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ELECTRON SCATTERING BY NATIVE DEFECTS IN UNIFORMLY AND MODULATION DOPED SEMICONDUCTOR STRUCTURES

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ABSTRACT

Formation of native defects in GaAs is described in terms of the amphoteric native defect model. It is shown that Fermi energy induced formation of gallium vacancies is responsible for the limitations of maximum free electron concentration in GaAs. The effect of the defects on electron mobility in heavily doped n-GaAs is quantitatively evaluated. Defect scattering explains the abrupt reduction of electron mobility at high doping levels. Also, it is demonstrated that native defects are responsible for the mobility reduction in inverted modulation doped GaAs/AlGaAs heterostructures. The amphoteric defect model also explains a distinct asymmetry in defect formation in n- and p-GaAs. In p-GaAs the Fermi level induced reduction of the defect formation energy is much smaller, and therefore the concentration of the native defects is negligible compared with the hole concentration.

I. INTRODUCTION

Native (or intrinsic) defects play a very significant role in determining electrical and optical properties of semiconductors. Numerous characteristics of semiconductor devices are critically affected by the presence of native defects introduced during crystal preparation and/or device fabrication. This issue is especially important in compound semiconductors where a large variety of different native defects can exist.

One of the long standing and poorly understood problems in semiconductors is the extent to which native defects affect free carrier mobilities. Carrier scattering by native defects is very often invoked to explain unusually low mobilities observed in semiconductor crystals. However, since in most cases the nature of the defects as well as their concentrations are not known, such explanations are only qualitative. A quantitative treatment would require a knowledge of the microscopic potential introduced by the defect as well as the concentration and distribution of the defects in the crystal. Dislocations were among the most extensively studied carrier scattering centers [1-4]. Both charged and neutral dislocations were considered in these early studies. In most cases, however, it was difficult to ascertain a detailed form of the potential necessary for a description of the scattering process. The situation is even more unclear in the case of scattering by point defects where both the microscopic nature of the defects and their concentrations are hardly ever known.

In this paper we present model calculations of the effects of native defects on carrier mobility in GaAs. We utilize the recently developed concept of amphoteric native defects [5,6] to calculate the concentration of ionized native defects in heavily doped GaAs crystals. We find a large difference in the defect formation in n- and p-type GaAs. This has important consequences for the limitations of free carrier concentration and carrier mobility in heavily doped material. We also consider the effect of native defects on electron and hole mobilities in modulation doped heterostructures. Free electron induced generation of native defects is found to be responsible for reduced two-dimensional electron-gas mobility in n-GaAs/AlGaAs inverted-modulation-doped-heterostructures. In normal-modulation-doped heterostructures mobilities are high because defect formation during epitaxy is not enhanced by free carriers.

II. NATIVE DEFECTS IN HEAVILY-DOPED GaAs

It has been known for more than two decades that heavy doping of GaAs with donors to levels higher than ~ 2×10^{18} cm^{-3} leads to a considerable reduction of the intensity of band edge luminescence [7,8]. The reduction of the luminescence intensity is always associated with the appearance of a broad luminescence line at about 1.2 eV [8]. Extensive annealing studies have

led to the conclusion that the center responsible for the deep luminescence is associated with Ga deficiency in the crystals. On the basis of these studies it has been proposed that the deep center is a donor-gallium vacancy complex (D-V_{Ga}) [9]. This identification of the deep center appears to be consistent with measurements of local vibrational mode absorption in Si-doped MBE grown GaAs, which indicates the presence of a large concentration of complexes with Si_{Ga} and an unidentified native defect on one of the next nearest neighbor sites [10]. Recent studies of MOCVD grown GaAs have confirmed that gallium deficiency promotes formation of the deep centers [11]. An increase in the concentration of deep centers is always associated with reduction of the concentration of electrically active donors [7,11] and electron mobility [11]. All these results are consistent with the assumption that the D-V_{Ga} complex is acting as a deep acceptor, compensating intentionally introduced donors and contributing to the electron scattering.

The doping induced formation of the deep centers in n-type GaAs can be contrasted with p-type material where it has been found that doping to the levels close to 10^{20} cm^{-3} does not significantly affect the band-edge luminescence [12]. Also, no prominent deep level-luminescence is detected in p-type material [12]. This indicates that unlike in n-type GaAs, the acceptor doping does not induce the formation of compensating native defects. Such a conclusion is consistent with the very high electrical activity observed for acceptors in heavily doped p-GaAs [13].

