Modulation of the absorption coefficient at $1.3\mu\text{m}$ in Ge/SiGe multiple quantum well heterostructures on silicon

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We report modulation of the absorption coefficient at $1.3\mu\text{m}$ in Ge/SiGe multiple quantum well heterostructures on silicon via the quantum-confined Stark effect. Strain engineering was exploited to increase the direct optical band-gap in the Ge quantum wells. We grew 9 nm-thick Ge quantum wells on a relaxed $\text{Si}_{0.22}\text{Ge}_{0.78}$ buffer and a contrast in the absorption coefficient of a factor of greater than 3.2 was achieved in the spectral range 1290–1315 nm. © 2011 Optical Society of America

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Existing silicon Mach–Zehnder modulators that exploit the carrier dispersion effect are typically either large and dissipate considerable amounts of power [1] or require the use of resonant cavities, which have temperature stabilization issues and are very sensitive to fabrication tolerances. The development of silicon-based electroabsorption modulators (EAMs) is desirable for emerging silicon photonics applications, including optical network interconnects and fibre-to-the-home, because such devices can have a small footprint, low power consumption, and good temperature stability. Several optical fiber telecommunications systems exploit the spectral ‘window’ around $1.3\mu\text{m}$, which corresponds to zero dispersion in standard single-mode fibers. In particular, some passive optical network (PON) architectures use $1.3\mu\text{m}$ radiation for upstream signals [2]. Therefore, it is desirable to fabricate optical modulators that can operate at this wavelength.

Ge/SiGe multiple quantum well (MQW) heterostructures can be epitaxially grown on silicon wafers using a relaxed buffer layer (see, e.g., [3]), where the alternating layers should be strain-balanced to the buffer layer so that no net strain accumulates in the MQW stack. Previous studies of the quantum-confined Stark effect (QCSE) in Ge/SiGe MQW structures have reported buffers layers with a Ge fraction of 90% or more [4–7]. Here, we describe absorption spectra for a strain-balanced stack of ten 9 nm thick Ge quantum wells and eleven 7 nm thick $\text{Si}_{0.4}\text{Ge}_{0.6}$ barriers grown on a relaxed $\text{Si}_{0.22}\text{Ge}_{0.78}$ buffer. The large compressive strain in the Ge quantum wells results in an increase in the direct bandgap compared with relaxed Ge, which results in a blue-shift of the absorption edge, and proper choice of the layer widths and compositions allows us to control the absorption edge of the structure [8].

The MQW heterostructures were grown using reduced pressure chemical vapor deposition (RP-CVD) on a relaxed $\text{Si}_{0.22}\text{Ge}_{0.78}$ buffer. The buffer was grown using reverse linear grading (RLG) from a relaxed Ge seed layer,

which was grown on an Si substrate [9]. Further details of the epitaxial growth can be found in [10]. Circular mesa devices of $80\mu\text{m}$ diameter were defined using optical lithography and reactive-ion etching and a Ti/Al metal stack was deposited and sintered at 400°C for 30 minutes to form electrical contacts. A schematic diagram of the cross section of the devices is shown in Fig. 1.

Absorption spectra were inferred from the wavelength-dependent photocurrent, which was measured using a 100 W xenon light source with a 3 nm bandwidth monochromator. The output from the monochromator was modulated using an optical chopper, allowing the use of lock-in detection to improve the sensitivity of the measurement and focussed using a $40\times$ microscope objective lens onto a $50\mu\text{m}$ core diameter optical fiber to bring the light onto the device. Figure 2 shows the photocurrent at a range of applied biases; a significant blue-shift is observed compared with existing structures and the zero-bias spectra shows that the direct absorption edge is at wavelengths shorter than $1.3\mu\text{m}$. Upon

Fig. 1. (Color online) Schematic diagram showing a cross section of the MQW *p-i-n* photodiodes. The MQW layer consists of ten wells and eleven barriers sandwiched between two 100 nm-thick intrinsic $\text{Si}_{0.22}\text{Ge}_{0.78}$ spacer layers in order to achieve a uniform electric field across each quantum well in the stack. The doping levels were $1 \times 10^{19} \text{ cm}^{-3}$ for both the *n* and *p* type layers. The structure was vertically illuminated using an optic fiber.

