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H. L. Chuang
Purdue University

Paul David DeMoulin
Purdue University

M. E. Klausmeier-Brown
Purdue University

Michael R. Melloch
Purdue University, melloch@purdue.edu

Mark S. Lundstrom
Purdue University, lundstro@purdue.edu

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H. L. Chuang, P. D. DeMoulin, M. E. Klausmeier-Brown, M. R. Melloch, and M. S. Lundstrom
School of Electrical Engineering, Purdue University, West Lafayette, Indiana 47907

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The electrical performance of Be-doped, p-p^+ GaAs homojunction barriers is characterized and analyzed. The results of the analysis show that minority-carrier electrons, at 300 K, have a mobility of 4760 cm^2/V s at a hole concentration of 2.3×10^{16} cm^{-3}, and that the effective recombination velocity for these homojunction barriers is about 6×10^4 cm/s. We present evidence that this unexpectedly high recombination velocity is a consequence of an effective reduction in band gap due to the heavy impurity doping. The effective band-gap shrinkage in this Be-doped material grown by molecular-beam epitaxy appears to be comparable to that already observed for Zn-doped GaAs grown by metalorganic chemical vapor deposition. This work demonstrates that so-called band-gap narrowing effects significantly influence the electrical performance of GaAs devices.

I. INTRODUCTION

Strong heavy doping effects have recently been reported for Zn-doped GaAs grown by metalorganic chemical vapor deposition (MOCVD).[1] In this paper we present evidence that comparable effects of similar magnitude also occur in Be-doped GaAs grown by molecular-beam epitaxy (MBE). The experiments utilized p-p^+ homojunction barriers commonly used to confine minority carriers in GaAs solar cells. Previous work showed that the effective recombination velocity associated with such a barrier was much too high for effective minority-carrier confinement.[2] Significantly better performance has been achieved by replacing the isotype homojunction barrier with an isotype heterjunction barrier in n-p GaAs solar cells.[3] The work we report suggests that the poor confinement of minority carriers by these homojunction barriers is due to an effective reduction in the band gap associated with heavy Be doping.

A successive etch technique[4] was employed to estimate the recombination velocity of a p-p^+ homojunction barrier. The barrier recombination velocity was found to be about 6×10^4 cm/s, and it increased as the width of the p^+-barrier layer decreased. Recombination through defects at the p-p^+ interface cannot explain these results, but the occurrence of band-gap narrowing effects[5] in p^+-GaAs can. The amount of band-gap narrowing deduced from the measurements is consistent with that measured for Zn-doped GaAs grown by MOCVD.[1] These results demonstrate that band-gap narrowing effects significantly influence the electrical performance of devices containing p^+-GaAs regions. They underscore the need to characterize such effects for Be-doped GaAs grown by MBE.

II. EXPERIMENTAL TECHNIQUE

The epitaxial layer structure for the solar cells used in this study is shown in Fig. 1. The films were grown in a Perkin–Elmer PHI-400 MBE system. The starting substrate was cleaved from a (100)-oriented, n-type GaAs wafer, and the thicknesses of the epitaxial layers were determined by counting oscillations in the intensity of the reflection high-energy electron diffraction pattern. Silicon was used as the n-type dopant and beryllium as the p-type dopant. Solar cells of dimension 0.1×0.1 cm^2 were defined by photolithography and subsequent wet etching. The p-type contact was a Au:Zn metal finger pattern which covered 18.4% of the cell area and formed a nonalloyed contact to the p^+-GaAs cap layer. The back contact metal was indium. The doping density of the P layer was measured as 2.3×10^{16} cm^{-3} by capacitance versus voltage profiling. Doping densities of the other layers were estimated from the growth rate of the film and the temperature of the dopant oven.