In principle, a quantitative evaluation of the effects of native defects on free carrier concentration and/or carrier mobility requires a detailed knowledge of the defect formation energy. It has been shown recently that formation of native defects can be described in terms of the so-called amphoteric native defect (AND) model [5,6]. According to the model, incorporation of simple vacancy-like defects is controlled by the defect reactions

$$As_{As} + V_{Ga} \leftrightarrow (V_{As} + As_{Ga}) \qquad (1a)$$

$$\underset{\text{Acceptor}}{(Ga_{As} + V_{Ga})} \leftrightarrow \underset{\text{Donor}}{V_{As} + Ga_{Ga}} \qquad (1b)$$

Both reactions involve a jump of a single atom between adjacent lattice sites. The most important feature of these reactions is that they lead to the transformation of an acceptor-like native defect (LHS of 1a and 1b) into a donor-like defect (RHS of 1a and 1b). Therefore one expects that the formation energy of the defects and directions of reactions (1a) and (1b) will depend on the location of the Fermi energy. Based on calculations of the total defect energies [14], one obtains in good approximation the formation energy of the defects given by reaction (1a):

$$E_f(D_\pm) = U_f^o \pm 3E_F \qquad (2)$$

where $D_- \equiv V_{Ga}$, $D_+ \equiv (As_{Ga}+V_{As})$, E_F is the Fermi energy measured with respect to the Fermi level stabilization energy, i.e., the energy at which $E_f(D_-) = E_F(D_+) = U_f^o$. For reaction (1a) E_{FS} is located in the band gap at about 0.6 eV above the valence band edge [6].

The Fermi level stabilization energy E_{FS} plays a crucial role in the amphoteric native defect model. It is an energy reference to calculate the Fermi level induced reduction of the defect formation energy. The location of E_{FS} with respect to the conduction and the valence band edges determines the propensity of the semiconductor to form acceptor- or donor-like native defects, which compensate intentionally introduced shallow donors or acceptors, respectively. Since, as is seen from eq. (2) the defect formation energy strongly depends on the Fermi level position, one expects that defect incorporation will be strongly affected by the type of doping and the doping level.

In the following we consider reaction (1a), which corresponds to arsenic-rich conditions. It should be noted, however, that similar considerations are applicable to reaction (1b) as well. According to eq. (2) in n-type GaAs, $E_F > E_{FS}$ with:

$$[V_{Ga}] = C' \exp\left(\frac{3E_F}{kT}\right) \qquad (3)$$

where C' is a constant dependent on U_f^o. The presence of triply ionized V_{Ga} acceptors reduces the free carrier concentration which is given by:

$$n = N_D^+ - 3[V_{Ga}] \tag{4}$$

where N_D^+ is the concentration of ionized shallow donors. A standard relationship between n and E_F is:

$$n = N_c F_{1/2}\left[(E_F - E_c)/kT\right] \tag{5}$$

Equation (5) completes the set of equations required to find $[V_{Ga}]$ and n as functions of the doping level N_D^+. In eq. (5) E_c and N_c are the conduction band edge energy and the density of states, respectively, and $F_{1/2}(2)$ is the Fermi-Dirac integral.

In order to solve the set of equations 3 to 5 one needs to know the location of E_c with respect to E_{FS} at elevated temperatures. As has been discussed previously [15] the major contribution to temperature dependence of the energy difference $E_c - E_{FS}$ comes from the lattice dilation. It has been found [15] that $d(E_c - E_{FS})/dT = -1.7\times10^{-4}$ eV/K. Since the temperature coefficient of the GaAs energy gap is -4.8×10^{-4} eV/K one obtains the value of -3.1×10^{-4} eV/K for the temperature coefficient of the energy difference $(E_{FS} - E_v)$.

