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Surface recombination velocity in GaAs and $In_{0.15}Ga_{0.85}As$ thin films

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The authors have made photoluminescence intensity (PLI) measurements on thin GaAs and InGaAs films to compare the surface recombination velocity at interfaces of the materials with AlAs and native oxide. An analytical expression for the internal quantum efficiency of the PLI method for thin semiconducting films is derived. This expression is applied to measurements on specially designed multilayer structures based on GaAs and InGaAs thin films. The results show that the native oxide on an $In_{0.15}Ga_{0.85}As$ film has a one order of magnitude lower surface recombination velocity than the native oxide on a GaAs film. © *2007 American Institute of Physics*. [DOI: 10.1063/1.2717533]

Si-device scaling will become increasingly difficult in the years to come. Especially the introduction of high-*k* dielectrics as a replacement for the SiO_2 gate oxide poses numerous problems. On the other hand, with the replacement of SiO_2 as the passivating layer of the channel surface, one of the most important advantages of Si over other semiconductors vanishes. Therefore, in recent years, the fabrication of metal-oxide-semiconductor field-effect transistors (MOSFETs) on alternative channel materials with high carrier mobility has gained increased interest, and much progress was made in this area. Notably Ge-,[1] GaAs-,[2–6] and InGaAs (Refs. 7 and 8)-based MOS devices show promising results. Nevertheless, especially for GaAs-based devices, which are very popular in the research community, serious surface passivation problems still persist.

In the following we will measure the surface recombination velocity of thin GaAs and $In_{0.15}Ga_{0.85}As$ films with the photoluminescence intensity (PLI) method[9] and demonstrate that InGaAs is a much more forgiving material, with a surface recombination velocity of the semiconductor-native oxide interface that is one order of magnitude lower than for the GaAs-native oxide interface. Whereas this observation has already been reported for thin InGaAs films with 53% of In,[10] where the surface recombination velocity was measured with a contactless laser-pumped carrier lifetime bridge, we extend this observation here to InGaAs films with a much lower In concentration of 15% by applying the PLI method.

The experimental setup of the PLI method consists of a HeNe laser with an intensity of 15 mW, which illuminates an area of 1 mm^2. Lenses in front of the sample capture the emitted photoluminescence (PL) light and focus it into a grating monochromator. The spectral response of the PL light is read out with a Si detector through a lock-in amplifier. The sample is mounted inside a cryostat, which allows for measurements at 77 K.

Our samples are thin multilayer films grown using metal organic vapor phase epitaxy. The layer structure of the different samples is shown in Fig. 1. The first sample consists of a Ge substrate on which a 100 nm thick buffer layer of GaAs was grown, followed by the deposition of a 20 nm thick AlAs layer, a 40 nm thick GaAs layer, and finally a 20 nm thick AlAs top layer. This structure has the particular property that only the PL light from the 40 nm thick GaAs quantum well will be detected. Indeed, the thin 100 nm GaAs buffer layer on the Ge substrate does not show any PL light at the band gap energy of GaAs (1.5 eV), as all charge carriers generated in the GaAs will quickly diffuse into the Ge, where they will be lost without emitting a photon.[11] The second sample is similar to the first sample, except that the top AlAs layer has been omitted. After removal of the sample from the reactor a thin native GaAs oxide grows, allowing the study of the GaAs-native oxide interface. The third sample consists of a semi-insulating GaAs substrate on which a 100 nm thick GaAs buffer layer was grown, followed by a 20 nm thick AlAs layer, a 40 nm thick $In_{0.15}Ga_{0.85}As$ layer, and a 20 nm AlAs top layer. For this sample the PL response from the thin InGaAs quantum well can be easily separated from the GaAs substrate and buffer layer because of the different spectral responses of the PL light. Whereas the InGaAs has a band gap at 77 K equal to 1.3 eV, the GaAs has a band gap of 1.5 eV. The small peak widths of the PL spectra at 77 K enable the secure separation of these lines, allowing the study of the PL response of the thin InGaAs quantum well without interference from the underlying layers. The fourth sample is similar to the third sample, except that the top AlAs layer was omitted. After removal of the sample from the growth chamber a thin native oxide layer forms, which enables the study of the $In_{0.15}Ga_{0.85}As$-native oxide interface. All four samples were grown with the top quantum well layer being approximately 5×10^{16} cm^{-3} *n*-type. A second set of four samples consisted of identical structures with approximately 5×10^{16} cm^{-3} *p*-type doped layers.

