliveira A G, Ferguson I and Stradling R A
1991 Local vibrational mode spectroscopy of Si donors and Be acceptors in MBE InAs and InSb studied by infrared absorption and
Raman scattering Semicond. Sci. Technol. 6 147–54
[35] Vegard L 1921 Die Konstitution der Mischkristalle und die Raumberfüllung der Atome Z. Phys. 5 17–26
[36] Rashband W S 1997–2015 Imagem, (http://imagej.nih.gov/ij/), (Bethesda, Maryland, USA: U.S. National Institutes of Health)
[37] Thompson M, Galvin G, Mayer J, Peercy P, Poate J, Jacobson A, Cullis A and Iwata N 1984 Melting temperature and explosive
crystallization of amorphous silicon during pulsed laser irradiation Phys. Rev. Lett. 52 2368–3
[38] Kodera H 1963 Diffusion coefficients of impurities in silicon melt Japan. J. Appl. Phys. 2 212
[39] Burton IA 1954 Impurity centers in Ge and Si Physica 20 843–54
[40] Ohtani H, Kobayashi K and Ishida K 2001 Thermodynamic study of phase equilibria in strained III–V alloy semiconductors J. Phase
Equilib. 22 276–86