In heavily doped semiconductors the location of the conduction (valence) band edge, E_c, (E_v) is modified by Coulomb interactions so that $E_{c,v} = E_{c,v}^o + \Delta E_{c,v}$. There are two contributions to the band edge shifts, $\Delta E_{c,v}$:

i) the shift of the conduction band edge due to electron-electron interaction [16]

$$\Delta E_1 = -\frac{2e^2 k_F}{\pi\varepsilon_o} - \frac{e^2\lambda}{2\varepsilon_o}\left[1 - \frac{4}{\pi}\tan^{-1}\left[\frac{k_F}{\lambda}\right]\right] \tag{6}$$

where the Fermi wave vector is given by the relation $k_F^3 = 3\pi^2 n$,
$\lambda = 2e/\hbar\,(3n/\pi)^{1/6}\,(m^*/\varepsilon_o)^{1/2}$ is the Thomas-Fermi screening parameter, m^* is the electron hole effective mass at the Fermi energy, or $\varepsilon_o = 12.9$ is the static dielectric constant, and n is either the electron or hole concentration.

ii) The second contribution to the band edge shifts comes from electron-ionized impurity interaction [16]

$$\Delta E_2 = -\frac{4\pi n e^2}{\varepsilon_o a_o \lambda^3} = -\frac{e\hbar}{2}\left(\frac{\pi^3 n}{\varepsilon_o m^*}\right)^{1/2} \tag{7}$$

where $a_o = \varepsilon_o\hbar^2/m^*e^2$ is the effective Bohr radius.

Using eqs. (6) and (7) one can calculate the location of E_c and E_v with respect to E_{FS}, and then with the help of eq. (5) determine the location of the Fermi energy as a function of electron or hole concentration. The results of the calculation are presented in fig. 1. The calculations were performed for T=900 K, which is the temperature typical for the MBE growth of GaAs. It is seen from fig. 1 that for low electron (hole) concentration, E_F is determined by the intrinsic concentration and located at about 0.3 eV above E_{FS}. In n-type GaAs E_F moves upward in energy and at $n=10^{19}$ cm^{-3}, it reaches the value of ~ 0.8 eV. This rapid shift of the Fermi level in n-type GaAs can be contrasted with a much slower dependence of E_F on the hole concentration. As is shown in fig. 1, even at the highest hole concentrations ~ 10^{21} cm^{-3} $(E_{FS} - E_F)$ is smaller than 0.6 eV. The asymmetry in the Fermi level behavior in

n- and p-type GaAs has important consequences for the relative abundance of native defects in the two types of materials.

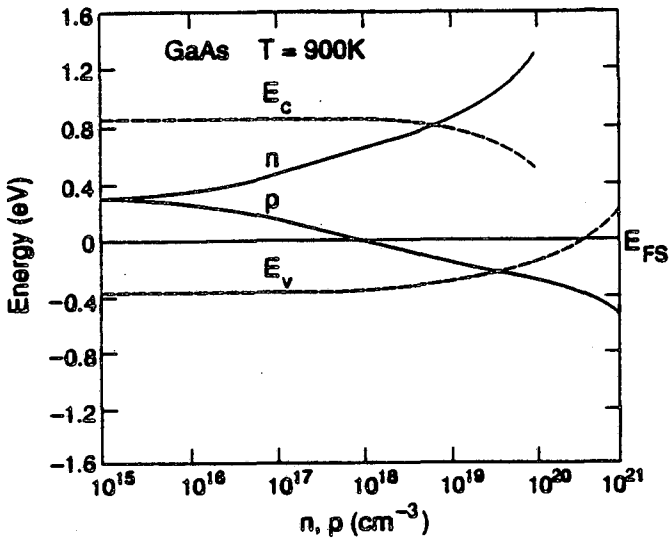

FIG. 1. Fermi energy, measured with respect to E_{FS}, as a function of free carrier concentration in n- and p-type GaAs (solid lines). Renormalized conduction (E_c) and valence (E_v) band edges are also shown.

It has been known from extensive studies [17,20] of GaAs that there is a limit on the maximum electron concentration which can be achieved by doping with shallow donors. Independently of the doping technique the portion of electrically active donors is considerably reduced for the doping level, exceeding about 10^{18} cm^{-3}. This reduction of electrical activity is not strongly dependent on the donor species and is very similar for amphoteric donors, such as Si and Sn, and for group VI donors S, Se, etc. This phenomenon can be easily understood within the amphoteric concept of Fermi-level induced formation of native defects.

To calculate the donor activation efficiency in GaAs we solve eqs. 3 to 5 to obtain $n(N_D^+)$. Fig. 2 shows the results of such calculations. It is seen that at low doping levels, when $[V_{Ga}]$ is small, $n=N_D^+$ and all donors are electrically active. With increasing doping levels an upward shift of the Fermi energy lowers the formation energy of $[V_{Ga}]$, which compensates shallow donors. This leads to a sublinear dependence in $n(N_D^+)$. At high doping levels one obtains $n \sim (N_D^+)^{1/3}$. The characteristic 1/3 power dependence reflects the fact that the compensating native defects are triply ionized acceptors. The calculations are in reasonably good agreement with experimental data on the activation of Se [17] and S [18] in GaAs.

