

Optimizing SiGe–SiO₂ Visible–Short-Wave Infrared Photoresponse by Modulating Interplay Between Strain and Defects Through Annealing

Muhammad Taha Sultan,* Ionel Stavarache,* Andrei Manolescu, Unnar Bjarni Arnalds,* Valentin Serban Teodorescu, Halldor Gudfinnur Svavarsson, Snorri Ingvarsson, and Magdalena Lidia Ciurea*

SiGe–SiO₂-based structures present high interest for their high photosensitivity from visible to short-wavelength infrared. Herein, two postdeposition annealing procedures, that is, rapid thermal annealing (RTA) and rapid-like furnace annealing (FA), are compared. Both RTA and FA are performed at 600 °C for 1 min for SiGe nanocrystals (NCs) formation in SiO₂ matrix in Si/SiO₂/SiGe/SiO₂ structures deposited by magnetron sputtering. The FA imitates RTA resulting in enhanced spectral response. X-ray diffraction, transmission electron microscopy, and Raman spectroscopy are carried out showing Ge-rich SiGe NCs with 11.3 ± 1.2 nm size for RTA and 9.4 ± 0.8 nm for FA. Photocurrent spectra for both structures show several peaks that are annealing dependent. The photocurrent intensity for FA samples is ≈ 7 times higher than RTA samples while cutoff wavelengths are slightly different, that is, 1365 nm for FA and 1375 nm for RTA. The FA structures show (at -1.5 V) over 4 A W^{-1} responsivity at 730 nm, 6.4×10^7 Jones detectivity at 735 nm, and 2.2×10^7 Jones at about 1210 nm. FA structures contain small SiGe NCs with incorporated residual strain, while RTA ones are formed of columnar SiGe NCs separated by SiGeO_x amorphous regions and show increased tensile strain in the SiGe.

high-quality Si/SiO₂ interface with low defect densities,^[9] and it is compatible with the well-established complementary metal-oxide semiconductor (CMOS) technology.^[10,11] The interest in Si_{1-x}Ge_x NCs arises especially because of their potential bandgap tuning by varying NC size and composition (x), as well as by strain engineering.^[11–14] Furthermore, several studies have revealed the effect of strain exerted by the surrounding matrix on NCs as a consequence of the crystallinity, thermal expansion coefficient, lattice mismatch, and degree of oxide crystallization.^[12,14] Attempts have been made to broaden the spectral domain toward short-wave infrared (SWIR) and to improve transport properties by reducing defect concentration, neutralizing dangling bonds, and diffusion of Si/Ge or the formation of unwanted oxides at the interface^[15] for Si_{1-x}Ge_x NCs-based applications in optoelectronics,^[16] light-emitting diodes,^[17] photodetectors,^[18,19] and solar cells.^[20] To achieve this, various approaches

to control the NC size and density have been proposed, for instance by incorporating buffer layer,^[5,21] determining suitable growth or postannealing temperatures, and procedures such as rapid thermal annealing (RTA) and/or conventional furnace annealing (FA),^[22,23] by doping,^[24] by varying the composition, that is, the Si/Ge ratio,^[1,13] and by hydrogen plasma treatment.^[25] SiGe NCs-based structures reported in literature are usually polycrystalline thin films of SiGe or SiGe NCs embedded in an oxide

1. Introduction

Over the past decades, substantial studies have been made on self-assembled Si_{1-x}Ge_x nanocrystals (NCs) in oxide matrices such as SiO₂,^[1–3] and high-dielectric-constant oxides such as TiO₂,^[4,5] Al₂O₃,^[6] and HfO₂.^[7] Of these, SiO₂ is the most studied, as it remains amorphous and provides lateral confinement to the NCs up to high annealing temperatures.^[8] Also, SiO₂ forms a

M. T. Sultan, U. B. Arnalds, S. Ingvarsson
Science Institute
University of Iceland
Dunhaga 3, 107 Reykjavik, Iceland
E-mail: muhammad@hi.is; uarnalds@hi.is

 The ORCID identification number(s) for the author(s) of this article can be found under <https://doi.org/10.1002/adpr.202300316>.

© 2024 The Author(s). Advanced Photonics Research published by Wiley-VCH GmbH. This is an open access article under the terms of the Creative Commons Attribution License, which permits use, distribution and reproduction in any medium, provided the original work is properly cited.

DOI: 10.1002/adpr.202300316

M. T. Sultan, A. Manolescu, H. G. Svavarsson
Reykjavik University
Menntavegur 1, 102 Reykjavik, Iceland

I. Stavarache, V. S. Teodorescu, M. L. Ciurea
National Institute of Materials Physics
405A Atomistilor Street, Magurele 077125, Romania
E-mail: stavarache@infim.ro; ciurea@infim.ro

V. S. Teodorescu, M. L. Ciurea
Academy of Romanian Scientists
050094 Bucharest, Romania

matrix. For polycrystalline SiGe layers, a common downside is the formation of nonradiative recombination centers at the grain boundaries that reduce photogenerated carriers concentration.^[26–28] In contrast, embedded NCs benefit from the passivation provided by the oxide layer and allow for controlling the NC size and minimizing the defects acting as recombination centers.^[29]

One of the standard nanostructuring procedures for the formation of SiGe NCs in oxide matrices is the use of high-temperature annealing to induce nanoparticle clustering and crystallization.^[11,30] In this case, a high thermal budget (TB) is required which can hinder the integration of such structures in sensitive devices (hybrid heterostructures on flexible substrates, nonvolatile memories, and Si-based optoelectronics), while reducing the energy efficient production of devices.^[27,31] In this respect, RTA has proven to be more efficient providing lower crystallization temperatures, that is, a significantly reduced TB, while minimizing film substrate interface reactions, thereby improving the electrical and optical properties of the structures.^[27,31] Such an effect has been attributed to the inherently different ramp rate for RTA compared to conventional furnace^[30,32] where the interfacial regions are more likely to be affected by annealing.^[33] FA produces variation in stress distribution within the structure and thereby affects the NC size and bandgap. It has been argued that in the case of conventional FA, the inherently increased exposure time results in increased NC sizes and stress relaxation accompanied with defects and dislocations,^[34] resulting in a deterioration of the optical and electrical properties.^[35,36] Studies of RTA-annealed samples^[32,37] have provided evidence of better crystallization at lower energy cost in short time while reducing stress in the film. However, with scaling of devices to smaller dimensions, the applicability of RTA processing is continually challenged, especially reducing the TB, as the peak temperature repeatability is challenging, thereby causing difficulty in process control and a degradation of thermal performance.

Light sensitivity performances is a very important characteristic of these materials. The literature presents a very wide range of responsivities for SiGe-based structures, from few mA/W up to more than a thousand A/W, the responsivities being strongly dependent on the manufacturing technology and parameters. So, Si-Ge photodetector under zero-bias illumination achieves 0.35 A W^{-1} at 1180 nm wavelength and 0.043 A W^{-1} around 1270 nm ones.^[38] Si-Ge phototransistors fabricated on $20 \text{ } \Omega \cdot \text{cm}$ Si substrates have 1.35 A W^{-1} responsivity at 50 MHz under phototransistor mode operation with 850 nm multimode fiber illumination.^[39] By SiGe-Si multiple-quantum-well heterojunction phototransistor, a responsivity over 1 A W^{-1} at 850 nm wavelength was obtained.^[40] Using SiGe heterojunction, bipolar phototransistors with multiple Ge quantum dot layers embedded in the base-collector junction achieve at normal incidence the responsivity of 3 mA W^{-1} at $1.3 \text{ } \mu\text{m}$. With waveguide geometry coupling light through the device edge, even higher responsivities of 130 mA W^{-1} at $1.3 \text{ } \mu\text{m}$ and 5 mA W^{-1} at $1.55 \text{ } \mu\text{m}$ were obtained.^[41] Back-side-illuminated Ge photodetector grown on Si wafers using thin SiGe buffer layers has a responsivity of 0.37 A W^{-1} at 0 V and 0.57 A W^{-1} above 2 V reverse bias at $1.3 \text{ } \mu\text{m}$.^[42] By another phototransistor photodetector in a $0.35\text{-}\mu\text{m}$ SiGe BiCMOS technology, that integrates a surface photodetector and a conventional SiGe heterojunction phototransistor,

highly increased responsivities of 5.2 A W^{-1} at 850 nm wavelength and 9.5 A W^{-1} at 670 nm are seen.^[43] Low-dark current Ge pin photodiodes on Si substrate achieve responsivities of 0.37 A W^{-1} at 0 V bias for the Ge diodes with 20 nm Si cap and 0.57 A W^{-1} above 2 V reverse bias for Ge diodes with 200 nm Si cap.^[44] Also, on the *p*-type SiGe quantum well infrared photodetectors on strained-Si-on-insulator substrates, a peak responsivity of 3.6 mA W^{-1} at 77 K and 0.5 V bias were reported.^[45] A remarkable responsivity of 1032 A W^{-1} at 2.8 V bias and 606 A W^{-1} at 1 V bias with input optical power of -50 dBm (10 nW) were obtained on Si/Ge waveguide integrated phototransistor with an engineered electric field distribution.^[46]

