Defects with deep levels in high-voltage gradual $p-i-n$ heterojunctions AlGaAsSb/GaAs

© M.M. Sobolev, F.Yu. Soldatenkov

Ioffe Institute,
194021 St. Petersburg, Russia
E-mail: f.soldatenkov@mail.ioffe.ru

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High-voltage gradual p^0-i-n^0 junctions of $Al_xGa_{1-x}As_{1-y}Sb_y$ with $x \sim 0.24$ and $y \sim 0.05$ in the i-region were studied using capacitance-voltage characteristics method and transient spectroscopy of deep levels. It has been established that the effective recombination trap in them is the DX-center of the Si donor impurity, with a thermal activation energy $E_t = 414$ meV, a capture cross section $\sigma_n = 1.04 \cdot 10^{-14}$ cm^2, and a concentration $N_d = 2.4 \cdot 10^{15}$ cm^{-3}. In the heterostructures studied, there were no deep levels associated with dislocations.

Keywords: AlGaAsSb, p^0-i-n^0 junction, DLTS, DX-center, liquid phase epitaxy.

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One of the main trends in modern pulsed power electronics is the use of wide-band semiconductor materials for the manufacture of devices with increased operating pulse repetition rates and operating temperatures. In the article [1], high-voltage lightly doped gradual p^0-i-n^0 junctions $Al_xGa_{1-x}As$, retaining rectifying properties up to 320°C, were obtained by liquid-phase epitaxy (LPE). It is possible to increase the performance of such devices due to the controlled formation of additional intrinsic structural defects during the growth of heterostructures with a given mismatch in the lattice parameter, for example, by introducing In or Sb [2] into lightly doped AlGaAs layers.

Epitaxial growth of high-voltage lightly doped gradual p^0-i-n^0 AlGaAsSb junctions was carried out using a modified LPE [1,2] method in a piston-type graphite container on p^+-GaAs substrates with (100) orientation, doped with zinc up to $5 \cdot 10^{18}$ cm^{-3}, from a limited solution-melt Al−Ga−As−Sb from 850 to 775°C, followed by growth of a tellurium-doped up to $2 \cdot 10^{18}$ cm^{-3} n^+-GaAs emitter layer. The composition of the liquid phase to obtain $Al_xGa_{1-x}As_{1-y}Sb_y$ solid solution layer of the required composition was determined from calculations in the quasi-regular solution model. The method for producing gradual $Al_xGa_{1-x}As_{1-y}Sb_y$ p^0-i-n^0 junctions from one solution-melt due to self-doping with background impurities is not fundamentally different from the method for producing $Al_xGa_{1-x}As$ p^0-i-n^0 junctions described earlier in the work [1]. Figure 1 shows the change (according to calculations in the regular solution model) of the composition x (curve *1*) and y (curve *2*) over the thickness of the studied epitaxial structure $Al_xGa_{1-x}As_{1-y}Sb_y$. The location of the $p-n$ junction boundary (Figure 1) was determined using an optical microscope from a chip of a sample oxidized by electrolytic anodization.

From Figure 1 it follows that the space charge region (SCR) of the reverse-biased $Al_xGa_{1-x}As_{1-y}Sb_y$ p^0-i-n^0 junction is located along the layer thickness in the composition range x, at which deep levels (DL) can appear in the DLTS spectra configurationally bistable DX-center [1], the model of which was first proposed by the authors of the work [3].