The completed cells were characterized by current versus voltage (I-V) measurements performed with a Hewlett–Packard 4145A semiconductor parameter analyzer. All I-V measurements were performed in the dark at about 23.3 °C. The measured current density versus applied voltage can be described by

$$J = J_{01}(e^{qV/kT} - 1) + J_{02}(e^{qV/2kT} - 1)\,, \qquad (1)$$

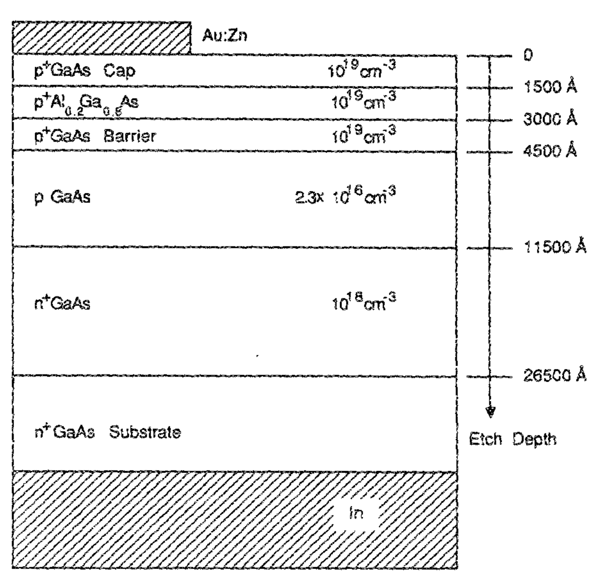

FIG. 1. Epitaxial layer structure of the solar cells used in this study.

where J_{01} and J_{02} are the saturation current densities associated with carrier recombination in the quasineutral and space-charge regions, respectively. The dark *I-V* characteristics were fitted to Eq. (1) to determine the two saturation current densities.

A successive etch technique was used to characterize the electron injection current.[4] The metal grid pattern was protected with photoresist, and the exposed semiconductor was removed in a series of short etches. Each etch was 20 s long in a solution of $[8H_2SO_4{:}4H_2O_2{:}400H_2O]$ at 26 °C and removed 375 Å of material as measured by step profiling. After each etch, the forward-biased dark *I-V* characteristic was measured.

III. ANALYSIS AND DISCUSSION

The dark *I-V* characteristic was measured after each etch step, and the resulting $n = 1$ saturation current density J_{01} is plotted in Fig. 2. J_{01} was roughly constant until the $Al_{0.2}Ga_{0.8}As$ layer was removed; it then increased as the p^+-GaAs barrier layer was thinned. When the p^+-barrier layer was completely removed, J_{01} increased sharply. This result clearly demonstrates that the heterojunction barrier is more effective than the homojunction barrier in minority-carrier confinement.

In a p-n^+ GaAs diode, the major component of J_{01} is due to electron injection in the *p*-GaAs and is given by

$$J_{01e} = \frac{qn_{io}^2}{N_A}\left(\frac{S + (D_n/L_n)\tanh(W_p/L_n)}{1 + S(L_n/D_n)\tanh(W_p/L_n)}\right), \quad (2)$$

where D_n and L_n are the minority-carrier electron diffusion coefficient and length, respectively, n_{io} is the intrinsic carrier concentration of lightly doped GaAs, W_p is the width, and S is the surface recombination velocity of the lightly doped *p* layer. If the *p* layer is thin ($W_p \ll L_n$) and the surface is unpassivated ($S \gg W_p/\tau_n$), then Eq. (2) can be simplified as

$$J_{01e} = \frac{qn_{io}^2 D_n}{N_A W_p}\,\frac{S}{S + D_n/W_p}. \quad (3)$$

Equation (3) should describe the electron injection current after the p^+-GaAs cap, the p^+-$Al_{0.2}Ga_{0.8}As$, and the p^+-GaAs barrier layers have been removed.

Since the measured $n = 1$ current component increased by more than a factor of 30 after the top three layers were removed, the measured J_{01} can be equated to Eq. (3). The width of the *p*-GaAs layer varied with time according to $W_p(t) = W_{p0} - Rt$, where W_{p0} is the width of the lightly doped thin *p* layer at $t = 0$, and R is the etch rate. Equation (3) can then be rearranged as

$$J_{01}^{-1} \simeq J_{01e}^{-1} = \left(\frac{N_A W_{p0}}{qn_{io}^2 D_n} + \frac{N_A}{qn_{io}^2 S}\right) - \left(\frac{N_A R}{qn_{io}^2 D_n}\right)t. \quad (4)$$

Figure 3 shows that a plot of J_{01}^{-1} versus etch time was linear with a slope of $N_A R/qn_{io}^2 D_n$, from which the product, $n_{io}^2 D_n$ at 23.3 °C, was determined to be 2.9×10^{14} cm^{-4} s^{-1}. From the intercept, a surface recombination velocity of 9.4×10^6 cm/s was deduced.