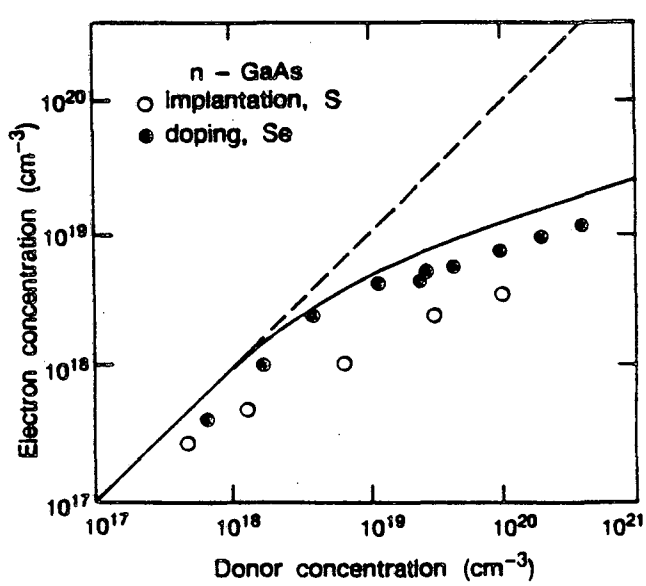

FIG. 2. Calculated electron concentration as a function of donor concentration in n-GaAs (solid line). Dashed line represents the case where all donors are electrically active. Experimental data was taken from Ref. 17 (Se) and Ref. 18 (S).

One can compare this with the case of p-type GaAs where it is known that shallow acceptors exhibit high activation efficiency even for very high doping levels exceeding 10^{20} cm^{-3} [13]. This is consistent with our present model since as is seen in fig. 1 holes produce a much smaller shift of the Fermi energy away from E_{FS}. Thus, for a hole concentration of about 10^{20} cm^{-3} $|E_{FS}-E_F|$ is less than half of the value of $|E_F-E_{FS}|$ in n-type GaAs, with the hole concentration of 10^{18} cm^{-3}.

III. ELECTRON AND HOLE MOBILITIES IN UNIFORMLY-DOPED GaAs

We have argued above that at high doping levels in n-type GaAs gallium vacancies compensate intentionally introduced donors. Since both compensating defects and donors are acting as ionized scattering centers, one expects that the increased incorporation of V_{Ga} will affect the electron mobility.

Calculations of electron mobility in n-GaAs are performed adopting standard procedure based on the variational solution of the Boltzman equation [21]. At high doping levels the electron mobility is largely determined by charged center scattering with a small contribution from the phonon scattering. The inverse electron mobility is

$$\mu^{-1} = \mu_{cc}^{-1} + \mu_{ph}^{-1} \tag{8}$$

where μ_{ph} is the room temperature phonon mobility limit ~ 8500 cm^2/V · s .

The mobility of electrons scattered by charged centers is [22]

$$\mu_{cc} = \frac{3\pi}{2}\frac{\varepsilon_o^2 \hbar^3}{e^3}\frac{n}{N_{cc} m^{*2} F_{cc}} \tag{9}$$

where $F_{cc} = \ln(\xi+1) - \frac{\xi}{\xi+1}$ and $\xi = 2k_F/\lambda$, and N_{cc} is the total concentration of scattering centers. The nonparabolicity of the conduction band has been incorporated in the calculation via an energy dependent effective mass m* [23]. Such an approach slightly overestimates the nonparabolicity effects by neglecting mixing of the valence and conduction band wavefunctions [22].