In this study, we investigate the beneficial effect of rapid-like FA compared to RTA on the photoconductive response of SiGe NC-based structures. The films were deposited via magnetron sputtering (MS) at room temperature (RT) followed by subsequent ex situ annealing at 600°C for 1 min. The structure and morphology of the films were studied with X-ray diffraction (XRD), Raman spectroscopy, and transmission electron microscopy (TEM). The results show Ge-rich SiGe NCs of $11.3 \pm 1.2 \text{ nm}$ (RTA) and $9.4 \pm 0.8 \text{ nm}$ (FA) size, and the strain was found to be dependent on the annealing procedure. Photoconductive measurements carried out on both types of structures treated by RTA and FA show a broad spectral response and cutoff wavelength 1375 and 1365 nm, respectively. The calculated responsivity and detectivity points out that FA samples compared with RTA ones present higher performances reaching 4 A W^{-1} and 6.4×10^7 Jones at around 730 nm and 1.5 A W^{-1} and 2.2×10^7 Jones in SWIR.

2. Results and Discussion

Figure 1 shows X-ray reflectivity (XRR) scans for as-grown structures with thin ($\approx 20 \text{ nm}$) SiGe layer (SiGe_{thin}). XRR fitting and results are SiGe layer thickness of $20 \pm 0.47 \text{ nm}$ with surface roughness of 1.3 ± 0.02 and $75 \pm 1.3 \text{ nm}$ buffer SiO₂ and $42 \pm 0.3 \text{ nm}$ top SiO₂.

Figure 2 shows grazing incidence XRD (GiXRD) scans for thin structures ($\approx 20 \text{ nm}$ SiGe layer), namely, for as-grown ones (S_{AG}) and for those annealed at 600°C by RTA (S_{RTA}) and in furnace

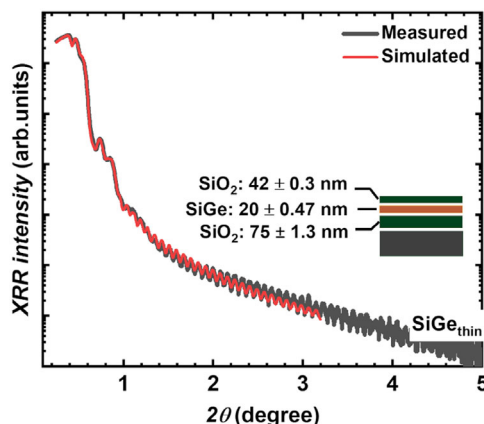


Figure 1. XRR plots measured on as-grown Si/SiO₂/SiGe/SiO₂ structures with a thin SiGe layer of $\approx 20 \text{ nm}$ thickness (labeled SiGe_{thin}). The black line represents the experimental data and red line shows the simulation ones fit.

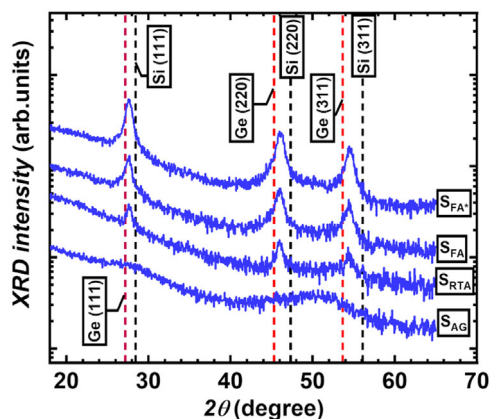


Figure 2. GiXRD scans: • S_{AG} , as-grown structure, • S_{RT} , S_{FA} 600 °C, 1 min thin structures (20 nm SiGe layer); • S_{FA*} , thick structure (200 nm SiGe layer) 600 °C FA, 1 min. SiGe crystallographic peaks (111), (220), and (311) are located between the standard positions of cubic Si ($2\theta = 28.443^\circ$, 47.304° , and 56.122° ; JCPD card: 00-27-1402) and cubic Ge ($2\theta = 27.284^\circ$, 45.306° , and 53.683° ; JCPD card: 00-004-0545), by the dashed lines.

(S_{FA}), along with GiXRD scans for thick structures (≈ 200 nm SiGe layer) annealed in furnace (S_{FA*}) also at 600 °C. For the as-grown structure, broad reflections implying the presence of amorphous SiGe in the structure (Raman spectroscopy is more sensitive, showing start of crystallization from the deposition without annealing, i.e., small SiGe NCs are already present in the as-grown sample).

For structures S_{RT} and S_{FA} , annealed at 600 °C for 1 min, the diffractograms present three distinctive peaks corresponding to $Si_{1-x}Ge_x$ being located between the standard cubic Si and Ge peak positions corresponding to (111), (220), and (311) planes, according to JCPD card: 00-027-1402 and 00-004-0545, respectively. SiGe NCs' size, composition, and strain were determined using crystallographic plane (111), using Scherrer's equation with shape factor $k = 0.9$, Vegard's law Equation (1) and (2), respectively. We acknowledge that the aforementioned approach to determine crystallite size and composition is indecisive, as the

consequence of microstrain broadening, lattice imperfections, and other instrumental error cannot be disregarded. However, the NC size is in good agreement with TEM analysis, whereas the determined concentration x in Si_xGe_{1-x} is within 0.5% error. The presence of crystalline peaks shows the formation of SiGe NCs. The d -spacing and SiGe lattice constant a were calculated for samples S_{RT} and S_{FA} using the crystallographic peak (111) positioned at 27.597° and 27.533° , respectively, and using Gaussian fit. The obtained values for S_{RT} and S_{FA} are $a_{RT} = 0.3230$ nm and $a_{FA} = 0.3223$ nm, the known values for Ge and Si are 0.326 and 0.314 nm, respectively, implying Ge-rich SiGe NCs. This also can be seen by the crystallographic (111) peak positions located closer to Ge standard position (27.284°). For finding the composition x and strain ϵ in these structures, we used

$$a_{SiGe} = 3^{1/2}d = 5.4309 + 0.20032x + 0.02674x^2 \quad (1)$$

$$\epsilon = (d_f - d_b)/d_b \quad (2)$$

where a_{SiGe} is the lattice constant. d_f and d_b represent the d -spacings for film and bulk SiGe.

The strain ϵ in the S_{FA} structure was determined to be $\epsilon_{FA} = 0.80\%$ compared to that of S_{RT} $\epsilon_{RT} = 1.02\%$ and compositions were determined to be $x_{FA} = 69\%$ and $x_{RT} = 74\%$. In conclusion from Figure 4, one can see that all samples, excepting the as-grown one S_{AG} , present SiGe NCs rich in Ge after RTA. In addition, by comparing samples S_{RT} and S_{FA} with similar thicknesses and the same temperature annealing (by RTA and FA), their compositions and strains are different.

