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Quantum-Confined Stark Shift Due to Piezoelectric Effect in InGaAs/GaAs Quantum Wells Grown on (111) A GaAs

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$In_{0.2}Ga_{0.8}As$/GaAs strained quantum wells (SQWs) were grown on GaAs (111)A just, 1° off and 5° off toward [110]- and [001]-oriented substrates. Dependence of strain relaxation on substrate orientation was studied by photoluminescence (PL) spectroscopy. Samples grown on GaAs (111)A 5° off toward [001]-oriented substrates showed the best optical characteristics and this substrate orientation was chosen for making *p-i-n* diodes. The PL spectrum shows the influence of a built-in electric field due to the piezoelectric effect. The blueshift of PL peaks with applied bias was demonstrated in a *p-i-n* structure. The PL peak corresponding to a 10 nm SQW blueshifted as much as 24 meV with only 1.2 V applied reverse bias.

1.INTRODUCTION

InGaAs/GaAs strained quantum wells (SQW) grown on substrates with orientation different from (100) are attracting attention because the presence of built-in electric fields originated in the piezoelectric effect. These electric fields redshift the optical transition energies accordingly to the quantum confined Stark effect (QCSE). If an external electric field opposite to the built-in electric field is applied, a blueshift of the optical transition energies can be observed[1]. This blueshift has been observed by photocurrent spectroscopy and electroreflectance spectroscopy in SQWs embedded in the insulating region of *p-i-n* diodes[2,3]. All but a few of these structures have been grown on (111)B-oriented substrates because it gives the largest built-in electric field. (111)A-oriented substrates give the same built-in electric field but crystal growth is by far more difficult.[4] The electric field in SQWs grown on gallium terminated (A) substrates has the opposite direction of SQWs grown on arsenic terminated (B) substrates. This property can be desirable when designing some devices.

In this paper, we report the first observation of blueshift of the optical transition energies by photoluminescence spectroscopy in $In_{0.2}Ga_{0.8}As$/GaAs SQWs grown on (111)A GaAs.

2. EXPERIMENTAL

$In_{0.2}Ga_{0.8}As$/GaAs SQWs were grown on undoped GaAs (111)A just, 1° off and 5° off toward [110]- and [001]-oriented substrates. Samples grown on GaAs (111)A 5° off toward [001]-oriented substrates showed the best optical characteristics and this substrate orientation was chosen for growing samples on p-doped substrates and making *p-i-n* diodes. The native oxide layer was removed using H_2SO_4:H_2O and the wafers were rinsed in deionized water for 5 min before etching for 80 sec with NH_4OH:H_2O_2:H_2O (2:1:95) at 25°C. After etching, the wafers were rinsed in deionized water for 5 min and dried with nitrogen. All samples were mounted on molybdenum holders without indium and loaded into a Varian Modular Gen II MBE chamber just after etching. The thermal cleaning temperature was 700°C. The As_4 beam equivalent pressure was 2.2×10^{-5} Torr and the As_4/Ga flux ratio was 6.5. This high As_4/Ga flux ratio has two purposes. Firstly, it is required to obtain smooth surfaces on (111)A oriented samples. Secondly, it assures that the amphoterous dopant silicon is incorporated as a donor. All GaAs layers were grown at 620°C. Substrate temperature was decreased during the last few minutes of GaAs growth before starting the growth of InGaAs. The InGaAs layers were grown at 560°C. The *p-i-n* diode structure consisted of: 120 nm Be-doped ($2 \times 10^{18} cm^{-3}$) GaAs buffer layer; 80 nm non-intentionally doped (NID) GaAs spacer; 10 nm NID $In_{0.2}Ga_{0.8}As$ SQW; 80 nm NID GaAs spacer; 5 nm NID $In_{0.2}Ga_{0.8}As$ SQW; 80 nm NID GaAs spacer; 2.5 NID $In_{0.2}Ga_{0.8}As$ SQW; 80 nm NID GaAs spacer; 120 nm Si-doped ($2 \times 10^{18} cm^{-3}$) cap layer. Undoped samples with the same structure were also grown. The pre-treatment and growth techniques developed in our laboratory allowed us to succeed in growing high optical quality SQWs[4].