Fig. 2. (Color online) Measured photocurrent data for the Ge/Si_{0.4}Ge_{0.6} MQW stack at a range of biases applied across the *p-i-n* diode. The light-hole exciton can be seen at approximately 1150 nm in the zero-field curve. This large light-hole/heavy-hole splitting is evidence of the large strain due to the high Si fraction of the virtual substrate. The curves at 8 V and 9 V appear larger than the lower-bias curves, which is attributed to avalanche gain within the device. The dashed curves show these data rescaled according to the simulated ratios of the absorption at the electron/heavy-hole exciton peak.

application of a reverse bias, a significant Stark shift was observed.

The photocurrent measured at 8 V and 9 V was significantly enhanced throughout the wavelength range of the experiment. This is attributed to avalanche gain within the device. The total thickness of the MQW region plus the two intrinsic spacer layers was 370 nm; the maximum applied bias of 9 V corresponds to an electric field of approximately 250 kV/cm, which is significantly above the breakdown bias for bulk Ge. This elevation in the breakdown field can be explained by considering that the active region is a MQW stack, so the hole wavefunctions and the *L*-valley electron wavefunctions are displaced from the band edges by the confinement energy, meaning that we can expect an increase in the ionization energy. Additionally, alloy disorder scattering in SiGe has been shown to suppress impact ionization [11], meaning that we can expect an increase in the breakdown field of a SiGe alloy compared with that of pure Ge. This avalanche gain is not expected to be a significant problem for modulator operation, as the only significant effect will be an increase in the photocurrent when the modulator is biased in the absorbing state. If we assume that the power of the laser is small and so the photocurrent is not significant to the power consumption of an EAM [12], then the avalanche gain is not expected to be a significant issue affecting device performance.

Figure 3 shows simulated data calculated based on the method described in [8]. The large compressive strain in the quantum wells results in a considerable splitting of the light and heavy-hole exciton peaks. In order to reproduce this splitting we have revised our choice of the uniaxial model solid deformation potential used in the bandstructure calculations so that we have $b = 2.86$ eV [13], together with a larger lattice bowing parameter of 0.026 nm [14]. These parameters also provide agreement with the experimental data from [3]. Interdiffusion of the

Fig. 3. (Color online) Simulated absorption spectra for the 9 nm Ge/Si_{0.4}Ge_{0.6} MQW heterostructures.

Ge and SiGe layers results in an increase in the confinement energy of the ground state Γ -valley electron wavefunction, which causes a blue-shift of the direct optical absorption edge. We find good agreement in the position of the absorption edge when we include an interdiffusion length of 2 nm.

In the simulated data plotted in Fig. 3, we have shown the absorption in units of fractional absorption per quantum well for light propagating perpendicular to the plane of the quantum wells. Additionally, the data is shown in terms of an effective absorption coefficient, α , which is defined using

$$\frac{I_L}{I_0} = e^{-\alpha L}, \quad (1)$$

where $1 - I_L/I_0$ is the fraction of light absorbed per period and L is the length of a period (i.e., the well thickness plus the barrier thickness). This provides a useful representation of the data, as it allows us to treat the MQW stack as an effective medium for which α gives the absorption coefficient for TE-polarized light propagating in the plane of the quantum wells.

The large compressive strain in the Ge quantum wells results in an increase in the splitting between the Γ - and *L*-valley conduction band edges, as well as a reduction in the energy of the Δ -valley relative to that of the Γ -valley. As such, we can expect an increase in the density of final states in the *L*-valleys and additionally, we can expect that $\Gamma \rightarrow \Delta$ scattering processes become allowed. Both of these effects will result in a reduced intervalley scattering lifetime compared with heterostructures grown on a Si_{0.1}Ge_{0.9} virtual substrate and an increase in the magnitude of the indirect absorption, which is slightly larger here than has been reported for similar structures grown on virtual substrates with a larger Ge fraction [3,7].

To eliminate the avalanche gain from the derived absorption spectra, the measured photocurrents were scaled according to the simulated ratios of absorption at the electron/heavy-hole exciton peak in order to obtain the results shown by the dashed curves in Fig. 2. The contrast in the absorption ratio between that at 0 V and at a given applied bias is shown in Fig. 4. The ratio between the absorption coefficient at 9 V and that at 0 V is 3.6 at 1300 nm and greater than 3.2 throughout the range 1290–1315 nm.

