Ge/SiGe Quantum Confined Stark Effect Modulators on Silicon

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1. Introduction

Silicon is the dominant semiconductor for electronics, but there is a growing need to integrate optoelectronics for telecommunications and computer interconnections with Si ICs. Silicon-based optical modulators have recently been successfully demonstrated [1,2]; but because the light modulation mechanisms in silicon are relatively weak [3], long (several mm) devices [1] or sophisticated high-Q resonators [2] have been necessary. The QCSE is routinely used in III-V quantum-well modulators. It possesses a number of attractive properties that enable modulators with only micrometers of path length that can be incorporated into large Si CMOS circuits. Waveguide QCSE devices typically have lengths of only, 100–400 μm . QCSE devices do not require carrier injection, and are typically operated as reverse-biased diodes. The resulting low power dissipation—of the order of 10 mW per channel—allow large arrays of optical interconnects operating at high data rates from silicon chips. Theoretically, the QCSE operates at sub-ps times and devices with 50 GHz modulation bandwidth have been demonstrated [4].

2. Quantum well design

Ge is an indirect band gap material, but it has a local minimum in the zone center which gives it a sharp absorption edge and high absorption coefficient. In order to have good quantum confinement, SiGe barriers are used since Si and Ge have a very high direct band gap difference, 3.4 eV. However, there is a 4% lattice constant mismatch between them and this imposes a critical thickness limitation on the growth. In order to monolithically integrate Ge/SiGe quantum well devices with silicon and also prevent the quantum wells and barriers from lattice-relaxing, a strain balanced structure is designed. A relaxed SiGe layer on silicon is used as a buffer with a silicon concentration nearly the same as the average of that in the MQWs, so the strain energy is annihilated after the growth of every quantum well/barrier pair. This extends the growth of the Ge quantum wells beyond the critical thickness limitation.

Since the Ge well is compressively strained, the heavy-hole and light-hole bands are split and the

heavy-hole will be the topmost valence band. The absorption edge is dominated by the direct band gap transition between electron and heavy-hole, which has a stronger Stark shift due to its heavier effective mass. However, both the strain and quantum confinement will cause the Ge absorption edge to blue shift. Figure 1 shows the resulting bandgap alignment. The band discontinuities of the heavy

Fig. 1 Energy band lineup for strained Ge/SiGe QW

hole, light hole, and electron at the Γ point between our wells and barriers are calculated to be 101 meV, 47 meV, and 400 meV respectively [22–24]. Note that this 400 meV Γ -point band discontinuity provides strong electron quantum confinement in the conduction band. There are, of course, lower-energy conduction band minima (L valleys) in the SiGe barriers. For quantum confinement, however, we expect the relevant bands in the barriers to be those with similar unit cell symmetries, as we expect that it is only to such bands that strong tunneling can exist. Hence, we consider here the bands at the Γ point in the barriers for the calculation of quantum confinement effects.

Because the Γ point in Ge wells is higher than the L valleys in the SiGe barriers, we do however expect that, even though the Γ point electron quantum well is relatively deep, electrons generated in the Ge wells

will be rapidly scattered out into these L valleys where they can be swept out by electric fields.

QW structures were grown in a commercially available, single-wafer, cold-wall, reduced-pressure, chemical vapor deposition (RPCVD) reactor, using hydrogen carrier gas and silane and germane precursor gases. After an in-situ high-temperature cleaning, the layers are grown at 400 °C, with annealing at higher temperatures. Two 250-nm $\text{Si}_{0.1}\text{Ge}_{0.9}$ films doped with $5 \times 10^{18} \text{ cm}^{-3}$ boron atoms are grown sequentially and annealed at 850 °C for 30 min and 700 °C for 5 min, respectively, to reduce dislocations caused by the lattice mismatch. A 100nm undoped $\text{Si}_{0.1}\text{Ge}_{0.9}$ spacer layer is grown, followed by ten pairs of MQWs (10nm Ge well/16nm $\text{Si}_{0.15}\text{Ge}_{0.85}$ barrier) and another 100 nm undoped $\text{Si}_{0.1}\text{Ge}_{0.9}$ spacer layer. Finally, the structure is capped by an arsenic-doped $\text{Si}_{0.1}\text{Ge}_{0.9}$ layer with a doping level of 10^{19} cm^{-3} . All materials, processing equipment and temperatures used are compatible with processes used in fabricating Si CMOS ICs [5].

Photocurrent spectra are taken at room temperature for different diode reverse bias voltages with light incident normal to the surface. We expect both light- and heavy-hole transitions in this configuration. The responsivity (current per unit optical power) inside the device is deduced from the light intensity incident upon the open area of the mesa surface, correcting for surface reflection. The corresponding effective absorption coefficient spectra (Fig. 2) are calculated assuming one electron of

Fig. 2. Measured absorption spectra vs. bias

current for every absorbed photon. Clear quantum confinement is seen, with strong exciton peaks that we assign to electron-to-heavy-hole and electron-to-light-hole transitions. The effective absorption coefficient at the e-hh exciton peak,

calculated on the basis of the total thickness of the wells and the barriers (0.26 μm), is $6,320 \text{ cm}^{-1}$. The half-width at half-maximum of the heavy-hole exciton peak is only 8 meV, and the heavy-hole exciton peak is still clearly resolved, even at $8 \times 10^4 \text{ Vcm}^{-1}$ (3V bias). As the quantum confinement energy originates primarily from the electron, the clarity of this quantum shift shows that the quantum confinement at the Γ point is strong, despite the possibility of scattering to the lower indirect valleys.

With a reverse bias from 0 to 4V, both peaks are red shifted by the QCSE. The e-hh exciton is shifted from 1.408nm (0 V) to 1.456nm (4 V). The maximum effective absorption coefficient change is $2,800 \text{ cm}^{-1}$ at 1,438nm under 3 V bias. This is the first efficient electro-absorption modulation observed in group-IV QW structures, and its performance is comparable to high-quality (direct gap) III-V materials at similar wavelengths.

3. Conclusions

We have demonstrated efficient QCSE in silicon-based structures, using strained Ge MQWs. The behavior of the exciton peaks, the band edge shift and the shift in absorption coefficient are comparable to those observed in III-V materials at similar wavelengths. Our materials and fabrication processes are completely CMOS compatible and suitable for mass production. This approach is therefore very promising for silicon-based electro-absorption modulators operating at high speed, low power, low operating voltage and with small device areas.

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