The concentration of the scattering centers N_{cc} depends not only on N_D^+ and V_{Ga}^{3-}, but also on their distribution in the crystal lattice. Thus, as has been discussed in the previous section, the positively charged donors tend to form complexes with negatively charged V_{Ga} [10,24]. Since such complexes act as doubly-charged centers, the total concentration of scattering centers is

$$N_{cc} = N_D^+ + 3[V_{Ga}] \tag{10}$$

The concentration of V_{Ga} was determined using eqs. 2 to 5. The calculations of the electron mobility, μ, were carried out using expression (10) for the concentration of charged scattering centers. The results of the calculations along with the experimental data on electron mobility in MBE grown n-GaAs [25,26] are shown in fig. 3. A rapid decrease of the electron mobility can be explained quantitatively by native defect scattering, assuming $U_f^o = 3.5$ eV for the formation energy of $[V_{Ga}]$ in eq. (2). The calculations also show that contrary to previous suggestions [25], the nonparabolicity effects are not very significant and cannot explain the abrupt decrease of the electron mobility. As is seen in fig. 3, conductivity in n-GaAs reaches a maximum value of about $2.4\times10^3\Omega^{-1}$ cm^{-1} at n ~ 1×10^{19} cm^{-3}. This appears to be a basic limit of electrical conductivity achievable in n-GaAs.

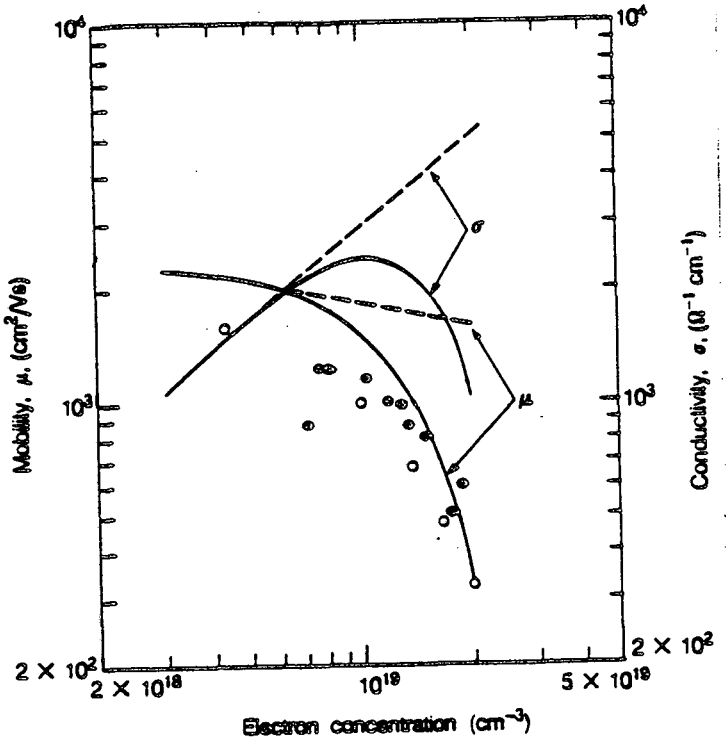

FIG. 3. Electron mobility and conductivity in heavily doped n-GaAs. The solid lines represent the calculations in which effects of native defects were included. The broken curves correspond to the standard case in which the concentration of charged scattering centers is equal to the carrier concentration. The experimental points represent the data of Ref. 25 (○) and Ref. 26 (●).

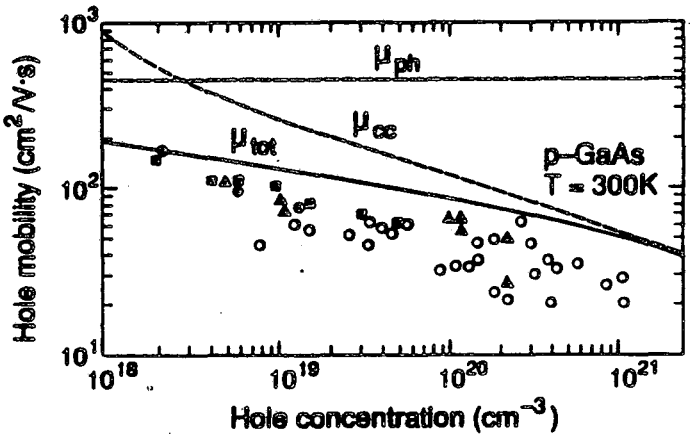

FIG. 4. Hole mobility in heavily doped p-GaAs. μ_{ph} and μ_{cc} are the mobilities due to phonon and charged center scattering, respectively. The combined hole mobility is represented by μ_{tot}. The experimental data for C (○ Ref. 13) and Be (●-Ref. 13, ❑-Ref. 28, ▲-Ref. 29) doped GaAs is also shown.