Effects of stress field developed in the SiGe film plane as well as strain exerted by the SiO_2 matrix in structures annealed by RTA are revealed by cross-sectional transmission electron microscopy (XTEM). **Figure 3a** shows XTEM micrographs recorded for S_{RT} structure. An interesting self-assembly of periodically arranged columnar NCs with the long axis perpendicular to film is observed. The graphs show no obvious presence of defects. The formation of such quasiperiodic NCs has been attributed to the crystallization process during annealing which develops a stress field in the SiGe plane along with the strain

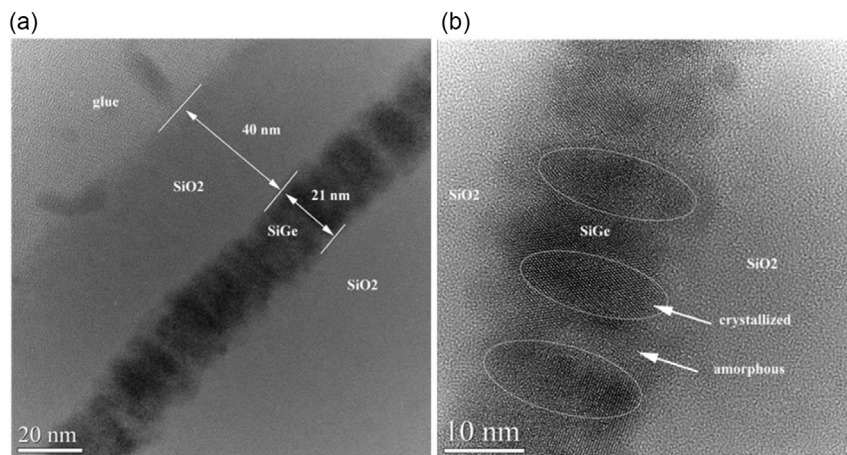


Figure 3. a) XTEM image of the S_{RT} structure annealed at 600 °C by RTA for 1 min. b) High-resolution TEM image showing the columnar assembly of SiGe NCs (shown by the marked areas), with an average periodicity of 12.5 nm separated by amorphous regions (with $SiGeO_x$).

exerted by the SiO_2 matrix resulting in the formation of corrugated edges by strain relaxation.^[3,47] The average size of the NCs is ≈ 10 nm with ≈ 2 – 5 nm distance between them (Figure 3b) comprising amorphous SiGeO_x with Si excess by oxygen diffusion from the surrounding oxide matrix, thus forming the inter-columnar film. The Si excess in SiGeO_x is explained in literature by SiO_2 formation that is more exothermic than that of GeO_2 and annealing at elevated temperatures will result in Ge segregation, leading to Ge-rich SiGe NCs.^[8] In addition, the size of NCs determined by Gaussian fitting of the (111) XRD peak, and using the Scherrer equation with an instrumental error (beam broadening of 0.012 and a shape factor k of 0.9), is in good agreement with results obtained from the lattice fringe area of NCs in the high-resolution transmission electron microscopy (HRTEM) image. For S_{RTA} and S_{FA} structures, the SiGe NC size varies, sizes being

11.3 ± 1.2 and 9.4 ± 0.8 nm, respectively, in good agreement with the results observed by TEM (Figure 3). The formation of columnar SiGe NCs during annealing can be due to SiGe nanocrystallites grown during MS deposition (revealed by 289.8 cm^{-1} Gauss deconvolution peak from Raman spectra of S_{AG} sample in Figure 5b), also observed in films of SiGe NCs in a TiO_2 matrix.^[4]

These nanocrystallites act as seed crystallites, resulting in heterogeneous nucleation at preferential sites, thereby reducing the nucleation barrier upon annealing. One can see that thin S_{RTA} structure is formed by the quite regular arrangement of columnar SiGe NCs with ≈ 12.5 periodicity in the film plane. SiGe NCs have 20 nm length (perpendicularly on film) and 5–6 nm width. For thick S_{FA} samples (≈ 200 nm SiGe film) (Figure 4a) annealed in furnace, the stress field developed is partially relaxed by the formation of defects in NCs mainly by coalescence of small NCs during annealing.^[3]

In Figure 4b, SiGe NCs size are evaluated from the lattice fringe contrast and it is similar to that obtained from XRD measurements. In contrast to thin S_{FA} structure, the film is polycrystalline, uniform, and shows no texture.^[3]

To elucidate the variation in strain and the composition of the $\text{Si}_{1-x}\text{Ge}_x$ NCs, Raman spectroscopy analysis was performed the results being shown in Figure 5a. The Raman peak positions were obtained by deconvoluting spectra, an example being shown in Figure 5b for S_{AG} structure. Four distinct peaks are observed, which are positioned around standard values of 300, 400, and 500 cm^{-1} corresponding to the first-order Ge–Ge (ω_{GeGe}), Si–Ge (ω_{SiGe}), and Si–Si (ω_{SiSi}) optical phonon modes of a crystalline $\text{Si}_{1-x}\text{Ge}_x$ alloy, and the peak at 521 cm^{-1} is due to Si substrate contribution. In Table 1, the results obtained for Raman spectra presented in Figure 5 are summarized. The shift in Raman peaks is attributed both to the variation in strain state and to the Ge content in $\text{Si}_{1-x}\text{Ge}_x$. The as-grown structure S_{AG} reveals Raman peak doublets for both ω_{GeGe} and ω_{SiGe} , where the peak at 289.81 cm^{-1} represents the Ge–Ge vibration mode in Ge-rich SiGe NCs, being shifted to lower energy (with

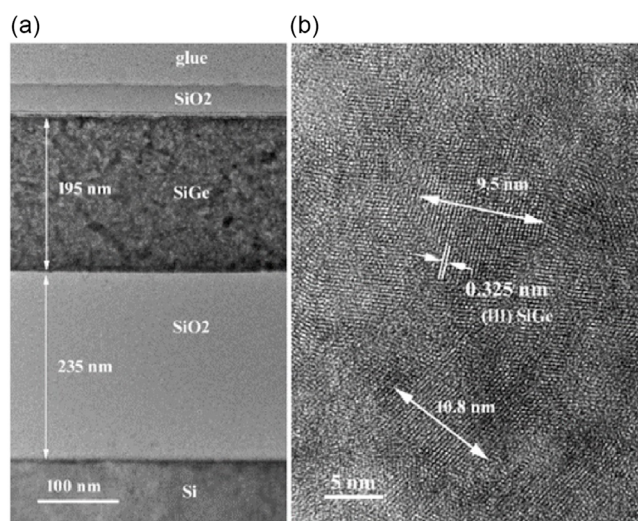


Figure 4. XTEM image of thick S_{FA} samples: a) low-magnification image and b) HRTEM detail.

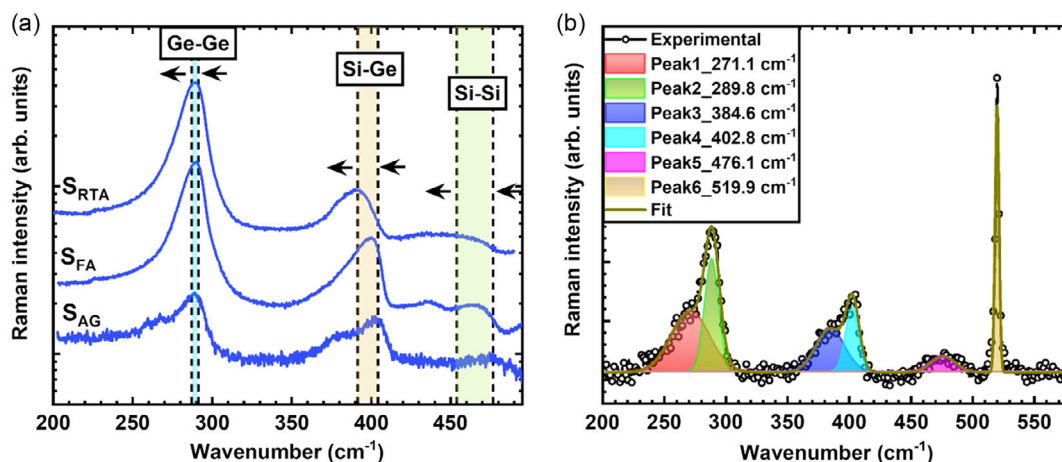


Figure 5. Raman spectra of S_{AG} , S_{FA} , and S_{RTA} structures. a) The peaks around 300, 400, and 500 cm^{-1} correspond to first-order Ge–Ge (ω_{GeGe}), Si–Ge (ω_{SiGe}), and Si–Si (ω_{SiSi}) optical phonon modes of a crystalline $\text{Si}_{1-x}\text{Ge}_x$ alloy (labeled respectively). The highlighted regions illustrate the shift in ω_{GeGe} , ω_{SiGe} , and ω_{SiSi} . b) Processing of Raman spectra for the S_{AG} structure, initially by subtracting background and followed by deconvolution of Gaussian peaks with corresponding Raman peaks tabulated in the graph.

Table 1. Data extracted from Raman spectroscopy analysis in Figure 5 for the S_{AG} , S_{RTA} , and S_{FA} structures together with composition (x_{Ge}) and strain in $Si_{1-x}Ge_x$ NCs.