The studies [1] of p^0-i-n^0 junctions $Al_xGa_{1-x}As$, fabricated by the LPE method, carried out in this work allowed to identify in them DX-centers with negative correlation energy U [4–6], associated with Te and Se impurities. It should be assumed that the epitaxial layers $Al_xGa_{1-x}As_{1-y}Sb_y$ studied in this work in the region of the $p-n$ junction may also contain DX-centers. To identify DX-centers in p^0-i-n^0 heterostructures, studies of $C-V$ characteristics were carried out at 300 and 87 K under different measurement conditions (Figure 2). In case

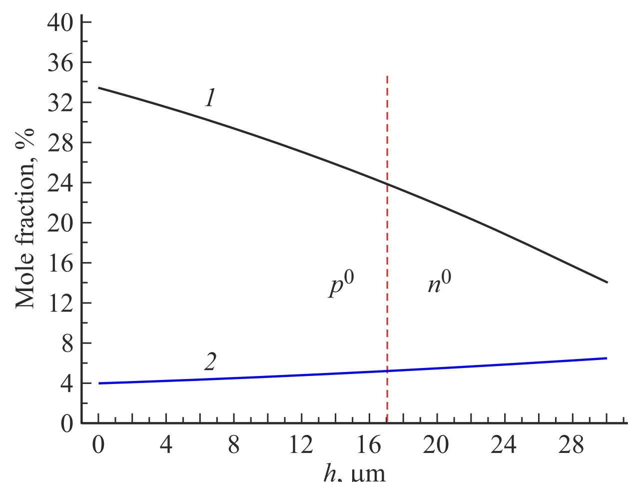

Figure 1. Distribution of mole fractions x (curve *1*) and y (curve *2*) over the thickness h of the epitaxial p^0-i-n^0 junction $Al_xGa_{1-x}As_{1-y}Sb_y$; the vertical line marks the border of the $p-n$ junction.

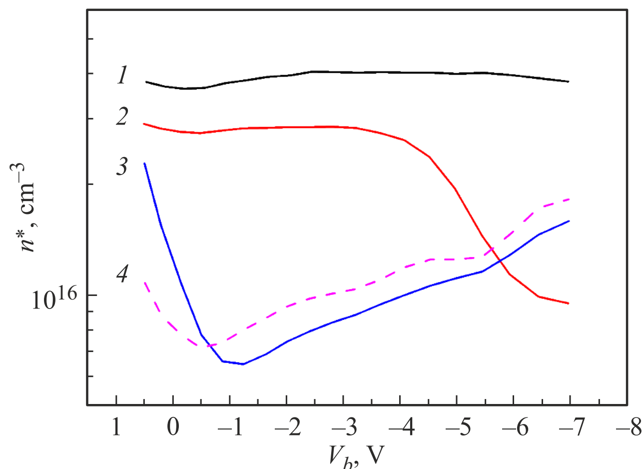

Figure 2. Distribution profiles of the effective concentration of free charge carriers n^* depending on the magnitude of the bias voltage V_b applied to the$Al_xGa_{1-x}As_{1-y}Sb_y$/GaAs $p^+-p^0-i-n^0-n^+$ heterostructure, measured at different temperatures T, K: *1* — 300, *2, 3, 4* — 87; in the dark (*1, 2, 3*) and with optical illumination (*4*); after pre-cooling the sample under the condition $V_r < 0$ (*2*), $V_r = 0$ (*3, 4*).

of measurements at 87 K, the sample was cooled either with the reverse bias voltage $V_r < 0$ or $V_r = 0$ turned on (Figure 2). We have previously used the technique of such measurements to identify bistable defects in work [1].