Photoluminescence (PL) measurements were performed at 12 K using an Ar^+ laser (488 nm wavelength) or a Ti-sapphire laser (844 nm wavelength), a monochromator and an InGaAs photomultiplier detector. The surface morphology of the samples was observed with a scanning electron microscope (SEM).

3. RESULTS AND DISCUSSION

The InGaAs layers are biaxially compressed in the plane of the SQWs. This strain increases the band-gap and shifts the band alignment at heterointerfaces. The strained bandgap was calculated after Pollak[5]; band alignment was calculated using the solid model of Van de Walle[6]. The strain generated electric field was calculated from bulk piezoelectric theory. For x=0.20 on the (111) direction, the theoretical electric field is E=280 kV/cm. The energy levels and wave functions were calculated by numerically solving the Schrödinger equation using the transfer matrix method.[7]

Figure 1 shows the potential profile of the electrons and holes' bands for the 10 nm SQW. An effective mass equal to $0.9m_0$ was used for the heavy holes' band. The band edge tilt due to the built-in electric field is so strong that both electrons and holes are confined in triangular quantum wells. A smaller value of electric field is used in Fig. 1 to fit experimental data, as is explained below. The electrons' confinement in the wells is improved by the tilted GaAs barrier.

Figure 2 shows the full-width at half-maximum (FWHM) and PL peak energy dependence on substrate orientation. Samples grown on GaAs (111)A just and 1° off-oriented substrates show red-shift of the PL peak and large FWHM. The redshift is originated in the bandgap reduction due to strain relaxation. The large FWHM indicates that strain relaxation is not spatially homogeneous.

The cross-section of the sample grown on GaAs (111)A just-oriented substrate was observed by TEM. Interface roughness was estimated to be less than three monolayers (1 nm approx.). The thickness of SQWs agree with the design value. We observed a dislocation density of about $1x10^{10}$ cm^{-2}. This dislocation density would produce 20% of strain relaxation, in agreement with the observed redshift of the PL peak for this sample.

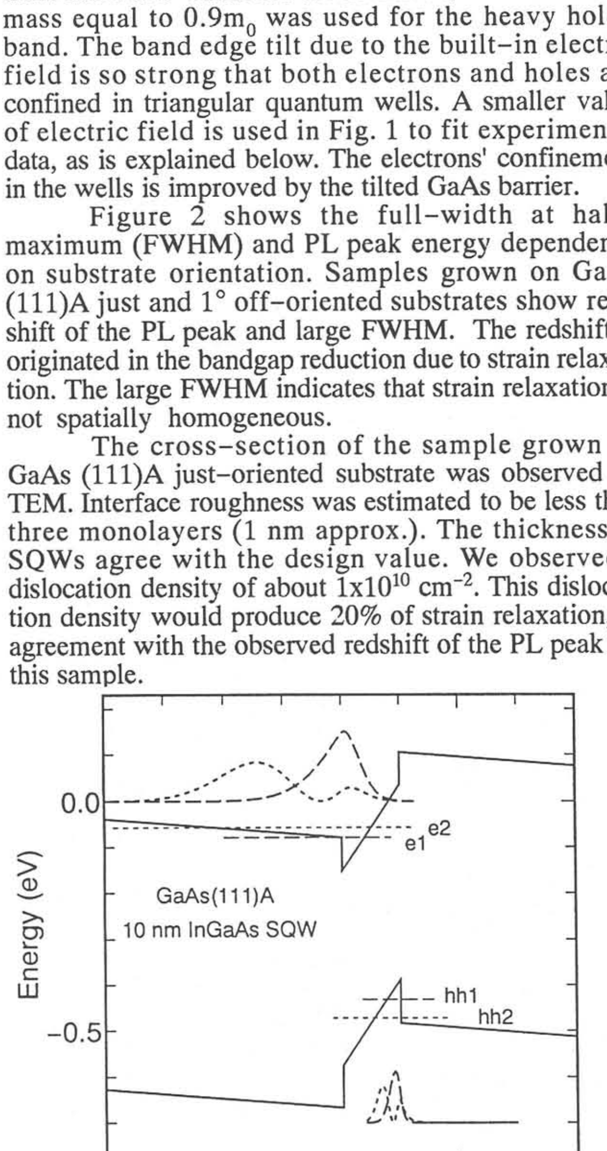

Figure 1. Potential profile for the 10 nm SQW. The two lower energy levels for electrons and holes and their wave functions are shown. The built-in electric field in the barrier helps to confine the electrons in the well. The bandgap is not at scale with the potential profile of the band edges.