The measured surface recombination velocity agrees well with the value expected for a bare GaAs surface.[6] From the measured $n_{io}^2 D_n$ product and the data of Blakemore for n_{io},[7] a minority-carrier electron diffusion coefficient of $D_n = 123$ cm^2/s was deduced. This value corresponds to a minority-carrier electron mobility of 4760 cm^2/V s at 300 K and is 25% lower than the mobility of minority-carrier electrons in uncompensated GaAs as predicted by Walukiewicz *et al.*[8] Low minority-carrier mobilities have also been reported in *p*-GaAs doped much more heavily than that employed here.[9,10]

Consider next the situation in which the p^+-GaAs barrier layer was present. A theoretical expression relating the barrier recombination velocity, S_{pp^+}, to the structural parameters of the barrier, valid for both homojunction and heterojunction barriers, has been given by DeMoulin, Lundstrom, and Schwartz[11] as

$$S_{pp^+} = \frac{D_n^+ N_A^-}{L_n^+ N_A^+}\,\frac{n_{ie}^2}{n_{io}^2}\coth\left(\frac{W_p^+}{L_n^+}\right), \quad (5)$$

where the $-$ and $+$ superscripts refer to the lightly and heavily doped sides of the junction, respectively, and n_{ie} is the effective intrinsic carrier concentration of the heavily

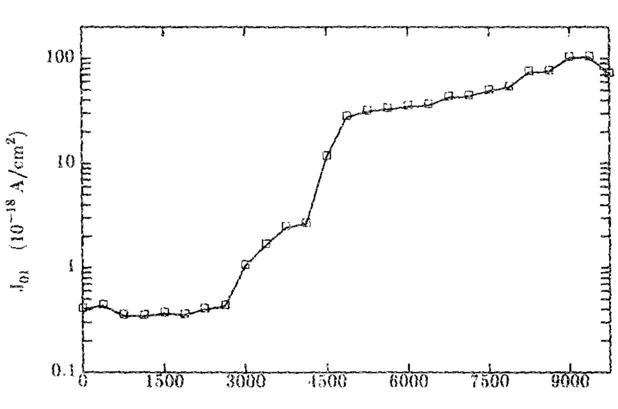

FIG. 2. $n = 1$ saturation current density J_{01} extracted from the measured dark current-voltage characteristic after each etch.

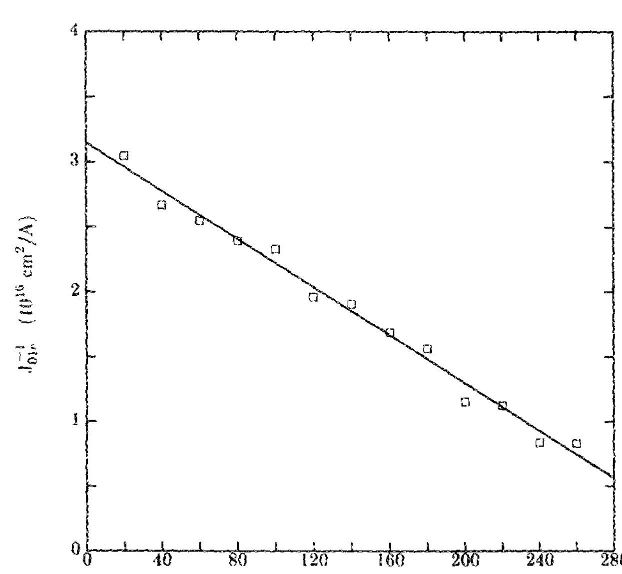

FIG. 3. J_{01e}^{-1} vs etch time was linear with a slope of $N_A R/qn_{io}^2 D_n$, from which D_n and μ_n were deduced.

doped side that accounts for the effective reduction in band gap associated with heavy impurity doping and for the influence of Fermi–Dirac statistics. In this case, since the p^+-GaAs barrier layer was only 0.15 μm thick, the assumption $W_p^+ \ll L_n^+$ is valid; Eq. (5) can thus be simplified as

$$S_{pp^+} = (D_n^+ N_A^- / W_p^+ N_A^+)(n_{ie}^2/n_{io}^2) . \qquad (6)$$

Thus, S_{pp^+} is expected to be inversely proportional to W_p^+.