For epitaxial p^0- and n^0-layers $Al_xGa_{1-x}As_{1-y}Sb_y$ from $C-V$ characteristics, using the differential capacitance method [7], the distribution profiles of the effective concentration of free charge carriers (n^*) over the thickness of the space charge region (SCR) were calculated (the figure is not shown) and depending on the magnitude of the bias voltage V_b applied to the $p-n$ junction (Figure 2). The following features of the $C-V$ characteristics of the studied structures were identified: 1) as the sample measurement temperature decreases from 300 to 87 K, the thickness of the SCR layer increases from ~ 0.5 to $0.9\,\mu$m; 2) in the case of measuring the $C-V$ characteristics of a sample at 87 K and pre-cooling it at $V_r = 0$ (Figure 2, curve *3*), the value of n^* decreases from $\sim 4 \cdot 10^{16}$ to $\sim 6 \cdot 10^{15}\,\mathrm{cm}^{-3}$ (Figure 2, curves *1* and *3*, respectively, at $V_b = -1$ V); 3) in addition, the $C-V$ characteristic obtained with illumination (Figure 2, curve *4*) remained unchanged for a long time at low temperature after turning off the lighting, which is explained by what is known as the residual photoconductivity effect [3,4]. All of the above features of the results of $C-V$ measurements are typical for samples in which configurationally bistable donor impurities (Si, Se, Te) are present, which have two types of donor states: shallow and deep DX-level, which is a self-capture trap. However, when the sample is cooled to 87 K with $V_r < 0$, with the subsequent measurement of the $C-V$ characteristic upon reaching $V_b \approx -4.0$ B, the effective concentration n^* drops quite quickly (Figure 2, curve *2*). At the same time, for the effective concentratio n^*, determined from the $C-V$ characteristic and measured when the sample was cooled to 87 K with $V_r = 0$, starting from the value $V_b \approx -1.0$ B, an increase in n^* is observed (Figure 2, curve *3*). This effect will be explained below.

Further, for the diode structure $Al_xGa_{1-x}As_{1-y}Sb_y$/GaAs $p^+-p^0-i-n^0-n^+$, DLTS spectra were measured at different values of filling pulses, $V_f = -0.01$, -1.00, -4.50 and $+0.50$ B, at bias voltages $V_r = -2.10$, -3.0 and -6.0 B (Figure 3). These spectra demonstrated the presence of three DLTS peaks, all with a positive polarity, but with different signal amplitudes. It should be noted that the positive sign of the DLTS $E1$ peak (Figure 3) is unusual for the DLTS spectrum measured for a fill pulse with $V_f = -0.01$ B. For a defect that is a trap of majority carriers, the DLTS peak at this value V_f should have a negative sign. Everything happens differently in $p-n$ structures with DX-centers with negative correlation energy U, which are characterized by the presence of three charge states: a shallow donor DX^+ and two levels with electrons in neutral (DX^0) and negatively charged (DX^-) states (acceptor-like level) [1,4–6]. The work [6] presents the most adequate configuration-coordinate diagram for different charge states of the DX-center in $Al_xGa_{1-x}As$. When measuring DLTS spectra for such defects at $V_r = -2.1$ B, an SCR is formed in the n^0-layer, in which electrons emitting from acceptor-like levels DX^- are carried away by the strong electric field of the SCR. The donor level, designated D^+, then becomes positively charged with a large electron capture cross section. At the same time, the deep DX^0 level turns out to be neutral and has a smaller capture cross section than that of the donor D^+ level. The capacity of the $p-n$ junction decreases. After applying a filling pulse with $V_f = -0.01$ B at a filling pulse duration sufficient to capture two electrons at the levels D^+ and DX^0, the first electron is

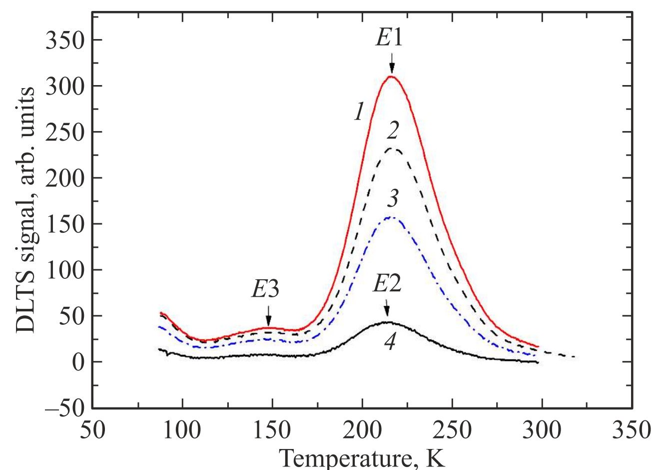

Figure 3. DLTS spectra of $Al_xGa_{1-x}As_{1-y}Sb_y$/GaAs $p^+-p^0-i-n^0-n^+$ diode, with window rate $200\,\mathrm{s}^{-1}$, at bias voltage $V_r = -2.1$ B and fill pulse voltage$V_f = +0.50$ B (*1*), $V_r = -2.1$ B and $V_f = -0.01$ B (*2*), $V_r = -3.00$ B and $V_f = -1.0$ B (*3*); $V_r = -6.00$ B and $V_f = -4.5$ B (*4*).