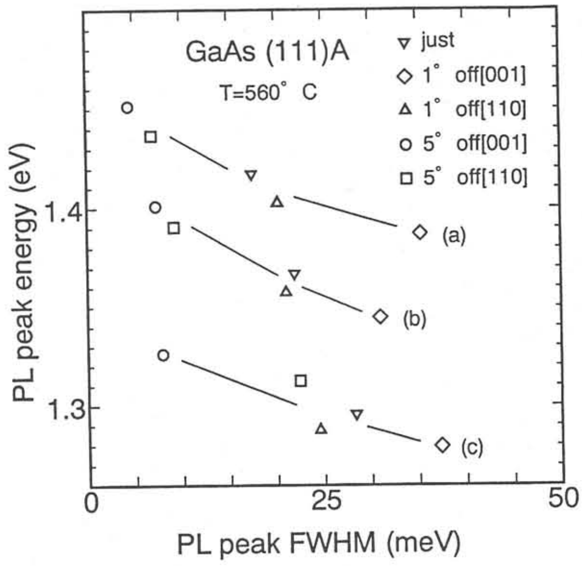

Figure 2. Photoluminescence peak energy and FWHM for various substrate orientations. Each sample has three SQW with differents thickness: (a) 2.5 nm. (b) 5 nm. (c) 10 nm.

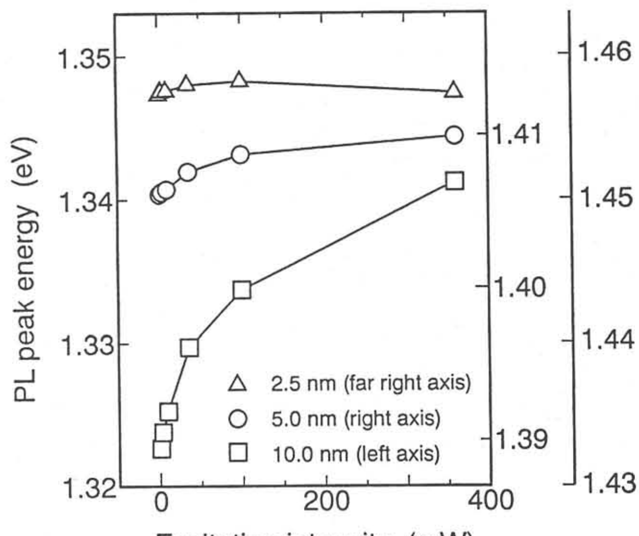

Figure 3. Photoluminescence peak energy dependence on excitation intensity for SQWs grown on GaAs (111)A 5° off toward [001]-oriented substrate.

The sample grown on GaAs (111)A 5° off toward [001] showed the best optical characteristics. The FWHMs under low intensity excitation for this sample are 4.5 meV, 7.3 meV and 7.9 meV for the 2.5 nm, 5.0 nm and 10.0 nm SQWs, respectively. These values compare well with the FWHMs of samples grown side-by-side on GaAs (100) substrates (not shown in the Fig. 2): 4.4 meV, 4.4 meV and 5.4 meV, respectively. The increase of FWHM with well thickness in the samples grown on (111)A 5° off substrates would indicate that even in these samples some strain relaxation occurs.

The experimental demonstration of the piezoelectric fields in an undoped sample is shown in Fig. 3. The PL peak energy blueshifts with increasing excitation intensity since the built-in electric field is screened by photogenerated carriers.