S_{pp^+} can be related to the measured $n = 1$ component of the dark current; since $W_p \ll L_n$ and S_{pp^+} may be comparable to W_p/τ_n, Eq. (2) is simplified as

$$J_{01e} = \frac{qn_{io}^2}{N_A} \frac{S_{pp^+} + W_p/\tau_n}{1 + (W_p/D_n)S_{pp^+}} , \qquad (7)$$

which can be solved for

$$S_{pp^+} = \frac{1 - (qn_{io}^2/N_A J_{01e})(W_p/\tau_n)}{(qn_{io}^2/N_A J_{01e}) - (W_p/D_n)} . \qquad (8)$$

Since J_{01e} is very nearly the measured J_{01}, and D_n has been determined as described above, with an appropriate τ_n, Eq. (8) can be employed to estimate S_{pp^+}; values of S_{pp^+} versus etch depth were calculated for a few different values of τ_n. By plotting S_{pp^+} vs $1/W_p^+$ (see Fig. 4), it is found that the relationship is linear, and that a τ_n of 1.1 ns caused the straight line to pass through the origin, thus satisfying Eq. (6). This implies that a minority-carrier electron in our MBE-grown material of doping $N_A = 2.3\times10^{16}\ \text{cm}^{-3}$ has a lifetime of 1.1 ns. Taking the square root of the product $D_n\tau_n$, the minority-carrier electron diffusion length L_n was determined to be 3.7 μm; thus, the assumption $W_p \ll L_n$ used in Eqs. (3) and (7) is valid. Using Eq. (6) and the slope of the straight line that passes through the origin in Fig. 4, $n_{ie}^2 D_n^+$ was determined to be $1.8\times10^{15}\ \text{cm}^{-4}\ \text{s}^{-1}$, which agrees with the data measured by Klausmeier-Brown, Lundstrom, Melloch, and Tobin.[1] They found $n_{ie}^2 D_n^+ = 1.8\times10^{15}\ \text{cm}^{-4}\ \text{s}^{-1}$ for MOCVD-grown GaAs, Zn doped at $1\times10^{19}\ \text{cm}^{-3}$. They reported that the product $n_{ie}^2 D_n^+$ was affected by band-gap narrowing through an increase in n_{ie}. This implies that the effective band-gap shrinkage in Be-doped GaAs grown by MBE is comparable to that observed for Zn-doped GaAs grown by MOCVD.

A plot of measured S_{pp^+} versus etch depth is displayed in Fig. 5, which shows that S_{pp^+} is about 6×10^4 cm/s and that it increases as the thickness of the p^+-layer decreases. Since S_{pp^+} was found to depend on the thickness of the p^+ layer, it cannot be controlled by recombination at the doping junction but must be related to the properties of the barrier. Assuming no band-gap narrowing effects, theoretical values of S_{pp^+} were calculated using Eq. (6), in which n_{ie} equals n_{io} from Blakemore[7] corrected for hole degeneracy, and using D_n given by Walukiewicz *et al.* for uncompensated *p*-GaAs.[8] Figure 5 compares the theoretical estimate of S_{pp^+} with the measured value computed from Eq. (8). The figure shows that when band-gap narrowing effects were not considered, the theoretical estimate of S_{pp^+} was about 10 times lower than the value deduced from the measurements. The results clearly suggest that the high barrier recombination velocity is a consequence of an effective narrowing of the band gap of p^+-GaAs.

Lower values of S_{pp^+} may be achieved by making the p^+-GaAs barrier layer thicker. Suppose the p^+-barrier layer is so thick that W_p^+ is several times greater than L_n^+. Equation (5) can then be simplified as

$$S_{pp^+} = (D_n^+ N_A^- / L_n^+ N_A^+)(n_{ie}^2/n_{io}^2) . \qquad (9)$$

Using the product $n_{ie}^2 D_n^+$ determined above, our doping densities, and n_{io} from Blakemore,[7] S_{pp^+} was estimated to be $\lesssim 10^4$ cm/s, provided that L_n^+ is greater than 0.82 μm. However, because it is very difficult to determine the value of L_n^+, the possibility of achieving such a low value of S_{pp^+} by increasing the thickness is uncertain.

