Defect identification in electron-irradiated GaAs

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(Received 30 June 1982)

The introduction rates of the main electron traps created by electron irradiations have been measured as a function of the composition of the $Ga_{1-x}Al_xAs$ solid solution system in the $0 < x < 0.5$ range; they are not x dependent. Discussing the recent results given in the literature, it appears that the defects created could be V_{As} and I_{As}.

When spectroscopic techniques are not applicable it is difficult to identify a defect and even, in a compound semiconductor, to determine the sublattice to which it belongs. In the case of irradiation-induced defects, two indirect ways can be used to make this identification possible. The first one consists in studying the anisotropy of the defect introduction rate R_d; by properly choosing the orientation of the crystal in relation to the irradiating beam (electrons) as compared to the crystallographic directions, the displacement of atoms of a given sublattice can be made more probable than for the other sublattice. This technique has been used in InSb (Ref. 1) and GaAs.[2,3,4] Recently the interpretation of results given in Ref. 4 has been confirmed for ZnSe.[5] The second one consists in using an alloy, i.e., a compound in which a third atomic component has been partly substituted to the atoms of one sublattice: for instance, a fraction x of Al atoms is substituted to Ga atoms in GaAs to form the $Ga_{1-x}Al_xAs$ alloy. Since the magnitude of the forbidden gap varies with the atomic composition x, it is in principle possible to deduce, from the variation with x of the position of the defect level, the nature of the defect, because the electronic structure of a given defect is made with a particular combination of wave functions built from wave functions of the valence and conduction bands.[3,6] Unfortunately, the theory is not yet able to provide unambigously the nature of the wave functions associated with a defect; moreover, the defect energy level varies with the nature of its atomic surrounding, i.e., with x. As a result, this technique cannot be used, as illustrated by the wrong answer it provided in GaAs.[3]

We propose in the present work another way of using the alloy by studying the defect introduction rate R_d of the induced defects as a function of x. If, in $Ga_{1-x}Al_xAs$, the defect results from arsenic atom displacement, its introduction rate should be relatively independent of x. If, on the other hand, the defect results from metal atom displacements, a dependence upon x is expected.

It is well known that electron irradiation induces five electron traps, labeled E_1 to E_5 (Ref. 7) and two hole traps (H_0,H_1). The electron traps have been extensively studied: their introduction rate as a function of the electron energy,[8] their annealing behavior,[9–11] and their electronic position in the gap as a function of x are known.[3,6] They have been shown to be associated with displacement in the As sublattice,[4] with a threshold energy consistent with a single-atom displacement. Discussing all these results and using the arguments that (1) the total defect introduction rate is practically equal to the calculated one, (2) the annealing follows first-order kinetics, (3) the annealed fraction is nearly 100%, and (4) no long-range diffusion is detected during annealing, Pons and Bourgoin[12] deduced that the traps E_1 to E_3 are associated with close pairs of primary defects of the As sublattice. They can be vacancy-interstitial pairs as in ZnSe (Ref. 13), but other types of pairs such as antisite As_{Ga}-V_{As} are possible.

The present study has been performed (1) to confirm the result[4] that E_1–E_3 belong to the As sublattice, and (2) to bring additional information in order to allow defects identification. The samples used are Sn-doped $Ga_{1-x}Al_xAs$ ($0 < x < 0.47$), grown by liquid-phase epitaxial technique on ⟨100⟩ GaAs substrates (10^{17} cm^{-3}, Si-doped, Bridgmann grown). The n-GaAs is grown on a n^+-GaAs substrate by vapor-phase epitaxy. Ohmic contacts are made on the substrate and a Schottky barrier is made by Au evaporation (thickness, 300 Å). For the irradiated samples, the Au deposition is performed after the irradiation in order to prevent a possible diffusion during this irradiation. The concentration and energy position of the traps is determined using the deep-level transient spectroscopy technique in the range 4 to 400 K. Electron irradiation is performed at room temperature with a Van de Graaff accelerator. The

TABLE I. Initial carrier concentration and native traps present in the samples before electron irradiation ($E = E_C - E_T$), trap position with respect to the conduction band).