Sample ID	ω_{GeGe} [cm^{-1}]	ω_{SiGe} [cm^{-1}]	ϵ [%]	x_{Ge} [%]
S_{AG}	289.81	402.7	0.53	61
S_{FA}	289.39	400.9	0.74	64
S_{RTA}	288.58	392	1.50	78

respect to the bulk value of 300.1 cm^{-1} due to quantum confinement and strain.^[5,48] The shoulder at 271.1 cm^{-1} indicates the disorder in SiGe component,^[49] while the peak from 476.1 cm^{-1} is due to amorphous Si.^[50] However, the structures processed via RTA and FA showed a redshift of the ω_{GeGe} and ω_{SiGe} Raman peaks compared to the results obtained for the S_{AG} structures, indicating the increase (reduction) in tensile (compressive) strain. Utilizing the data obtained from the Raman spectroscopy using Tsang's model (Equation (3) and (4) below),^[5] the strain ϵ present in the structures and composition (Ge content x_{Ge}) of $Si_{1-x}Ge_x$ compound were obtained and are presented in Table 1.

$$\omega_{GeGe} = 280.3 + 19.4x - 450\epsilon \quad (3)$$

$$\omega_{SiGe} = 400.1 + 24.5x - 4.5x^2 - 33.5x^3 - 570\epsilon \quad (4)$$

So, one can see that all samples present strain that is increased by FA and RTA processing, the highest tensile strain being by RTA. The Ge concentration of 0.74 obtained via XRD is practically similar to that from Raman spectroscopy (0.78) for RTA sample.

Both the composition and strain account for the shifts in the peak position.

2.1. Photoelectrical Characterization

To explore the influence of the annealing procedures on the photosensitivity of the structure, current–voltage (I – V) characteristics were measured in the dark and under illumination with an incandescent lamp (20 W) after Al electrodes deposition in a coplanar configuration with an active area of $8 \times 10\text{ mm}^2$. Figure 6 shows I – V characteristics for bias voltage in the range of -1.5 – 1.5 V . For the S_{AG} structure (Figure 6a), the I – V plot shows ≈ 2 orders of magnitude increase in current intensity upon illumination for both voltage polarities (from -1.5 to $+1.5\text{ V}$), while for S_{FA} and S_{RTA} structures (Figure 6b,c), the photocurrent increases by 1 and 3 orders of magnitude, respectively. The low photocurrent for S_{FA} with respect to S_{AG} can be attributed to residual strain acting as nonradiative recombination centers,^[28,41] in contrast to S_{RTA} where the main contribution is given by SiGe NCs formation. One can observe differences in I – V curves taken in the dark and under illumination on FA, RTA, and FA*. So, FA samples present high dark current and lower photocurrent compared with RTA ones (Figure 6b compared with Figure 6c). This discrepancy can be explained considering the two annealing processes. FA involves slower ramp rate than RTA, that commonly promotes the formation of defects^[11] including recombination centers, within the oxide matrix.^[11,47] So, the lower photocurrent observed in the FA film (compared to RTA films) is due to the presence of those defects acting as recombination centers for photogenerated carriers. The higher

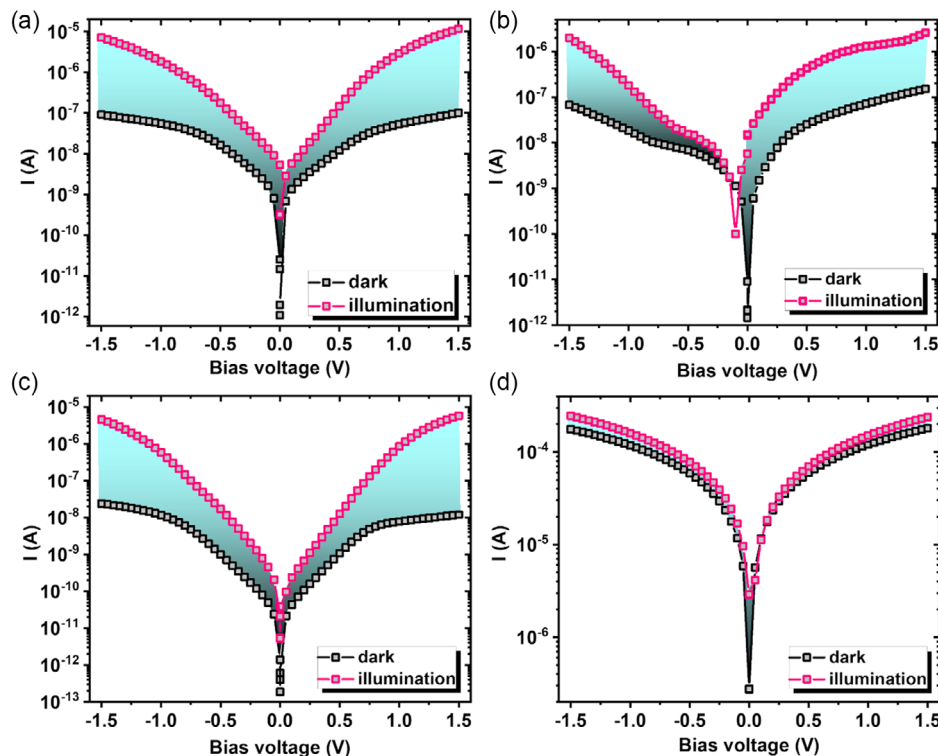


Figure 6. RT I – V characteristic, in dark and under illumination with an incandescent lamp (20 W): a) S_{AG} , b) S_{FA} , c) S_{RT} , and d) S_{FA}^* .

dark current for FA films compared to RTA is also due to participation of some defects and Ge-related states formed by fast diffusion^[51] to charge carrier transport in the dark. The I - V curves recorded on FA* structure with the thick SiGe film (Figure 6d) can be similarly explained. The dark current for FA* structure is with 3 orders of magnitude higher than that for FA while the photocurrent is increased with only two orders, both currents being due to the increased contribution of the thick SiGe layer.

Figure 7 shows the RT photocurrent spectra, with -1.5 V external bias, for thin structures (≈ 20 nm SiGe layer) in the as-grown state and those annealed using RTA and FA. For comparison, we also represented the spectrum obtained on the FA* sample in which the thickness of SiGe increases significantly at 200 nm. A broad interval of photosensitivity from ≈ 400 to 1375 nm was observed for all structures (Figure 7a–d), with a spectral responsivity (R) cutoff wavelength (presented in insets) higher than the Si absorption limit (≈ 1100 nm). To evaluate the origin of the peaks, the photocurrent spectra were deconvoluted using Gaussian fitting with the Origin software. The peaks (and shoulders) located below 1100 nm are attributed to Ge-related states in the SiO₂ layers by fast diffusion^[51] and, to strain in SiGe NCs and/or NCs–oxide interface. The shoulder positioned at 1100 nm represents the contribution of the Si substrate^[11] and peak located around 1200 nm is due to Ge-rich SiGe NCs. It is observed that for S_{FA} sample the overall spectral intensity increased almost three times compared to that observed for

S_{AG} structures, along with an increase in the relative intensity of the SiGe NCs peak 6 with respect to peaks 1–4 produced by strain and interface.^[33,34] implying residual strain within the structure. For RTA structure, the spectral intensity was reduced to $\approx 1/4$ of that for S_{AG} structure, with a relative increase in peak intensity from the SiGe NCs' peak 6 with respect to peaks 1–4. At the same time, the reducing of 1–4 peaks' intensity could be induced by healing localized states and (low degree) stress relaxation in structure (analyzed by TEM and supported by results from XRD and Raman measurements). In the literature,^[36,37] it has been argued that annealing in a conventional furnace at high temperatures and for a long time (several minutes to hours) results in stress relaxation with increased NCs size, thereby resulting in an increased defect concentration and dislocation density in NCs and at the interface. In comparison, RTA for short exposures tends to form rather small NCs at the beginning (growing further during RTA) while releasing residual strain in the structure that is not in equilibrium.^[32,37] Considering the above two cases, for our case, we can state that transient FA for 1 min may have provided enough energy to obtain crystallization without reaching equilibrium and thereby limiting the stress relaxation in the structures, that is, remaining residual stress.