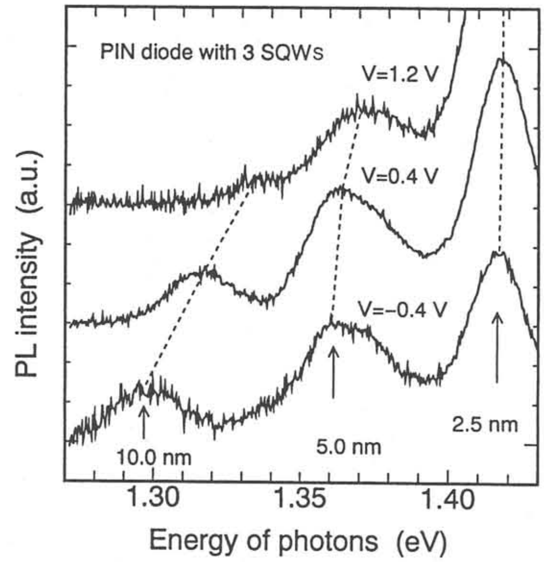

Figure 4. Photoluminescence spectrum of a $p-i-n$ diode. The PL peak shifts toward higher energy as reverse bias is increased.

The as-grown doped samples were processed into 0.4 mm square mesas. The top contact had a 0.2 mm window for optical access. Figure 4 shows the PL spectra of the diodes with applied bias. Low excitation intensity (less than 1 W/cm^2) was used to avoid screening the electric field with photogenerated carriers. The PL peak corresponding to the 10 nm SQW blueshifted as much as 24 meV with only 1.2 V applied reverse bias. The larger FWHM of the PL peaks, as compared to the undoped samples, can be originated in the annular contact that produces an inhomogeneous electric field in the SQWs.

Figure 5 shows the PL peak dependence on applied electric field. The dotted lines are calculated using the piezoelectric field as fitting parameter. A piezoelectric constant 30% smaller than the calculated value was used to obtain agreement between theory and experiment. The discrepancy in the piezoelectric field can be related to partial strain relaxation and/or to the presence of surface charge at the heterointerfaces. This charge would partially screen the piezoelectric field. The blueshift is nearly linear with applied reverse bias because more than 6 V are required to reach the flat band condition in the SQW.

4. CONCLUSION

In conclusion, we grew InGaAs/GaAs SQWs on GaAs (111)A just, 1° off and 5° off toward [110]- and [001]-oriented substrates. Dependence of strain relaxation on substrate orientation was studied by PL spectroscopy. Samples grown on GaAs (111)A 5° off toward [001]-oriented substrates showed the best optical characteristics and this substrate orientation was chosen for making $p-i-n$ diodes. The PL spectrum shows the influence of a built-in electric field due to the piezoelectric effect. The blueshift of PL peaks with applied reverse bias was demonstrated in a $p-i-n$ structure. The large blueshift and good optical properties observed open the path to optical modulators for integration with other devices on (111)A GaAs that we are currently investigating.

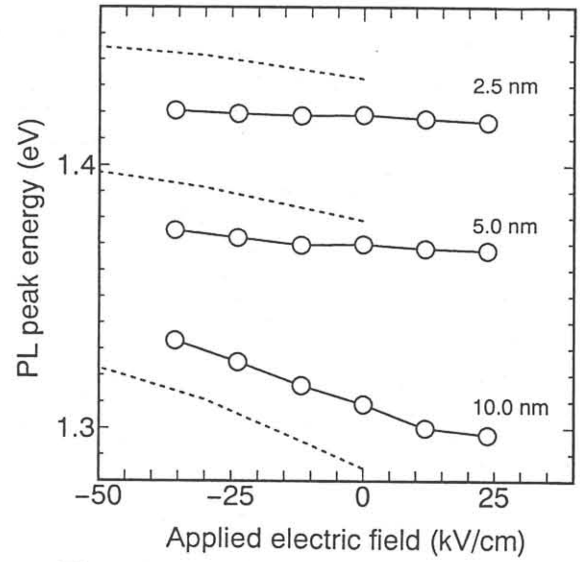

Figure 5. Photoluminescence peak energy dependence on applied electric field. The dotted line is the calculated value for a piezoelectric field of 196 kV/cm.

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