captured at the donor level D^+, and the defect goes into the state DX^0 in accordance with the reaction $D^+ + e^- \to DX^0$. After this, the defect almost immediately captures a second electron, and becomes DX^- ($DX^0 + e^- \to DX^-$). At the same time, the capacity of the $p{-}n$ junction increases. Then, after turning on the fill pulse and reverse biasing with $V_r = -2.1$ B, two electrons are emitted, closely following each other. Since the emission of the first electron takes significantly longer than the emission of the second electron, the observed rate of emission will actually be the same as that of the first electron [1].

This means, firstly, that deep states DX^-, which are traps for majority carriers, give a positive DLTS signal; secondly, the DLTS signature of level $E1$, found from the Arrhenius dependence, may belong to the DX-state. For $E1$ thermal activation energy level E_t, capture cross section σ_n and concentration N_d were respectively $E_t = 414$ meV, $\sigma_n = 1.04 \cdot 10^{-14}\,\mathrm{cm}^2$, $N_d = 2.4 \cdot 10^{15}\,\mathrm{cm}^{-3}$. The parameters E_t and σ_n were identical for the DLTS peaks of the spectra *1, 2* and *3* (Figure 3), and were close in parameters to the DX-center of the donor impurity Si. For the spectrum *4*, the amplitude of the DLTS peaks decreased by approximately 7 fold, the DLTS signature of the level $E2$ also changed: $E_t = 345$ meV, $\sigma_n = 2.43 \cdot 10^{-15}\,\mathrm{cm}^2$. As it was shown above (see Figure 2), when the sample is cooled to 87 K with $V_r < 0$ and subsequent measurement of the $C{-}V$ characteristic when $V_b \approx -4.0$ V is reached, the effective concentration n^* drops quite quickly. These metamorphoses can be caused by a gradient change in the composition of the $Al_xGa_{1-x}As_{1-y}Sb_y$ solid solution along the thickness of the epitaxial n^0-layer (see Figure 1), in which the deep DX-level appears higher or in resonance with the Γ-minimum of the conduction band. In this case, the deep DX^-state becomes metastable, and the DX^+ effective mass state associated with the Γ-valley becomes stable. In this case, taking into account that the epitaxial layers contain dislocations, as a result of the presence of mechanical pressure on the lattice, the ability of the DX-center to capture electrons arises, even when it is located above the bottom of the conduction band. In this case, a peak associated with the DX-center can be observed in the DLTS spectra, but with a significantly lower amplitude [4,5].

In addition to the levels of the deep DX^-state, a peak $E3$ is detected in the DLTS spectra, with the following parameters: $E_t = 220$ meV, $\sigma_n = 1.61 \cdot 10^{-16}\,\mathrm{cm}^2$, $N_d = 2.4 \cdot 10^{14}\,\mathrm{cm}^{-3}$. The appearance of this peak may be due to the presence of Se and Te [1] donor impurities in the layer.

It should be noted that, although previously (see work [2]) deep levels associated with dislocations were identified in the DLTS spectra of $GaAs_ySb_{1-y}$ layers with y from 1.5 to 3%, in the studied $Al_xGa_{1-x}As_{1-y}Sb_y$/GaAs heterostructures with y up to 6.5% they were absent. This may be caused by a change in the dislocation structure of the heterolayers with an increase in Sb content, and a resulting mismatch in the lattice parameters between the layer and substrate.

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Conflict of interest

The authors declare that they have no conflict of interest.

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