The approximation of sample responsivity was obtained by applying the formula $R = I_{\text{photo}}(\lambda)/P_{\text{in}}(\lambda)$, where I_{photo} represents the spectral photocurrent and P_{in} represents the illumination power at normal incidence which falls on the active area of sample. An important observation is the fact that the responsivity for

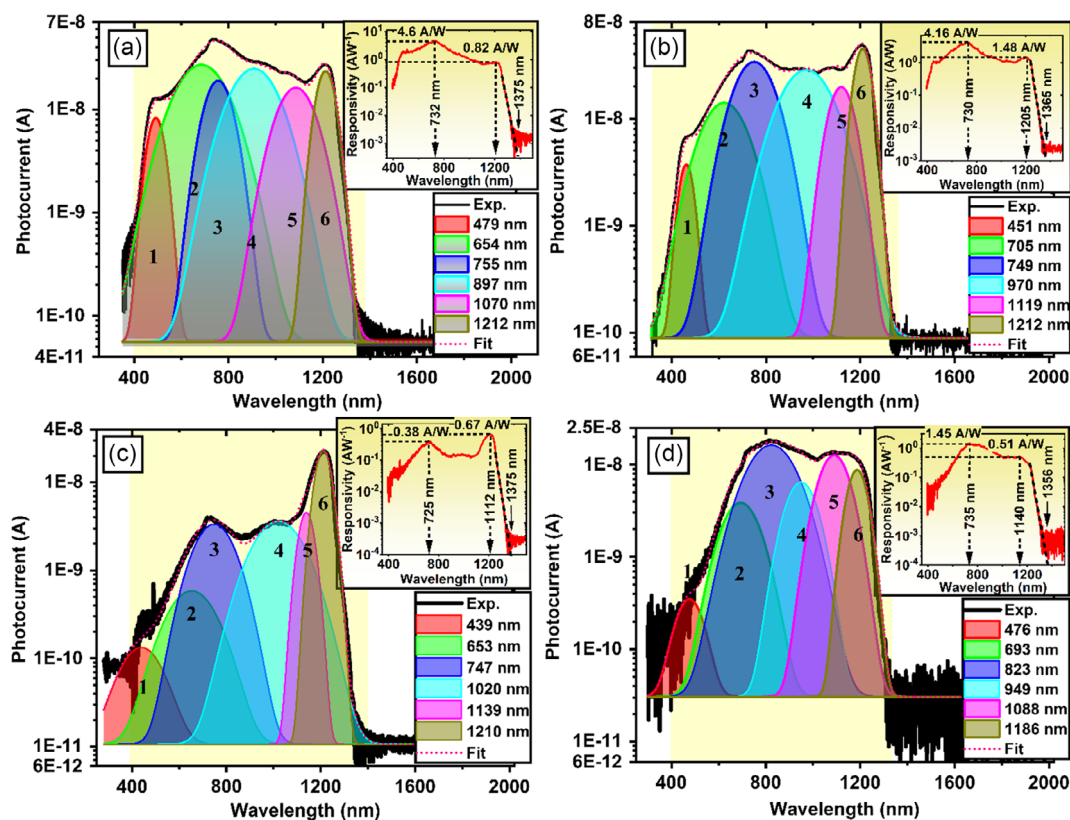


Figure 7. RT photocurrent spectra obtained on a) S_{AG}, b) S_{FA}, c) S_{RT}, and d) S_{FA*} structures. The inset shows the spectral responsivity cut-off wavelength for the respective structures.

the FA samples is superior to that obtained on the RTA samples both in the visible and in the NIR domain. The FA structures have a maximum responsivity greater than 4 A W^{-1} at 730 nm, and in the infrared, the maximum responsivity is $\approx 1.5 \text{ A W}^{-1}$ determined at 1205 nm. The intensities weights change in the case of RTA structures, where the maximum responsivity greater than 0.6 A W^{-1} is determined at 1200 nm, and the visible peak is almost 0.4 A W^{-1} at 730 nm. These values are consistent with those reported in peer-reviewed journals for SiGe-based structures, discussed in the Introduction.

To validate the peak origin due to the SiGe NCs, we explore the temperature dependence of photocurrent spectra for the S_{FA} structures in **Figure 8**. A blueshift in the position of the peak corresponding to NCs, from 1214 to 1168 nm and then to 1127 nm, is observed when the photocurrent spectra were measured at 300, 200, and 100 K, respectively (together with the shift of the cut-off wavelength from 1365 to 1324 and then to 1243 nm), implying the quantum confinement effect.^[1,13,52] At the same time, the positions of peaks attributed to strain and interface show insignificant change. Additionally, to elucidate further the origin of the NCs peak, the conduction mechanism between different electronic states can be taken into consideration. It is acknowledged that at low temperatures (below 200 K), the hopping of charge carriers to interface-related states is weak while the thermally activated processes dominate (2–3 orders of magnitude stronger than hopping at a given low T).^[49] Therefore, at low T , the exciton is more localized in NCs under illumination. This can explain the relative increase of NCs' peak intensity at low T compared to that from the localized states and/or strain (Figure 8). At higher temperatures, that is, above 200 K, the interface localized states can contribute to the transfer of charge carrier from NC to NC, resulting in an increased NC peak intensity.

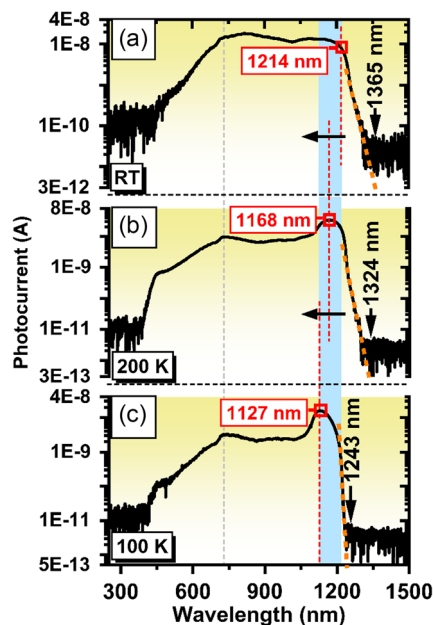


Figure 8. Photocurrent spectra measured for the S_{FA} structure at 300, 200, and 100 K. The dashed line represents the peak positions, and the arrows represent the shift in peak position as a function of temperature.

The peaks attributed to strain, Ge-related states, and interfaces present different relative intensities than as-grown samples by different stress fields correlated with defects healing. Strain present in structures causes NCs to deform in size and shape which therefore results in altering the structures' bandgap. We consider that in the case of the S_{FA} structure (1 min FA) the stress is partially relaxed, whereas in case of the S_{RTA} structure, the strain exerted by the oxide matrix tends to relax in the film plane by forming "corrugated" NC edges while the NCs are stress free on the normal direction.^[3] This is also supported by Raman spectroscopy analysis (Table 1), implying an increase of tensile strain from S_{AG} to annealed structures. Therefore, one can observe changes in peak intensity as a strain indicator for S_{FA} and S_{RTA} structures.

To analyze the effect on peak position due to strain, the structure S_{FA*} having thick SiGe layer was considered. Such structures were studied in ref. [53], where the strain is significantly relaxed by forming planar defects. Figure 7d shows photocurrent spectrum for the S_{FA*} structure with a thick SiGe layer. Unlike S_{FA} (Figure 7b), the sensitivity observed in S_{FA*} (Figure 7) has a cutoff wavelength at 1356 nm, with the photocurrent intensity being reduced 7 times. The Gaussian fit of spectrum shows peak 5 due to the Si substrate that further contributes to the alteration in the photocurrent intensity (by capacitive coupling)^[54] and also peak 6 that is due to SiGe NCs. Peaks 1–4 obtained at wavelengths below 950 nm are mainly due to Ge-related states (low strain).^[51] The relative increase of peaks 2–4 intensity to that of peaks 5–6 (Figure 7d compared to Figure 7b) can be related to the defects in NCs and their surface passivation. The defects contribution explains the low photocurrent with respect to dark current (Figure 7d).

Figure 9 shows the specific detectivity (D^*), which is the key parameter to evaluate the possibility of using these materials as optical sensors. The detectivity was determined from recorded spectral photoresponse using the formula

$$D^* = \frac{A^{1/2} R}{(2qI_{\text{dark}})^{1/2}} = \frac{R}{(2qJ_{\text{dark}})^{1/2}} \quad (5)$$

where I_{dark} is the current in the dark, R is responsivity, and A is the effective area. J_{dark} is the dark current density, and q is electronic charge.