Examination of the results in figs. 2 and 3 shows that the effects of $[V_{Ga}]$ on the free electron concentration and mobility are becoming noticable for electron concentration $n \simeq 3$ to 5×10^{18} cm^{-3}. We find from fig. 1 that in n-GaAs this corresponds to $|E_F-E_{FS}| \simeq 0.7$ eV. In p-GaAs the value of $|E_F-E_{FS}|$ is much smaller and does not reach this level even for the highest hole concentration of 10^{21} cm^{-3}. This indicates that the concentration of the native donor ($As_{Ga}+V_{As}$) defect is always negligible compared with the free carrier concentration. This explains why acceptors show a much higher activation efficiency in GaAs. Also, it suggests that the contribution of the native defects to hole scattering should be negligible.

To test this suggestion we have calculated the hole mobility in p-GaAs. We use a model of partially coupled valence bands [27]. In this model the intra- and inter-band scatterings are included to calculate light- and heavy-hole mobilities. The mobilities are then combined in a standard way to obtain Hall mobility [27], which can be compared with experiments. The results of the calculations are shown in fig. 4. It is seen that at room temperature the phonon mobility limit is about 450 cm^2/V·s. This value of the mobility is determined mostly by the deformation potential and the polar optical phonon scatterings. The contribution of acoustic phonons is relatively insignificant. We have also calculated the mobilities resulting from scattering by charged centers. Assuming that the concentration of charged centers is equal to the

concentration of free holes, one finds that μ_{cc} depends very weakly on the hole concentration. For $p \geq 10^{19}$ cm^{-3} one finds $\mu_{cc} \sim p^{-1/3}$. This is the same dependence as that found for the mobility in heavily doped, uncompensated n-GaAs. It is also seen in fig. 4 that this weak dependence of the mobility on the hole concentration correctly predicts the experimentally observed mobilities for samples in a very wide concentration range [13,28,29]. This again confirms our finding that native defects do not play any significant role in p-GaAs. Discrepancies between the experimental and theoretical mobilities at very high hole concentrations can be attributed to the limitation of our model calculations which do not properly account for the anisotropy and energy dependence of the valence band effective masses at very high doping levels.

IV. ELECTRON MOBILITY IN MODULATION DOPED HETEROSTRUCTURES

In the case of uniformly doped n-type GaAs, the native defects are formed in heavily doped regions containing high concentrations of electrons and ionized impurities. However, since the formation of the defects is controlled by the Fermi energy, which depends on the electron concentration only, one can also expect an enhancement of defect formation in the systems in which electrons are separated from the impurities. Excellent examples of such systems are modulation doped heterostructures (MDH) [30], in which electrons confined in the quantum well are spatially separated from donors located in the barrier [31]. There are two kinds of MDHs [32]. In a normal-MDH (N-MDH), the undoped, well-forming layer is grown prior to the heavily doped barrier, whereas in an inverted-MDH (I-MDH), the heavily doped barrier is grown first followed by the undoped quantum well-forming layer. The basic difference between those two types of heterostructures is that in I-MDH the nominally undoped well is grown in the presence of a high concentration of free electrons transferred from the barrier forming layer. In N-MDH the electrons are transferred after the well is fully grown. Early studies of charge transport in AlGaAs/GaAs MDHs have shown that electron mobilities in I-MDH are always significantly lower than those in N-MDHs [32]. There were several attempts to explain this phenomenon either in terms of interface roughness [32] or impurity diffusion [33,34]. Here we show that the difference in the mobility values in these two structures can be understood easily in terms of the increased scattering by native defects formed in I-MDH.

In I-MDH the GaAs well is grown in the presence of electrons transferred from the AlGaAs barrier. This leads to the situation where $E_F \gg E_{FS}$, i.e., a condition which favors the formation of native defects. Such defects act as electron scattering centers. To estimate the effect of the defects on electron mobility we consider I-MDH with 2D electron gas density n_{2D}. The equivalent 3D density in the quantum well is $n = n_{2D}/W$, where $W = 2(\hbar^2\varepsilon_0/33\pi e^2 m^* n_{2D})^{1/3}$ is an effective width of the triangular quantum well [35]. The concentration of V_{Ga} is calculated using eqs. 3 to 5. For a typical I-MDH with $n_{2D} = 5\times10^{11}$ cm^{-2} we obtain $[V_{Ga}] \approx 6\times10^{14}$ cm^{-3}. Here we use the same value of the parameter U_f^o - 3.5 eV as the one determined for uniformly doped GaAs. Since the free electron concentration is much lower than in the case of uniformly doped GaAs, we can neglect the conduction band edge renormalization given by eqs. (6) and (7). The concentration of V_{Ga} in I-MDH is much lower than the free carrier concentration, therefore, the compensating effect of the defects is negligible. However, since in MDH, charged impurities are spatially separated from the electrons, the presence of even this small concentration of charged native defects in the quantum well can very significantly affect the 2D-electron gas mobility.

