Low surface recombination velocity and contact resistance using p^+/p carbon-doped GaAs structures

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(Received 22 January 1990; accepted for publication 2 April 1990)

A reduction of the GaAs surface recombination velocity due to a heavily carbon-doped GaAs overlayer is reported. Metalorganic molecular beam epitaxy using trimethylgallium, triethylgallium, and elemental arsenic sources has been used to grow an epitaxial structure consisting of 1000 nm of $p = 1\times10^{17}$ cm^{-3} capped with 10 nm of $p = 1\times10^{20}$ cm^{-3} GaAs. Time-resolved photoluminescence (PL) and PL excitation spectroscopy showed this p^+/p structure to have a 3.2 ns carrier lifetime and strong band-edge PL emission, whose intensity was nearly constant over an excitation photon energy range of 1.5–3.3 eV. The same wafer with the p^+ cap etched off exhibited a much shorter carrier lifetime and PL intensity that decreased exponentially with increasing photon energy, which is indicative of carrier losses to surface recombination. The specific contact resistivity of nonalloyed ohmic contacts to these heavily doped layers was observed to be in the mid 10^{-7} Ω cm^2 range, independent of measurement temperature from 77 to 340 K, suggesting a tunneling contact due to the narrow surface depletion layer.

Surfaces of semiconductor materials with native oxides typically exhibit a large density of surface state which tend to fix the location of the surface Fermi level at some characteristic energy.[1] For GaAs this energy is approximately midgap, i.e., about 0.7 eV from a band edge. Such high surface state densities have hampered the development of ideal photonic and electronic devices made of GaAs and related materials such as GaAlAs, GaInAs, etc.[2] In particular, midgap Fermi level pinning leads to large surface recombination velocities (SRVs), high values of metal/semiconductor contact resistance, scaling problems for GaAs metal-semiconductor field-effect transistors, circuit passivation problems, and no practical metal-oxide-semiconductor field-effect transistors to date. The introduction of lattice-matched heterostructures such as GaAlAs on GaAs has provided a limited solution for many photonic and electronic device applications at the expense of increased complexity of the epitaxial and device fabrication processes. Alternative methods of dealing with the GaAs surface Fermi level pinning problem such as photochemically generated oxides[3] and reactive chemical coatings[4] have thus far resulted in only temporary passivating effects. A simple and stable surface passivation technique for GaAs remains to be developed.

In this letter, we explore the use of a thin, very highly doped p-type layer placed at the GaAs surface to simultaneously reduce the GaAs SRV and metal contact resistance. This idea was previously employed to improve the "blue" response of GaAlAs/GaAs solar cells.[5] In the present work, we make use of the observation that GaAs, grown by the metalorganic molecular beam epitaxy method (MOMBE), can contain extremely high levels of an electrically active carbon dopant when trimethylgallium (TMG) is used as a group III precursor.[6-9] In addition, the diffusion coefficient for carbon has been estimated to be at least two orders of magnitude smaller than for Be.[8,10] This property of the carbon dopant makes possible the growth of thin, heavily carbon-doped layers of GaAs without severe broadening of the carbon doping profile.

The carbon-doped, p-type GaAs structures used in this study were grown on semi-insulating (001)GaAs substrates in a Varian Gen II Modular Gas Source MBE system using TMG and triethylgallium (TEG) as group III precursors and solid As_4 from a conventional effusion cell as a group V source. The metalorganic precursors are introduced through a low-pressure injector maintained at 80 °C and are transported by palladium-purified hydrogen carrier gas metered with mass flow controllers. The structure consists of the following layers in order of growth with the group III precursor indicated in parentheses: (1) 500 nm GaAs buffer layer (TEG), (2) 50 nm GaAs barrier layer (TMG), (3) 1000 nm GaAs active layer (TEG), and (4) 10 nm GaAs cap (TMG). The growth rates of the TEG- and TMG-based layers were both calibrated to be 1 μm/h with reflection high-energy electron diffraction (RHEED) oscillation measurements taken immediately prior to growth of the structure. The substrate temperature during growth was 600 °C as measured with an infrared pyrometer.