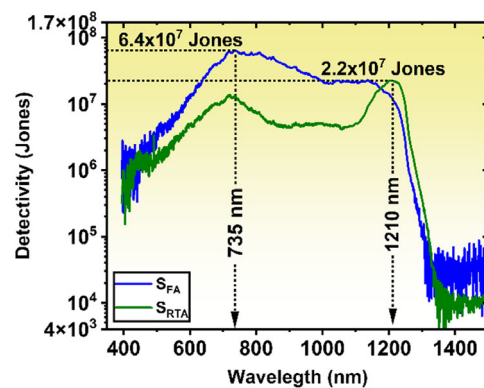


Figure 9. Calculated spectral specific detectivity at RT for S_{FA} and S_{RTA} samples.

The detectivity spectra present two important peaks at about 735 nm and around 1210 nm for both samples FA and RTA, respectively. The highest peak for FA samples is at 735 nm, in the visible range, reaching 6.4×10^7 Jones and a shoulder in the near infrared. In comparison, the RTA samples present the highest peak in near infrared at about 1210 nm, where the detectivity value reaches 2.2×10^7 Jones and one in visible at 735 nm where detectivity value exceeds a little over 1×10^7 Jones. The obtained results are in good agreement with the results reported in literature.^[55]

Therefore, the structure treated by RTA (S_{RTA}) has shown increased tensile strain, as observed from the shift of peaks in Raman spectra. The crystallization process during RTA, when a stress field develops in the SiGe film plane, is therefore responsible for producing a quasiperiodic SiGe NC arrangement. TEM images reveal the presence of SiGe NCs which are stress free in the normal direction on the film and show no internal defects, and the structure has been relaxed by forming corrugated edges of the SiGe layer, thus minimizing the stress field developed during RTA. This results in the significant increase in NC peak and the reduction of strain and interface peaks.

In contrast, structures processed by FA showed enhancement of intensity of peaks due to strain. This can be explained by the short annealing time for NC formation without reaching equilibrium (meaning restricted stress relaxation and therefore residual strain in the structure).

3. Conclusion

In this work, we investigated the dependence of vis–SWIR sensitivity of SiGe NCs in SiO_2 matrix on thermal treatment by RTA and in furnace using the rapid-like procedure imitating RTA. The structures $\text{Si/SiO}_2/\text{SiGe/SiO}_2$ (20 nm SiGe layer) were deposited by MS, namely, SiO_2 layers by reactive direct current MS (dcMS) and SiGe layer by dcMS and high-power impulse MS (HiPIMS) from Si and Ge targets, respectively. The samples were thermally annealed, some of them by RTA and the others by FA at the same temperature of 600 °C for 1 min. The structure and morphology studies show crystalline Ge-rich SiGe NCs with 9.4 ± 0.8 and 11.3 ± 1.2 nm size for S_{FA} and S_{RTA} , respectively. We have shown that the S_{FA} structures contain small SiGe NCs and have incorporated residual strain. On the contrary, S_{RTA} structures are formed of the columnar assembly of SiGe NCs with an average periodicity of 12.5 nm separated by amorphous regions (SiGeO_x) and show increased tensile strain. During the crystallization process (during RTA), the stress field develops in the SiGe film plane. At the same time, we demonstrated that in the normal direction on the film, SiGe NCs are stress free, they don't contain internal defects, and structure has been relaxed by forming corrugated edges of the SiGe NCs, thus minimizing the stress field developed during RTA.

The results regarding photosensitivity of S_{FA} and S_{RTA} samples are in good agreement with the above discussion. S_{FA} samples present a broad spectral sensitivity (400–1365 nm) with increased photocurrent intensity (≈ 7 times). The photocurrent spectrum taken on S_{RTA} structures also present a broad spectral sensitivity (400–1375 nm) and significant increase in NCs' peaks together with the reduction of peaks corresponding to strain and interface peaks.

The highest responsivity and detectivity were obtained on FA samples reaching more than 4 A W^{-1} and 6.4×10^7 Jones in visible range (around ≈ 730 nm) and in SWIR those were found to be 1.5 A W^{-1} and 2.2×10^7 Jones, respectively.

These results provide suitable approaches for applications in discrete optical sensors and integrated SWIR photodetectors, the structure preparation being compatible to CMOS technology and requiring a low TB.

4. Experimental Section

Materials and Method: The structures investigated ($\text{Si/SiO}_2/\text{SiGe/SiO}_2$) in this study comprised SiGe layer sandwiched between SiO_2 layers deposited on $10 \times 10 \text{ mm}^2$ Si (001) substrates with resistivity of 10–100 Ωcm (Figure 10a).

Preparatory cleaning of substrates before loading into the deposition chamber was carried out in an ultrasonic bath by immersing samples in acetone, methanol, and isopropanol followed by deionized water and blow drying with N_2 . Before deposition, the substrates were prebaked at 600 °C for 30 min in vacuum chamber at base pressure of 5×10^{-7} Pa and then cooled down before proceeding with deposition. The deposition of SiO_2 layer (on Si substrate in $\text{Si/SiO}_2/\text{SiGe/SiO}_2$ structure) was carried out by reactive dcMS at a power of 150 W from a Si target. Argon was used as the working gas with a flow rate of $q_{Ar} = 30$ sccm, along with O_2 as the reactive gas with a flow rate of $q_{O_2} = 2$ sccm. The deposition pressure in the chamber was adjusted to a value of 0.8 Pa. The SiGe layer (thin 20 nm) was cosputtered from individual Si and Ge targets, using dcMS and HiPIMS respectively, and the deposition rates were adjusted to obtain a 50:50 SiGe atomic ratio. The Si deposition was made at a constant power of 180 W using Advanced Energy MDX500 power supply. The Ge deposition was performed using HiPIMS with a SPIK1000A pulse unit (Melec GmbH) operating in unipolar negative mode (charged by a DC power supply, ADL GS30) at a constant square-wave pulse of constant voltage. A combined current transformer and a voltage divider unit (Melec GmbH) were used to monitor the discharge current and voltage. The data were recorded using a digital storage oscilloscope (Agilent 54624 A) and a PCI-e DAQ card. The pulse length, repetition frequency, and pulse amplitude were set to 100 and 300 Hz at 445 V, giving an average power of 103 W with a current and peak power density of 233 mA cm^{-2} and 107 W cm^{-2} over the whole target area. The cap SiO_2 was deposited on SiGe layer under similar conditions as the buffer one (on Si substrate). Thick $\text{Si/SiO}_2/\text{SiGe/SiO}_2$ structure (SiGe ≈ 200 nm) was prepared for strain relaxation comparison with thin structures. The Al electrodes (deposited by electron beam evaporation) were $1 \times 10 \text{ mm}^2$ in size, in a coplanar configuration with an active area of $8 \times 10 \text{ mm}^2$ between them shown in Figure 10a. The transport mechanisms between electrodes were 1) inside the active layer (formed by elongated SiGe NCs, perpendicularly oriented on substrate, being separated by amorphous SiGeO_x (Figure 3)), the transport takes place by electron tunneling between SiGe NCs and 2) inside the adjacent SiO_2 layers, the transport takes place by hopping on Ge-related states produced by Ge fast diffusion.^[51] In the case of photocurrent spectra (Figure 7), the contribution of Si substrate was visible due to the absorption of light with corresponding wavelengths. The contact junctions were Schottky barriers (positioned back to back). Schematic of the device band structure is drawn in Figure 10b.