The electron mobility has been calculated using the method presented in Ref. 36. Acoustic and optical phonons, alloy disorder, remote and background charged center scattering are included in the calculations. Since the native defects are located in the quantum well they contribute to the background charged center scattering. The results presented in fig. 5 were obtained for the calculated concentration of V_{Ga} equal to 6×10^{14} cm^{-3}. This corresponds to a background concentration of charge scattering centers of 5.4×10^{15} cm^{-3}. It is seen in fig. 5 that this low concentration of native defects very profoundly affects the mobility. The low temperature mobility in a I-MDH is about an order of magnitude lower than the mobility in an equivalent N-MDH in which background defect scattering is negligible. This is in general agrement with experimental data which show that mobilities in N-MDH are considerably higher than those measured in a I-MDH. Lower mobilities in AlGaAs/GaAs I-MDH were previously explained by assuming additional scattering by charged centers located at the interface [37]. Without elaborating on the nature of the charged centers the authors of Ref. 37 concluded that

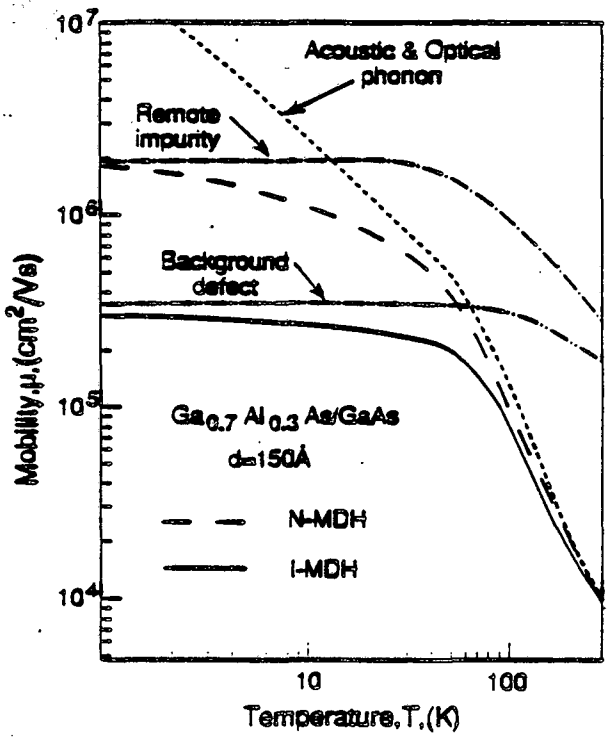

FIG. 5. Temperature dependent 2D-electron gas mobilities in $Al_{0.3}Ga_{0.7}As/GaAs$ MDHs with the spacer width d = 150Å. The broken and solid lines represent electron mobilities in normal- and inverted-MDHs, respectively. Contributions to the total mobilities resulting from different scattering mechanisms are also shown.

the interface charge density required to account for the experimental data has to be in the range 5×10^9 cm^{-2} to 3×10^{10} cm^{-2}. Since there is a great similarity between interface charge and background charge scatterings [36,38], the results of ref. 37 can be interpreted also in terms of scattering by background charged defects uniformly distributed in the well. The equivalent concentration of the background charges is 5×10^{15} cm^{-3} to 3×10^{16} cm^{-3}. It should be noted that our estimate of the effective charged center concentration of $N_{cc} = 5.4\times10^{15}$ cm^{-3} lies in this concentration range.