The hole concentration in layers (2) and (4) are expected to be 1×10^{20} cm^{-3} on the basis of Hall effect and secondary-ion mass spectrometry (SIMS) measurements on calibration layers grown with TMG, while the hole concentrations in layers (1) and (3) are 1×10^{17} cm^{-3} as determined by capacitance-voltage (C-V) profiling on the actual epitaxial layer. The function of layers (1) and (2) is to isolate the minority carriers (electrons) generated by optical pumping from any radiative and nonradiative recombination centers that might exist at the substrate-epilayer interface. Only the luminescence properties of the upper interface of the active layer (3) are pertinent to this study. To evaluate the contribution of the cap layer (4) to the surface passivation of the active layer (3), we have compared the optical properties of the structure with the cap layer (4) intact (p^+/p structure)

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to those with it etched off (p structure). The minority-carrier recombination dynamics were characterized using photoluminescence excitation (PLE) spectroscopy and time-resolved photoluminescence (TRPL) spectroscopy. The metal contact properties were determined using the transmission line method (TLM)[11] and C-V profiling.

For the PLE measurement, light of photon energy greater than the GaAs band gap from a tunable monochromatic excitation source (monochromator and xenon arc lamp) was chopped, passed through a beam splitter, and focussed onto the sample at normal incidence. The PL spectrum was filtered by a second monochromator (at the GaAs band-gap energy) and detected by an S-1 photomultiplier connected to a lock-in amplifier. The PLE signal was normalized to constant absorbed photon flux by detecting a portion of the incident light from the beam splitter. The absorbed photon flux was determined from this incident power by multiplying the incident power by $[1 - R(E)]/E$, where E is the photon energy and $R(E)$ is the photon energy-dependent, normal incidence reflectivity of the GaAs surface.[12] The normalized PLE signal was then measured continuously by varying the wavelength of the excitation monochromator. The PLE response normalized in this manner is proportional to the conversion efficiency of photon-excited electron-hole pairs to band-edge PL as a function of the depth at which the pairs are generated.

The normalized PLE response for the p^+/p and p-GaAs structures is shown in Fig. 1. It is clearly seen that for the p^+/p structure, the PLE signal is strong and only weakly decreasing with photon energy. It has been previously shown that this type of response is indicative of relatively low SRVs,[13,14] i.e., $< 10^4$ cm/s. This behavior is expected from a nearly flatband p-type GaAs sample with a surface-state density of $< 10^{12}\ \mathrm{cm}^{-2}\,\mathrm{eV}^{-1}$. This sample behaves similarly to a p-GaAlAs/p-GaAs sample except that there is no high-energy cutoff due to absorption by the GaAlAs layer. On the other hand, the response for the p structure shows a much weaker normalized PLE signal that decreases exponentially with photon energy due to nonradiative surface recombination losses that increase with decreasing depth of carrier generation. This type of response is therefore indicative of relatively large SRVs, i.e., $> 10^6$ cm/s and is typical of unpassivated p-GaAs with a native oxide.[15]

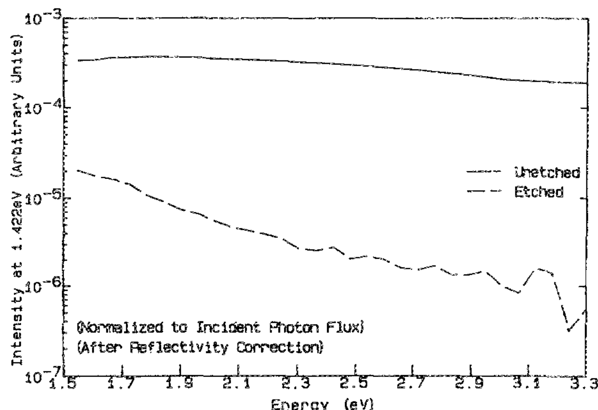

FIG. 1. Comparison of photoluminescence excitation spectra for p^+/p (unetched) and p (etched) GaAs structures. The spectra have been normalized to constant incident photon flux as the incident photon energy is varied.