We have to highlight that the annealing by RTA and rapid-like FA was made before the Al electrode depositions. Two variants of annealing methods were utilized, that is, RTA and rapid-like FA. We denoted the structure in the as-grown state as S_{AG} and the structures annealed under RTA and FA were referred to as S_{RTA} and S_{FA} , respectively. For the FA-processed structure, the furnace was preheated and held at the desired temperature until a uniform heat distribution over the alumina tube was obtained. The sample was then inserted and held in the tube for 150 s in N_2 gas ambient flow. The annealing time included 90 s of ramp-up time ($\approx 5\text{--}6^\circ\text{C s}^{-1}$) (determined earlier via a dummy sample attached to a thermocouple) and 60 s of the intended annealing time. After annealing, the alumina tube

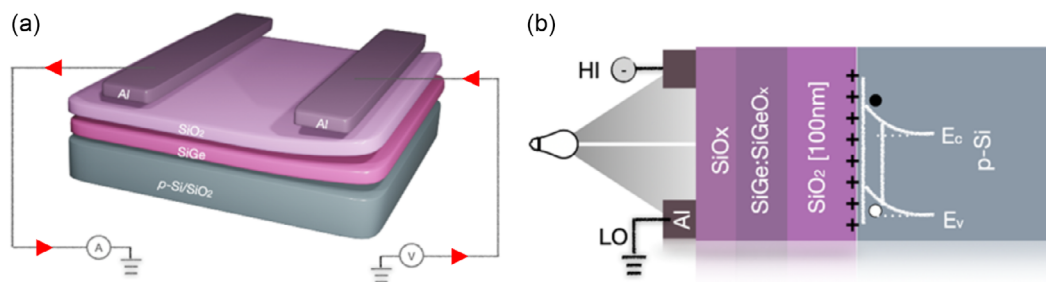


Figure 10. a) Schematic of structure (Si/SiO₂/SiGe/SiO₂), with coplanar Al contacts (1 × 10 mm²), and b) the device band structure.

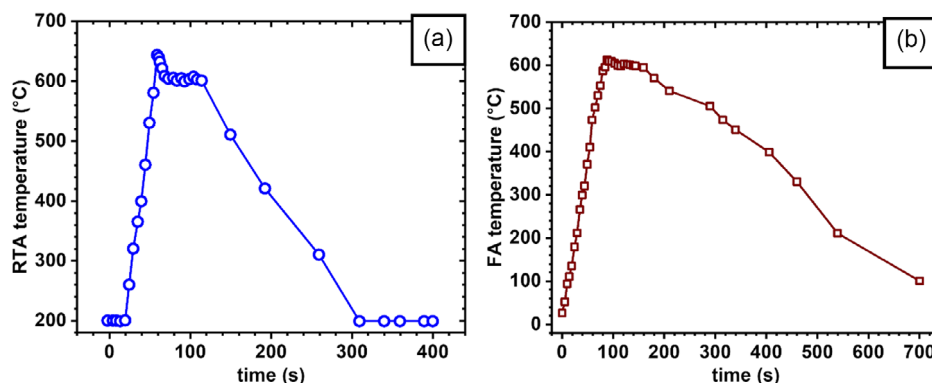


Figure 11. Annealing cycle for: a) RTA (open circles) determined using pyrometer and b) FA (open squares) determined using a thermocouple attached to the dummy sample. It should be noted that the pyrometer cannot read temperature values below 200 °C.

was extracted and was allowed to cool down in air while maintaining the N₂ flow. In case of RTA, the sample was initially heated to 600 °C over a period of 55 s using a power setting of 22% (i.e., a ramp rate of ≈13–16 °C s^{−1}) in N₂ ambient, followed by a holding temperature using pyrometer control for 60 s.

After the heating process, the samples were allowed to cool down in N₂ ambient. The annealing cycles for both RTA and FA are presented in **Figure 11a,b**, respectively. The temperature of 600 °C was chosen for nanostructuring Ge-rich SiGe films, and it was suitable for obtaining optimum results regarding film crystallization (including stress and NC sizes) correlated with photoconductive properties (*I*–*V* photocurrent characteristics and photocurrent spectra) as we have shown in our previous work.^[53]

Characterization: Structural characterization of the films was conducted using XRR and GiXRD using a Panalytical X'pert diffractometer (CuKα, 0.15406 nm), Raman spectroscopy (Horiba LabRam Evolution), XTEM, and energy-dispersive X-ray spectroscopy (Zeiss Supra 35). A detailed description is given in ref. [3]. Raman spectroscopy analysis was made using an air-cooled frequency-doubled Nd:Yag (100 mW/1 MHz) 532 nm excitation line along with 1800 mm high-resolution grating. Each Raman measurement was composed of 20 integrated spectra with an acquisition time of 30 s each. The laser power setting was adjusted to 3% in order to avoid any heating effects during measurements. The electrical characterization setup contained an electrometer (Keithley 6517A), a LakeShore331 temperature controller, and a Janis CCS-450 cryostat. The current–voltage (*I*–*V*) characteristics were measured under illumination using an incandescent tungsten halogen lamp of 20 W. The optical power on the active area of sample was 20 mW cm^{−2}, measured using a Thorlabs powermeter PM100D coupled with a thermal sensor S401C. The photocurrent spectra were recorded under illumination by an incandescent quartz tungsten halogen lamp of 100 W, coupled with an SR830 lock-in amplifier, an SR540 chopper, and MS257 monochromator with an adjustable set of filters (details in ref. [11]).

Acknowledgements

This work was funded by the Icelandic Research Fund, grant nos. 218029-051 and 207111-051, and by Romanian Ministry of Research, Innovation and Digitalization in the framework of PN-III-P4-ID-PCE-2020-1673 (contract no. IDEI191/2021), PN-III-P1-1.1-TE-2021-1491 (TE71/2022), and Core Program PC2-PN23080202.

Conflict of Interest

The authors declare no conflict of interest.

Data Availability Statement

Research data are not shared.

Keywords

magnetron sputtering, photocurrents, rapid-like furnace annealing, SiGe nanocrystals, stress field and defects, structure and morphologies

Received: November 10, 2023

Revised: May 9, 2024

Published online: June 19, 2024

[1] E. Tuğay, S. İlday, R. Turan, T. G. Finstad, *J. Lumin.* **2014**, *155*, 170.