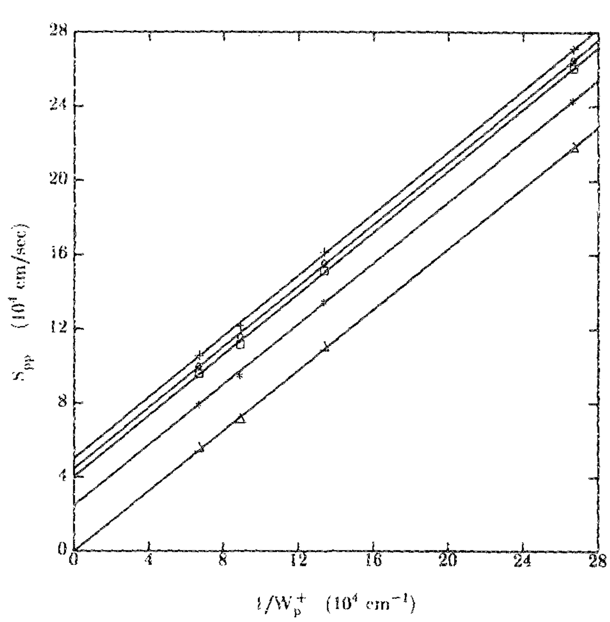

FIG. 4. Barrier recombination velocity S_{pp^+} vs thickness of the p^+-GaAs barrier layer plotted, from top to bottom, for minority-carrier electron lifetimes of 100, 10, 5, 2, and 1.1 ns, respectively.

IV. CONCLUSIONS

In this paper we employed a successive etch technique to study electron injection currents in GaAs p-n^+ diodes grown by molecular-beam epitaxy. The results show that the

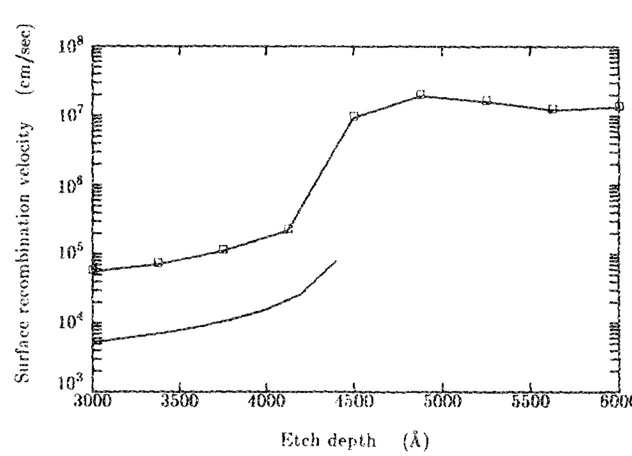

FIG. 5. Surface recombination velocity vs etch depth starting from the p^+-GaAs barrier layer. Top curve represents measured data. Bottom curve represents theoretical data without band-gap narrowing effects.

effective recombination velocity of the p-p^+ homojunction barriers in these diodes is about 6×10^4 cm/s. Analysis of these results strongly suggests that the high barrier recombination velocity is a consequence of an effective reduction in band gap caused by heavy Be doping. These effects appear to be comparable in magnitude to those reported for Zn-doped GaAs grown by metalorganic chemical vapor deposition.[1] The results also show that minority-carrier electrons, at 300 K, have a mobility of 4760 cm^2/V s at a hole concentration of 2.3×10^{16} cm^{-3}, thus confirming that the minority-carrier electron mobility in this moderately doped p-GaAs is lower than the majority-carrier mobility in comparably doped, uncompensated n-GaAs.

The results reported in this paper demonstrate that heavy doping effects significantly influence the performance of GaAs bipolar devices. This work helps to explain the substantial increase in solar cell performance that was observed when a homojunction p-p^+ barrier was replaced with an isotype heterojunction barrier.[3] We conclude that heterojunction barriers are essential for maximizing the efficiency of n-p GaAs solar cells.

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