Parameters (cm^{-3}) \ x	0	0.14	0.25	0.47
Initial carrier concentration	7.4×10^{15}	7×10^{16}	2×10^{16}	2.5×10^{16}
Native traps E(eV), concentration	0	0.37, 8.3×10^{14}	0.12, 1.4×10^{15} 0.25, 8.5×10^{14} 0.44, 8×10^{14}	0.17, 8×10^{15} 0.25, 4.2×10^{15}

beam is scanned over the various samples of variable x in order to insure that they received the same dose. Two doses, 10^{15} and 5×10^{15} e cm^{-2}, of 1-MeV electrons have been used. The accuracy in the measurement of the defect concentration, i.e., on the introduction rates, varies with x. Indeed, when x increases, the concentration of the native traps (see Table I) increases sharply and the concentration of the electron-induced traps becomes more and more difficult to determine. As shown in Table II, the position of the trap levels we obtained as a function of x is in reasonable agreement with the results of Lang, Logan and Kimerling[3] and Kravchencko and Prints.[6] The introduction rates obtained are given in Table III. For $x = 0$, the results are in reasonable agreement with the literature for 1-MeV electron irradiation.[8,14,15] The mean R_d value measured is 2 cm^{-1} for E_1, E_2, and 0.7 cm^{-1} for E_3. The apparent deviation of R_d we observe with x around this mean value is always smaller than the variation of x itself and we can therefore conclude that there is practically no variation of R_d with x for E_1, E_2, and E_3.

If we look at the primary intrinsic defects that can be created in gallium arsenide, we find (i) V_{As} and V_{Ga} (arsenic and gallium vacancy), (ii) I_{As} and I_{Ga} (arsenic and gallium interstitial); and (iii) As_{Ga} and Ga_{As} (antisite defects). In gallium aluminum arsenide, we find in addition to those defects cited above, (iv) V_{Al}, I_{Al}, As_{Al}, Al_{As}, and Al_{Ga}, Ga_{Al}. Al_{Ga} is not a defect but a substitution; we cannot separate between the following species: $As_{Al} \equiv As_{Ga}$, $Ga_{Al} \equiv Ga$, or $V_{Ga} \equiv V_{Al}$. We are then left for the GaAlAs material with two additional defects, I_{Al} and Al_{As}. The defects dependent on x(Al) are the following: I_{Al} and Al_{As}, proportional to x, and I_{Ga} and Ga_{As}, proportional to $1-x$. $V_{Ga} + V_{Al}$ is certainly dependent on x: the displacement cross section for an atom of mass M and atomic number Z is, in the Rutherford approximation and for electron energy, well above the threshold for atom displacement (our case), proportional to Z^2/M.[16] So $V_{Ga} + V_{Al}$ would be proportional to $1-0.6x$. ($|Z^2/M|Ga/|Z^2/M|Al \simeq 2.5$.) In our case we have shown that E_1, E_2, and E_3 traps are not x dependent and that they are present in pure GaAs as well as in GaAlAs. We can then conclude that they are related to the following simple intrinsic defects: V_{As}, I_{As}, and As_{Ga}, confirming the results of Pons and Bourgoin that the defects are associated with displacment in the As sublattice. After neutron and electron irradiation, more recently[17,18] the defect associated with As_{Ga} has been identified by ESR and can also be

TABLE II. Energy-level position ($E_C - E_T$ in eV) from the conduction band for the E_1, E_2, and E_3 traps vs x, the aluminum content. The results of Kravchenko and Prints (Ref. 6, column A) of Lang, Logan, and Kimerling (Ref. 3, column B) are compared with the present results (column C).

Traps \ Results	$x=0$			0.14			0.25			0.47		
	A	B	C	A	B	C	A	B	C	A	B	C
E_1	· · ·	0.08	0.04 ±0.02	· · ·	0.14	0.12	· · ·	0.17	0.17	· · ·	· · ·	0.30
E_2	· · ·	0.18	0.18	· · ·	0.23	0.20	· · ·	0.26	0.34	· · ·	· · ·	0.43
E_3	0.31	0.33	0.33	0.44	0.48	0.49	0.58	0.62	0.68	0.75	· · ·	0.85

TABLE III. Introduction rates R_d (cm^{-1}) in $Ga_{1-x}Al_xAs$ as a function of x, measured for two doses of irradiation (a: 5×10^{15} e cm^{-2}, and b: 10^{15} e cm^{-2}).