- [2] C. Pinto, R. J. Kashtiban, A. G. Rolo, M. Buljan, A. Chahboun, U. Bangert, N. P. Barradas, M. Gomes, *Thin Solid Films* **2010**, 518, 2569.
- [3] M. T. Sultan, V.-A. Maraloiu, I. Stavarache, J. T. Gudmundsson, A. Manolescu, V. S. Teodorescu, M. L. Ciurea, H. G. Svavarsson, *Beilstein J. Nanotechnol.* **2019**, 10, 1873.
- [4] M. T. Sultan, J. T. Gudmundsson, A. Manolescu, V. S. Teodorescu, M. L. Ciurea, H. G. Svavarsson, *Appl. Surf. Sci.* **2020**, 511, 145552.
- [5] A.-M. Lepadatu, C. Palade, A. Slav, O. Cojocaru, V.-A. Maraloiu, S. Iftimie, F. Comanescu, A. Dinescu, V. S. Teodorescu, T. Stoica, M. L. Ciurea, *J. Phys. Chem. C* **2020**, 124, 25043.
- [6] F. Vieira, C. Pinto, S. Levichev, A. G. Rolo, A. Chahboun, M. Buljan, N. P. Barradas, E. Alves, S. Bernstorff, O. Conde, M. Gomes, *Microelectron. Eng.* **2011**, 88, 509.
- [7] R. Gupta, W. J. Yoo, Y. Wang, L. Tan, G. S. Samudra, S. Lee, D. S. H. Chan, K. P. Loh, L. K. Bera, N. Balasubramanian, D.-L. Kwong, *Appl. Phys. Lett.* **2004**, 84, 4331.
- [8] H. B. Chew, W. K. Choi, Y. L. Foo, F. Zheng, W. K. Chim, Z. J. Voon, K. C. Seow, E. A. Fitzgerald, D. Lai, *Nanotechnology* **2006**, 17, 1964.
- [9] D. Lehninger, J. Beyer, J. Heitmann, *Phys. Status Solidi A* **2018**, 215, 1701028.
- [10] N. P. Barradas, F. Vieira, A. Parisini, O. Conde, J. Martín-Sánchez, A. G. Rolo, A. Chahboun, M. Gomes, *Nucl. Instrum. Methods Phys. Res., Sect. B* **2014**, 331, 89.
- [11] I. Stavarache, C. Logofatu, M. T. Sultan, A. Manolescu, H. G. Svavarsson, V. S. Teodorescu, M. L. Ciurea, *Sci. Rep.* **2020**, 10, 3252.
- [12] R. Bahariqushchi, R. Raciti, A. E. Kasapoğlu, E. Gür, M. Sezen, E. Kalay, S. Mirabella, A. Aydinli, *Nanotechnology* **2018**, 29, 185704.
- [13] X. Wang, Z. Jian-guo, D. Liu, B. Cheng, W. Ge, J. Yu, Q. Wang, *Phys. Rev. B* **2005**, 72, 195313.
- [14] G. Zatyrb, A. Podhorodecki, J. Misiewicz, J. Cardin, F. Gourbilleau, *Nanoscale Res. Lett.* **2013**, 8, 40.
- [15] N. A. P. Mogaddam, A. S. Alagöz, S. Yerci, R. Turan, S. E. Foss, T. G. Finstad, *J. Appl. Phys.* **2008**, 104, 124309.
- [16] A. K. Sood, J. W. Zeller, R. A. Richwine, Y. R. Puri, H. Efstathiadis, P. Haldar, N. K. Dhar, D. L. Polla, *Advances in Optical Fiber Technology: Fundamental Optical Phenomena and Applications* (Eds: M. Yasin, H. Arof, S. W. Harun) InTech **2015**, <https://doi.org/10.5772/59065>.
- [17] K. Cheng, R. Anthony, U. R. Kortshagen, R. J. Holmes, *Nano Lett.* **2011**, 11, 1952.
- [18] P. Liu, S. Cosentino, S. T. Le, S. Lee, D. C. Paine, A. Zaslavsky, D. Pacifici, S. Mirabella, M. Miritello, I. Crupi, A. Terrasi, *J. Appl. Phys.* **2012**, 112, 083103.
- [19] M. Aouassa, R. M'gaieth, B. Azeza, I. Berbézier, L. Favre, *J. Mater. Sci.: Mater. Electron.* **2021**, 32, 16800.
- [20] A. Dey, D. Das, *J. Phys. Chem. Solids* **2021**, 154, 110055.
- [21] M. Glover, A. Meldrum, *Opt. Mater.* **2005**, 27, 977.
- [22] I. Stavarache, V. S. Teodorescu, P. Prepelita, C. Logofatu, M. L. Ciurea, *Sci. Rep.* **2019**, 9, 10286.
- [23] W. K. Choi, S. Kanakaraju, Z. X. Shen, W. S. Li, *Appl. Surf. Sci.* **1999**, 144–145, 697.
- [24] S. Takeoka, K. Toshiaki, M. Fujii, S. Hayashi, K. Yamamoto, *J. Lumin.* **2000**, 87–89, 350.
- [25] A. I. Yakimov, V. V. Kirienko, V. A. Armbrister, A. B. Dvurechenskii, *Semicond. Sci. Technol.* **2014**, 29, 085011.
- [26] M. L. Ciurea, I. Stavarache, A.-M. Lepadatu, I. Pasuk, V. S. Teodorescu, *Phys. Status Solidi B* **2014**, 251, 1340.
- [27] I. Berbezier, A. Ronda, J.-N. Aqua, L. Favre, T. Frisch, *Nanostructured Semiconductors: From Basic Research to Applications*, Pan Stanford Publishing, New York, NY **2014**.
- [28] M. L. Ciurea, S. Lazanu, I. Stavarache, A.-M. Lepadatu, V. Iancu, M. R. Mitroi, R. R. Nigmatullin, C. M. Baleanu, *J. Appl. Phys.* **2011**, 109, 013717.
- [29] F. Vieira, J. Toudert, A. G. Rolo, A. Parisini, J. P. Leitão, M. R. Correia, N. Franco, E. Alves, A. Chahboun, J. Martín-Sánchez, R. Serna, M. Gomes, *Nanotechnology* **2017**, 28, 345701.
- [30] D. Hiller, S. Goetze, M. Zacharias, *J. Appl. Phys.* **2011**, 109, 054308.
- [31] M. Zschintzsch, C. J. Sahle, J. von Borany, C. Sternemann, A. Mücklich, A. Nyrow, A. Schwamberger, M. Tolan, *Nanotechnology* **2011**, 22, 485303.
- [32] Z. Wan, S. Huang, M. A. Green, G. Conibeer, *Nanoscale Res. Lett.* **2011**, 6, 129.
- [33] M. Draghici, M. Miu, V. Iancu, A. Nassiopoulou, I. Kleps, A. Angelescu, M. L. Ciurea, *Phys. Status Solidi* **2000**, 182, 239.
- [34] V. Ioannou-Sougleridis, A. G. Nassiopoulou, M. L. Ciurea, F. Bassani, F. A. d'Avitaya, *Mater. Sci. Eng., C* **2001**, 15, 45.
- [35] X. Zhang, T. Zhang, M. Wong, Y. Zohar, in *Proc. IEEE The Tenth Annual Inter. Workshop on Micro Electro Mechanical Systems. An Investigation of Micro Structures, Sensors, Actuators, Machines and Robots*, IEEE, Piscataway, NJ **2002**, <https://doi.org/10.1109/memsys.1997.581922>.
- [36] S. Prucnal, J. M. Sun, A. Muecklich, W. Skorupa, *Electrochem. Solid-State Lett.* **2007**, 10, H50.
- [37] D. Y. C. Lie, J.-H. Song, M.-A. Nicolet, N. D. Theodore, *Appl. Phys. Lett.* **1995**, 66, 592.
- [38] M. de Cea, D. Van Orden, J. Fini, M. Wade, R. J. Ram, *APL Photonics* **2021**, 6, 041302.
- [39] Z. G. Tegegne, J. Nanni, C. Viana, G. Tartarini, J.-L. Polleux, *Electron. Lett.* **2019**, 55, 656.
- [40] Z. Pei, C. S. Liang, L. S. Lai, Y. T. Tseng, Y. M. Hsu, P. S. Chen, S. C. Lu, M.-J. Tsai, C. W. Liu, *IEEE Electron Device Lett.* **2003**, 24, 643.
- [41] A. Elfving, G. V. Hansson, W.-X. Ni, *Physica E* **2003**, 16, 528.
- [42] J. Oh, Z. Huang, J. C. Campbell, *Appl. Phys. Lett.* **2004**, 85, 3286.
- [43] K.-S. Lai, J.-C. Huang, K. Y.-J. Hsu, *IEEE Electron Device Lett.* **2007**, 28, 800.
- [44] S. Park, S. Takita, R. Ichikawa, Y. Ishikawa, K. Wada, *ECS Trans.* **2008**, 16, 591.
- [45] J. Aberl, M. Brehm, T. Fromerz, J. Schuster, J. Frigerio, P. Rauter, *Opt. Express* **2019**, 27, 32010.
- [46] Y. Gao, R. Das, Y. Xie, F. Guo, P. Mascher, A. P. Knights, *Opt. Express* **2024**, 32, 227.
- [47] M. T. Sultan, A. Manolescu, J. T. Gudmundsson, K. Torfason, G. A. Nemmes, I. Stavarache, C. Logofatu, V. S. Teodorescu, M. L. Ciurea, H. G. Svavarsson, *Appl. Surf. Sci.* **2019**, 469, 870.
- [48] M. T. Sultan, J. T. Gudmundsson, A. Manolescu, V. S. Teodorescu, M. L. Ciurea, H. G. Svavarsson, *Nanotechnology* **2019**, 30, 365604.
- [49] C. Palade, A.-M. Lepadatu, A. Slav, V. S. Teodorescu, T. Stoica, M. L. Ciurea, D. Ursutiu, C. Samoilă, *Materials* **2021**, 14, 7040.
- [50] T. Tah, C. K. Singh, S. Amirthapandian, K. K. Madapu, A. Sagdeo, S. Ilango, T. Mathews, S. Dash, *Mater. Sci. Semicond. Process.* **2018**, 80, 31.
- [51] V. S. Teodorescu, C. Ghica, A. V. Maraloiu, M. Vlaicu, A. Kuncser, M. L. Ciurea, I. Stavarache, A. M. Lepadatu, N. D. Scarisoreanu, A. Andrei, V. Ion, M. Dinescu, *Beilstein J. Nanotechnol.* **2015**, 6, 893.
- [52] J. Martín-Sánchez, I. Capan, A. Chahboun, C. Pinto, F. Vieira, A. G. Rolo, M. Gomes, *Appl. Surf. Sci.* **2013**, 280, 632.
- [53] E. Tuğay, R. Turan, *J. Nanosci. Nanotechnol.* **2016**, 16, 4052.
- [54] A.-M. Lepadatu, A. Slav, C. Palade, I. Dascalescu, M. Enculescu, S. Iftimie, S. Lazanu, V. S. Teodorescu, M. L. Ciurea, T. Stoica, *Sci. Rep.* **2018**, 8, 4898.
- [55] A. I. Yakimov, A. V. Dvurechenskii, A. I. Nikiforov, Y. Y. Proskuryakov, *J. Appl. Phys.* **2001**, 89, 5676.