The effect of the p^+-GaAs cap on minority-carrier recombination dynamics has been further explored with measurements of TRPL on the same epitaxial structure. The TRPL was measured at the GaAs band edge using time-resolved photon counting.[16] Two different excitation sources were employed: a mode-locked Nd:YAG laser (80 ps pulses at 532 nm) and a synchronously pumped dye laser (10 ps pulses at 656 nm). A Hammamatsu R2658 photomultiplier with an InGaAs photocathode was used to detect the transient PL signal. The minimum decay time that can be measured using this system without deconvolution techniques is about 0.5 ns. Injected carrier densities were always kept below $10^{15}\ \mathrm{cm}^{-3}$ to ensure that the observed PL decay times were independent of laser power.

The TRPL response of the p^+/p and p structures are shown in Fig. 2. For the p^+/p structure, the PL transient is a simple single-mode exponential decay with 3.2 ns effective lifetime, which is invariant with respect to the wavelength of the excitation pulses. This observation indicates that surface recombination does not completely dominate the decay of excess minority carriers. The decay time, which is determined by the collective effects of bulk nonradiative, bulk radiative, and interfacial recombination, is given by[17]

$$\frac{1}{\tau} = \frac{1}{\tau_{\mathrm{nr}}} + \frac{1}{\phi\tau_r} + \frac{2S_i}{d}, \quad (1)$$

where τ_{nr} and τ_r are the bulk nonradiative and radiative lifetimes, respectively, ϕ accounts for photon recycling, d is the thickness of the active layer, and S_i is the interfacial recombination velocity at the upper and lower interfaces of the active layer (assumed equal for simplicity). An upper limit for S_i can be estimated for the p^+/p structure by assuming that the interfacial recombination term dominates Eq. (1). For a decay time of 3.2 ns, S_i must be less than 1.5×10^4 cm/s, which agrees with our earlier qualitative expectation based on the PLE spectrum in Fig. 1. Of course, if other bulk nonradiative mechanisms contribute to the minority-carrier decay, then the SRV would be even lower than this value. In this experiment, we have not varied the thickness of the active layer, nor have we attempted to evaluate the effect of photon recycling on the observed external quantum efficiency, and hence we are unable to place a lower limit on the value of the SRV for the p^+/p sample. The PL decay of the p-GaAs structure in Fig. 2 is much more rapid than that of the p^+/p sample. In addition, the decay is nonexpon-

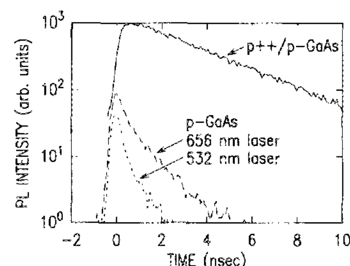

FIG. 2. Photoluminescence decay curves for the p^+/p and p-GaAs structures. The decay curve for the p^+/p GaAs structure is identical for excitation by 532 and 656 nm laser pulses, with an effective lifetime of 3.2 ns. The decay curves for the p-GaAs structure are nonexponential.

ential and occurs more rapidly as the exciting photon energy increases, indicating the dominance of surface recombination in the minority-carrier dynamics.