In the present calculations we have assumed $n_{2D} = 5\times10^{11}$ cm^{-3}, which is a typical concentration of 2-D electron gas in GaAs/AlGaAs MDH. A more detailed analysis of the native defect formation in I-MDH would require a proper description of the charge transfer from doped AlGaAs to the GaAs quantum well at the growth temperature. Such an analysis can be easily done for a specific structure with known doping levels and spacer width. The single most important factor controlling defect formation is the location of the Fermi energy in 2-D quantum wells. Therefore, the strategy to obtain an I-MDH with a low defect concentration is to lower the doping level and thus also the concentration of 2-D electron gas. Recently, a new type of I-MDH has been devised [39] which consists of an undoped GaAs/AlGaAs superlattice and a thick undoped GaAs layer. The electrons in the quantum well originate from planar doping in GaAs at a distance of 2500 Å away from the quantum well. It has been shown that the 2-D electron concentration can be controlled by the gate voltage and mobilities as high as 2×10^6 cm^2/V·s can be achieved in such a structure. This result provides strong support for our explanation of the lower mobilities in I-MDH. In this new structure the GaAs quantum well is grown without any electrons present. Therefore, it resembles the case of N-MDH where the doping is performed only after the well is fully grown.

All the above considerations apply to n-type AlGaAs/GaAs MDHs in which there is a substantial Fermi level induced reduction of the defect formation energy. As has been shown in section II, there is no significant reduction of the defect formation energy in p-type GaAs. Therefore, one does not expect any difference in the native defect scattering in p-type I-MDH and N-MDH. This provides an opportunity to differentiate the effects of native defect scattering from other scattering mechanisms [33,34] which were suggested to be responsible for lower mobilities in I-MDHs. Unlike the native defect scattering, the effectiveness of scattering processes due to interface roughness [33] and scattering by impurities diffusing from the heavily doped region [34] into the well, does not depend on the type of doping and therefore they should play an equally important role in p- and n-type MDHs. Unfortunately, to our knowledge, no comparative studies of mobilities in inverted and normal p-type AlGaAs/GaAs MDHs have been reported.

Besides the extensive studies of AlGaAs/GaAs MDHs, much information is available on InP/InGaAs MDHs. In the well-forming $In_{0.47}Ga_{0.53}As$ layer, the Fermi level stabilization energy $E_{FS} \equiv E_C - 0.25$ eV is located close to the conduction band edge. Thus, for the typical carrier concentration the effect of the Fermi level induced defect formation is very small. Consequently, one does not expect any difference in the mobility values in I-MDH and N-MDH. This conclusion is in agreement with results of a recent study [40] which has shown that practically the same mobilities are observed in inverted and normal InP/InGaAs MDHs. This again indicates that, at least in this system, the interface roughness and impurity diffusion effects appear not to be important.

CONCLUSIONS

The model of amphoteric native defects has been used, in recent years, to understand a large variety of phenomena in semiconductors [6] and at metal-semiconductor interfaces [5]. The most important feature of the model is that it provides the basis for a quantitative analysis of the native defect formation process. The incorporation of native defects is controlled by the location of the Fermi level measured with respect to the Fermi level stabilization energy, E_{FS}. Therefore, the location of E_{FS} with respect to the band edges determines how the native defect formation process depends on the type of doping and the doping level.

It has been shown previously that the model accounts well for the Schottky barrier heights at metal-semiconductor interfaces and explains pinning of the Fermi level by native defects during metal deposition on a semiconductor surface [5]. The Fermi level dependent formation of native defects has been used to explain the doping enhanced superlattice intermixing [6,41] and saturation of the free carrier concentration in semiconductors [6]. Also, an analysis of the effects of doping on supersaturation of native defects has provided the basis for a semi-quantitative treatment of the long-standing and controversial issue of doping induced suppression of dislocation formation in semiconductors [42,43].

In this paper we have demonstrated that within the same concept of amphoteric defects, one can calculate the contribution of native defects to carrier scattering. In particular, the calculations indicate that the abrupt reduction of electron mobility in heavily doped n-GaAs results from the Fermi-level induced formation of ionized gallium vacancies.

An interesting case is represented by modulation doped heterostructures in which electrons are spatially separated from the parent donors. The native defects are formed in the quantum well. There they act as very efficient scattering centers and play a dominant role in limiting the electron mobility in GaAs/AlGaAs inverted-modulation doped heterostructures.

It is found that there is a distinct asymmetry in defect formation in n- and p-GaAs. Even in very heavily doped p-GaAs the reduction of the defect formation energy and thus also an enhancement of the defect concentration is small and does not produce any of the effects observed in n-GaAs. Although the present considerations were limited to GaAs based systems, the concept of amphoteric native defects has a much broader scope. The known location of E_{FS} can be used, in principle, to describe the formation of defects in any semiconductors and semiconductor structures.

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