Fig. 4. (Color online) The absorption coefficient plotted normalized to that at 0 V bias. The contrast in the absorption ratio between 9 V and 0 V exceeds a factor of 3.2 from 1290 nm to 1315 nm.

Reduction of the required electric field while maintaining operation at $1.3\ \mu\text{m}$ can be achieved by a combination of increasing the Si fraction in the substrate and increasing dimensions of the quantum wells. The limit on doing so is the critical thickness of an individual layer in the structure. According to the method described by Matthews and Blakeslee, we can expect a strain-balanced stack of 12 nm thick Ge quantum wells and 25 nm thick barriers to be stable on a $\text{Si}_{0.28}\text{Ge}_{0.72}$ relaxed buffer [8,15,16]. Because of the increased compressive strain in the Ge quantum well layers, such an MQW structure can also be expected to operate at $1.3\ \mu\text{m}$, but with a reduction in the required electric field of a factor of approximately two. Additionally, a reduction in the applied bias can be achieved by decreasing the thickness of the intrinsic space layers.

We have described the observation of the quantum-confined Stark effect in a Ge/SiGe MQW heterostructure epitaxially grown on a Si(100) wafer and demonstrated modulation at $1.3\ \mu\text{m}$. Strain engineering was used to increase the direct optical bandgap in the Ge quantum

wells and a contrast in the absorption coefficient of a factor of more than 3.2 was observed for wavelengths in the range 1290–1315 nm.

References

1. L. Liao, A. Liu, D. Rubin, J. Basak, Y. Chetrit, H. Nguyen, R. Cohen, N. Izhaky, and M. Paniccia, *Electron. Lett.* **43**, 1196 (2007).
2. C. Lam, *Passive Optical Networks Principles and Practice* (Elsevier, 2007).
3. Y.-H. Kuo, Y. K. Lee, Y. Ge, S. Ren, J. E. Roth, T. I. Kamins, D. A. B. Miller, and J. S. Harris, *Nature* **437**, 1334 (2005).
4. Y.-H. Kuo, Y. K. Lee, Y. Ge, S. Ren, J. E. Roth, T. I. Kamins, D. A. B. Miller, and J. S. Harris, *IEEE J. Sel. Top. Quantum Electron.* **12**, 1503 (2006).
5. R. K. Schaevitz, J. E. Roth, S. Ren, O. Fidaner, and D. A. B. Miller, *IEEE J. Sel. Top. Quantum Electron.* **14**, 1082 (2008).
6. C. Lange, N. S. Köster, S. Chatterjee, H. Sigg, D. Chrastina, G. Isella, H. von Känel, M. Schäfer, M. Kira, and S. W. Koch, *Phys. Rev. B* **79**, 201306 (2009).
7. P. Chaisakul, D. Marris-Morini, G. Isella, D. Chrastina, X. L. Roux, E. Gatti, S. Edmond, J. Osmond, E. Cassan, and L. Vivien, *Opt. Lett.* **35**, 2913 (2010).
8. L. Lever, Z. Ikonić, A. Valavanis, J. Cooper, and R. Kelsall, *J. Lightwave Technol.* **28**, 3273 (2010).
9. V. A. Shah, A. Dobbie, M. Myronov, D. J. F. Fulgoni, L. J. Nash, and D. R. Leadley, *Appl. Phys. Lett.* **93**, 192103 (2008).
10. X.-C. Liu, M. Myronov, A. Dobbie, R. J. H. Morris, and D. R. Leadley, *J. Phys. D* **44**, 055102 (2011).
11. K. Yeom, J. M. Hinckley, and J. Singh, *J. Appl. Phys.* **80**, 6773 (1996).
12. M. E. Chin and W. S. C. Chang, *IEEE J. Quantum Electron.* **29**, 2476 (1993).
13. M. Chandrasekhar and F. H. Pollak, *Phys. Rev. B* **15**, 2127 (1977).
14. D. De Salvador, M. Petrovich, M. Berti, F. Romanato, E. Napolitani, A. Drigo, J. Stangl, S. Zerlauth, M. Mühlberger, F. Schäffler, G. Bauer, and P. C. Kelires, *Phys. Rev. B* **61**, 13005 (2000).
15. J. Matthews and A. Blakeslee, *J. Cryst. Growth* **32**, 265 (1976).
16. D. J. Paul, *Semicond. Sci. Technol.* **19**, R75 (2004).