We will now derive the expression for the internal quantum efficiency of a thin semiconducting layer excited with a constant laser illumination. We start from the one-dimensional continuity equations for charge carriers inside a semiconductor.[12] The one-dimensional approach is justified by the large extension of the laser spot. This allows us to neglect all variations in the lateral directions,

$$\frac{\partial n}{\partial t} = n\mu_n\frac{\partial E}{\partial x} + \mu_n E\frac{\partial n}{\partial x} + D_n\frac{\partial^2 n}{\partial x^2} + G - R_{\mathrm{rad}} - R_{\mathrm{bulk}} - R_{S1} - R_{S2}, \quad (1)$$

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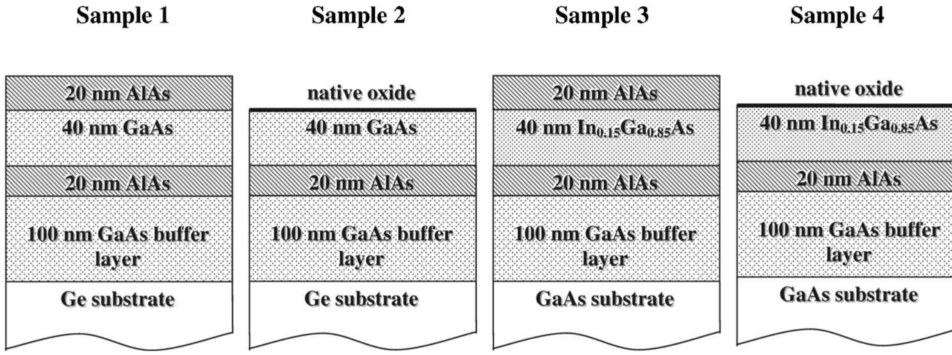

FIG. 1. Schematical representation of the four different samples.

$$\frac{\partial p}{\partial t} = -p\mu_p\frac{\partial E}{\partial x} - \mu_p E\frac{\partial n}{\partial x} + D_p\frac{\partial^2 p}{\partial x^2} + G - R_{\text{rad}} - R_{\text{bulk}} - R_{S1} - R_{S2}. \quad (2)$$

Here, n and p are the electron and hole densities, E the electric field, μ_n and μ_p the electron and hole mobilities, D_n and D_p the electron and hole diffusion constants, G the carrier generation rate, R_{rad} the bulk radiative recombination rate, R_{bulk} the nonradiative bulk recombination rate, R_{S1} the surface recombination rate at the front interface, and R_{S2} the surface recombination rate at the rear interface. Several approximations allow the simplification of the general equations: The thin film approximation allows us to neglect all spatial derivatives of the carrier concentrations and the electric field and consider these quantities as homogeneous inside the thin layer. This approximation holds for films for which the diffusion length λ of the charge carriers is much larger than the film thickness: $\sqrt{D\tau} \gg t$, where D is the carrier diffusion constant, τ is the carrier lifetime, and t is the film thickness. This condition is largely established in our samples as we have diffusion lengths of the order of microns and a film thickness of 40 nm. Furthermore, because the laser illumination is continuous in our measurement setup, we can neglect the time derivative and use the steady-state equations. Finally, we neglect the nonradiative bulk recombination rate, which is justified considering our high quality dislocation-free films and the small thickness of the films, which makes losses at the interfaces much more important than bulk losses. This leaves us with one equation, which governs the equilibrium between carrier generation and recombination in our thin films,

$$G - R_{\text{rad}} - R_{S1} - R_{S2} = 0. \quad (3)$$

The radiative recombination rate is given by

$$R_{\text{rad}} = B(np - n_i^2), \quad (4)$$

where B is the radiative recombination coefficient, equal to $1.3 \times 10^{-10}\ \text{cm}^3/\text{s}$ in GaAs,[13] and n_i is the intrinsic carrier concentration, which can be safely neglected with respect to np if there is external illumination. The surface recombination rate R_S is approximated by the Shockley-Read-Hall recombination rate for a single level recombination center at energy E_t,[14]

$$R_S = \frac{np - n_i^2}{(1/\sigma_p v_{\text{th}} N_t)[n + n_i \exp((E_t - E_i)/kT)] + (1/\sigma_n v_{\text{th}} N_t)[p + n_i \exp(-(E_t - E_i)/kT)]}, \quad (5)$$

where σ_n and σ_p are the electron and hole capture cross sections of the recombination center, v_{th} the carrier thermal velocity, N_t the trap density, and E_i the intrinsic Fermi energy. In our thin layers the trap density is equal to $N_t = D_{\text{it}}/t$, where D_{it} is the interface state density of the surface and t is the film thickness. In a low illumination regime, when injected carriers are much fewer in number than the majority carriers, and with trap levels near the midgap energy, which is a very reasonable approximation for GaAs,[15] expression (5) can be simplified as