The reduction in SRV with the p^+/p structure is attributable to several details of the internal electric field distribution near the GaAs surface. To illustrate these points, the band diagram for the p^+/p structure is illustrated in Fig. 3 with surface Fermi level pinning assumed at 0.6 eV above the valence-band maximum. Electrons generated in the p layer deeper than 100 Å will encounter a potential barrier of about 180 meV (ignoring any band-gap narrowing in the p^+ layer) that will effectively confine them away from the surface. Similarly, minority carriers generated in the undepleted portion of the p^+ layer, from 40–100 Å in Fig. 3, that diffuse to the p^+/p interface will experience a drift field that accelerates them away from the surface. Minority carriers generated within about 40 Å of the surface are likely to be driven into surface states due to the reverse band bending associated with surface Fermi level pinning. Without the p^+-GaAs layer, a bare p-GaAs layer doped at 10^{17} cm^{-3} would lose minority carriers generated within a surface depletion region of about 1000 Å, and no potential barrier would exist to prevent electrons generated in the neutral region of the layer from diffusing into the surface field region. A final consideration is the possibility that the amount of surface band bending in the p^+ layer may actually be less than 0.6 eV due to saturation of the surface states at this extreme doping level. In this regard, it should be noted that for 10^{20} cm^{-3} doping, only 0.1 eV of band bending is required to saturate a surface-state density of 10^{13} cm^{-2}. Further work is needed to evaluate the likelihood of this possibility.

Measurement of contact resistance to these carbon-doped p^+-GaAs layers has been made to verify that excellent ohmic contacts can be fabricated at the same time that low GaAs SRV is achieved. Contact resistance measurements were made on 0.25-μm-thick p^+-GaAs films grown under similar conditions as the structures used for the optical measurements described above. Mesa-isolated TLM patterns were utilized for the contact resistance measurements. A Ti/Pt/Au metallization stack with a sheet resistance of less than 0.025 Ω/□ was employed and linewidths were measured with a scanning electron microscope on all samples to assure accuracy of the contact resistance measurement. The specific contact resistance for a sample with a doping of 8×10^{19} cm^{-3} was measured to be in the range $(3\text{–}9)\times10^{-7}$ Ω cm^2. The temperature dependence of the contact resistivity has also been measured from 77 to 340 K, with the result that the contact resistivity typically increased from about 7.5×10^{-7} Ω cm^2 at 77 K to 7.9×10^{-7} Ω cm^2 at 340 K. This weak dependence of the contact resistance on temperature indicates a tunneling contact arising from the narrow surface depletion layer. Furthermore, temperature-dependent Hall effect measurements indicate that such heavily carbon-doped GaAs samples exhibit negligible carrier freezeout down to the lowest measurement temperature of 10 K, which ensures the persistance of a narrow depletion layer for hole tunneling at quite low temperatures.

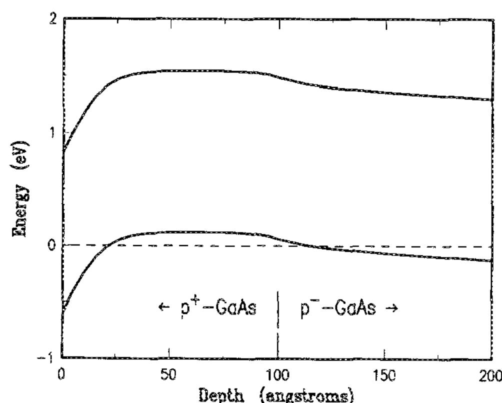

FIG. 3. Energy-band diagram for the p^+/p GaAs structure assuming a barrier height of 0.6 eV due to surface Fermi level pinning.

In summary, we have shown that carbon-doped p^+/p-GaAs structures are effective in reducing both the SRV and metal contact resistance simultaneously. For other passivating structures, e.g., GaAlAs/GaAs, the interface recombination velocity is quite low, but the high contact resistance to p^+-GaAlAs could limit the performance of some devices. Thus, the p^+/p structure should be most useful in device applications requiring both low SRV and simplified ohmic contact technology.

The authors DY and FP wish to acknowledge partial support of the New York State Science and Technology Foundation as part of its Centers for Advanced Technology program and support through an IBM Shared University Research contract. D. McInturff is thanked for his assistance with C-V measurements and A. C. Warren for use of HETMOD (a heterostructure Poisson equation solver) for the band diagram calculation.

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