Traps		$x=0$	0.14	0.25	0.47
E_1	a	· · ·	2.26	2.10	1.5
	b	1.8	2.14	2.10	· · ·
E_2	a	· · ·	2.25	1.82	2
	b	1.8	2.10	1.88	· · ·
E_3	a	0.70	0.70	0.68	0.73
	b	0.70	· · ·	0.73	0.73

eliminated because its annealing occurs only between 450 and 500 °C far from the 200 °C annealing step of (E_1-E_5,H_1). Then we are left in presence of two candidates (V_{As} and I_{As}) for many traps.

If the assumption related to the primary intrinsic defect nature of the traps created during electron irradiation is verified, we are then led to say that one defect can give rise to several energy levels in the band gap, which is not impossible. This can be the case for E_1, and E_2 traps, in agreement with the recent Pons proposal.[12] Aother hypothesis will be different microscopic configurations of Frenkel pairs as observed in electron-irradiated ZnSe.[13]

ACKNOWLEDGMENTS

The authors are grateful to M. Gavand and A. Laugier of the Laboratoire de Physique de la Matière, Institut National des Sciences Appliquées for providing the liquid-phase epitaxial samples. This work is supported by the Délégation Générale à la Recherche Scientifique et Technique Contract No. 79.0724.

[1]F. H. Eisen, Phys. Rev. 135, A1394 (1964).
[2]G. W. Arnold and G. Gobeli, in *Radiation Effects in Semiconductors,* edited by F. L. Vook (Plenum, New York, 1968), p. 435.
[3]D. V. Lang, R. A. Logan, and L. C. Kimerling, Phys. Rev. B 15, 4874 (1977).
[4]D. Pons, and J. Bourgoin, Phys. Rev. Lett. 47, 1293 (1981).
[5]G. D. Watkins, Phys. Rev. Lett. 48, 568 (1982).
[6]A. F. Kravchenko and V. Y. Prints, Fiz. Tekh. Poluprovodn. 12, 1612 (1978) [Sov. Phys. Semicond. 12, 952 (1978)].
[7]D. V. Lang and L. C. Kimerling, in *Lattice Defect in Semiconductors—1974,* IOP Conf. No. 23 (IOP, Bristol and London, 1975), p. 581.
[8]D. Pons, P. M. Mooney, and J. Bourgoin, J. Appl. Phys. 51, 2038 (1980).
[9]D. V. Lang, in *Radiation Effects in Semiconductors—1976,* IOP Conf. No. 31 (IOP, Bristol and London, 1977), p. 70.
[10]D. V. Lang, L. C. Kimerling, and S. Y. Leung, J. Appl. Phys. 47, 3587 (1976).
[11]D. Pons, A. Mircea, and J. Bourgoin, J. Appl. Phys. 51, 4150 (1980).
[12]D. Pons, and J. Bourgoin, in the Third Lund International Conference on Deep Level Impurities in Semiconductors, Southbury, Conn., May, 1981 (unpublished).
[13]G. D. Watkins, Phys. Rev. Lett. 33, 223 (1974).
[14]J. L. Pinsard, and A. Zylbersztein, in *GaAs and Related Compounds—1980,* IOP Conf. No. 56 (IOP, Bristol and London, 1981), p. 435.
[15]A. H. Kalma and R. A. Berger, IEEE Trans, Nucl. Sci. 19, 209 (1972).
[16]C. G. Darwin, Philos. Mag. 25, 301 (1913).
[17]R. Worner, U. Kaufmann, and J. Schneider, Appl. Phys. Lett. 40, 141 (1982).
[18]N. K. Goswani, R. C. Newman, J. E. Whitehouse, Solid State Commun. 40, 473 (1981).