$$R_S = \frac{pS_p}{t} \quad \text{for } n\text{-type material}, \quad (6)$$

where $S_p = \sigma_p v_{\text{th}} D_{\text{it}}$ is the hole surface recombination velocity at the interface. Defining the internal quantum efficiency (IQE) of the PLI method as the ratio between the radiative recombination rate R_{rad} and carrier generation rate G, we can write for n-type material by combining Eqs. (3), (4), and (6),

$$\text{IQE} = \frac{Bn}{Bn + (S_{p1} + S_{p2})/t}, \quad (7)$$

where S_{p1} is the hole surface recombination velocity at the front interface and S_{p2} is the hole surface recombination velocity at the rear interface. For sufficiently thin films the $\text{IQE} \ll 1$, and we can neglect the radiative recombination with respect to the surface recombination. We then finally get the expression for the IQE in n-type material,

$$\text{IQE} = \frac{Bnt}{S_{p1} + S_{p2}}. \quad (8)$$

Similarly, for p-type material we get the expression

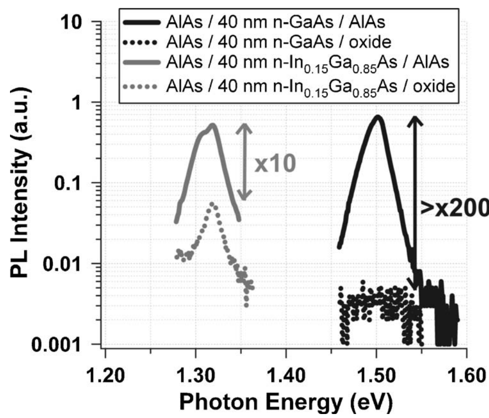

FIG. 2. Photoluminescence response of the four n-type samples.

$$\mathrm{IQE} = \frac{Bpt}{S_{n1} + S_{n2}}. \tag{9}$$

The internal quantum efficiency of the PLI method is therefore, in the approximation of very thin films, steady-state and low illumination conditions, inversely proportional to the sum of the minority carrier surface recombination velocities at the front and rear interfaces. A variation in the surface recombination velocity of the minority carriers is therefore directly translated into the intensity of the PL response.

Figure 2 shows the experimental PL responses from the four n-type samples, whereas Fig. 3 shows the PL spectra for the four p-type samples. The difference in intensity between the oxidized layers and the AlAs-passivated layers can be clearly identified. It is clear that the intensity ratio between the oxidized and passivated films is much smaller for the $In_{0.15}Ga_{0.85}As$ films as compared to the GaAs films. For the n-type layers, the reduction of the PL intensity measured on the oxidized layers as compared to the AlAs-passivated films is at least a factor of 20 more for the GaAs films, as compared to the InGaAs films. Note that the AlAs terminated films have approximately the same PL intensity. This implies, according to Eq. (8) with all other quantities being equal, a more than 20 times lower hole surface recombination velocity S_p of the $In_{0.15}Ga_{0.85}As$-native oxide interface as compared to the GaAs-native oxide interface. Similarly, from the intensity ratios of the p-type samples we deduce a 7.5 times lower electron surface recombination velocity S_n at the $In_{0.15}Ga_{0.85}As$-native oxide interface as compared to the GaAs-native oxide interface. The five times higher total PL intensities for the p-type InGaAs samples is due to a slightly higher doping concentration in these films ($\sim 2 \times 10^{17}$ cm^{-3}). This higher background concentration is translated into a higher total IQE, according to Eq. (9). The differences in surface recombination velocity for both electrons and holes between the GaAs and the InGaAs oxide passivated samples are much larger than what can be expected from the differences in the band gap alone (1.5 vs 1.3 eV). This indicates that even these relatively low In concentrations (15%) cause significant changes on the semiconductor-oxide interface that alter the electronic defect structure and either pushes the defect energy levels out of the band gap or reduces the overall defect density. Although the exact origin of this effect is not yet understood, it might be related to different oxide species forming on the GaAs and InGaAs surfaces.[16,17]

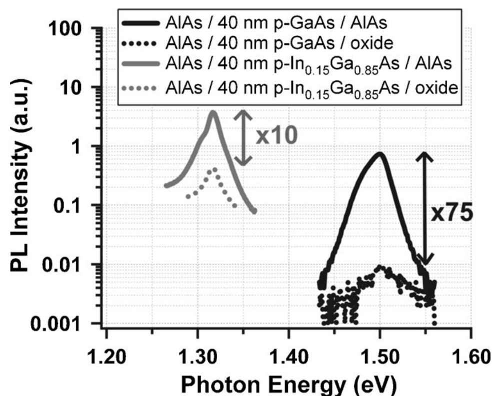

FIG. 3. Photoluminescence response of the four p-type samples.

As a conclusion the results presented in this work suggest that for MOSFET applications InGaAs may be a more forgiving material, for which it is easier to unpin the Fermi level and produce inversion-mode devices.

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