

# Some radiation effects in II-VI quantum dots

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The effect of electron irradiation ( $E=0.6-1.8$  MeV) on the optical characteristics (photoluminescence, PL) of CdTe/ZnTe structures with quantum dots, QDs, was investigated in the temperature range from 4.2 to 250 K. The data on the influence of irradiation on the temperature dependence of PL intensity, energy position and PL line width,  $W$ , from QDs were obtained. The narrowing of PL band and the blue shift of the QD peak position are explained by quenching of the low energy component connected with larger QDs. A slight decrease in PL intensity for both QDs and the buffer ZnTe layer as well as a sharp drop in PL intensity for 1 ML CdTe quantum wells, QWs, and fragments accompanied by a change in the activation energy of the PL quenching is explained by the radiation defects localized near the QD interface. The obtained results show that under e-beam irradiation QDs are more stable than QWs, which is in agreement with our previous investigations.

**Keywords:** CdTe/ZnTe quantum dots, photoluminescence, atomic force microscopy, electron beam irradiation.

## Introduction

Self-organized quantum dots, QDs, are of great interest for fundamental physics, as they are related to the theoretically predicted discrete atom-like energy levels,  $g$ -shape density of states and high excitonic oscillator strength due to extremely localized exciton wave function (comparable to the excitonic Bohr radius) [1]. The growth of semiconductor QDs is an object of special interest for application in new optoelectronic and low-dimensional photonic devices (single-electron transistors, optical memory structures, highly efficient semiconductor lasers etc.). Much interest was focused on the growth of the self-organized QDs on the basis

of III-V semiconductors (strained InAs layers on GaAs substrates [2] mainly) by well-established molecular-beam epitaxial, MBE, technique [3]. However, but only several groups reported the fabrication of QDs on the basis of II-VI materials with good optical quality: CdSe/ZnSe [4], ZnCdSe/ZnSe [5], CdTe/ZnTe [6-7]. It can be explained by larger ionicity and weaker surface atom diffusion that hampered the growth of these materials.

Recently, in [8-9] it was suggested to form QDs in ultrathin CdTe/ZnTe quantum wells, QWs, with high quantum efficiency in a broad range of visible spectrum. The obtained quantum-dot structures demonstrated intensive photoluminescence (PL) at low temperatures, but rapid PL quenching at temperatures about 150 K, suggesting the existence of defect-related states either in the CdTe dots or in the vicinity of interfaces.

Using the above technique for formation of QDs structures from ultra thin CdTe/ZnTe QWs, we obtained the experimental data on investigation of the optical characteristics (PL and reflectivity) of the samples with ultrathin CdTe/ZnTe layers [10]. These characteristics, including PL temperature dependence in the temperature range from 4.2 K to 250 K were studied before and after post growth treatments by high energy electron irradiation. Such irradiation ( $E=0.6-1.8$  MeV) is known to result in a controlled generation of point defects and electron-hole pairs. These treatments were used to reveal the mechanisms controlling the PL quenching processes caused by relaxation of carrier and the origin of the defect states in these structures. On the other hand, we tried to change the optical properties of these structures in a controlled way, an analogy to the post-growth rapid thermal treatments, recently used to modify the structural and optical properties of the InAs/GaAs QDs [11-13]. Investigation of such irradiation effects on QD characteristics of QDs may be useful for modification of their optical properties and elucidation of degradation mechanism in II-VI based devices.

## Experimental technique

The structures with ultra thin ZnTe layers with the average layer thickness ranging from 1 to 6 monolayers (ML) were grown by elemental source molecular beam epitaxy, MBE, (I type) and 1 ML layers were obtained by atomic layer epitaxy ALE (II type). Both types of layers were grown on (001) GaAs semiinsulating substrates. After oxide desorption GaAs substrates were cooled from 580 °C down to room temperature and covered by an amorphous to room temperature and covered by an amorphous 5 nm thick ZnTe layer, which was then recrystallized by heating up to the growth temperature. The subsequent procedure was the growth of a 1.5  $\mu\text{m}$  thick ZnTe buffer layer. The growth rate was  $\approx 0.6$  nm/h. The growth temperatures were  $\approx 350^\circ\text{C}$  for ZnTe and  $\approx 300^\circ\text{C}$  for CdTe layers [14-15]. Next the growth order was an ultrathin CdTe layers about (1-6) MLs thick with ZnTe barriers. The sample parameters are presented in the Table.

PL was excited by 488.0 nm and 514.5 nm line of  $\text{Ar}^+$  ion laser and detected with a photomultiplier tube in the current-flow regime using the grating spectrometer MDR-23. Surface morphologies were investigated by atomic force microscopy,

AFM (Figure1). ZnTe based structures were irradiated with 1.8 MeV energy electrons. The electron irradiation was performed at a beam density of  $1 \text{ mA} \cdot \text{cm}^{-2}$  up to a dose of  $4 \cdot 10^{16} \text{ sm}^{-3}$  (the flux was  $\approx 6 \cdot 10^{12} \text{ sm}^{-2} \cdot \text{s}^{-1}$ ) using an electron accelerator. The electron irradiation created Frenkel defects and generated e-h pairs in ZnTe.

Table 1.

Parameters of the samples under study.

N	Sample	Buffer ZnTe layer, $\mu\text{m}$	CdTe layer, ML	Over layer, nm	CdTe (next) layer, MI.	Cap layer, nm	$\lambda_{pL}$ , nm ( $\lambda_{exc}=514.5$ nm T=77 K)	Method of the growth
1	2	3	4	5	6	7	8	9
1	145	1.5	1	20	(<1ML)x 2	40	$\approx 543$	MBE
2	142	1.5	4	20	1	40	$\approx 623$	MBE
3	158	1.5	1			40	$\approx 555$ $\approx 570$ $\approx 580$	ALE
4	198	1.5	4			30	$\approx 620$	ALE
5	198	1.5	6			30	$\approx 640$	ALE

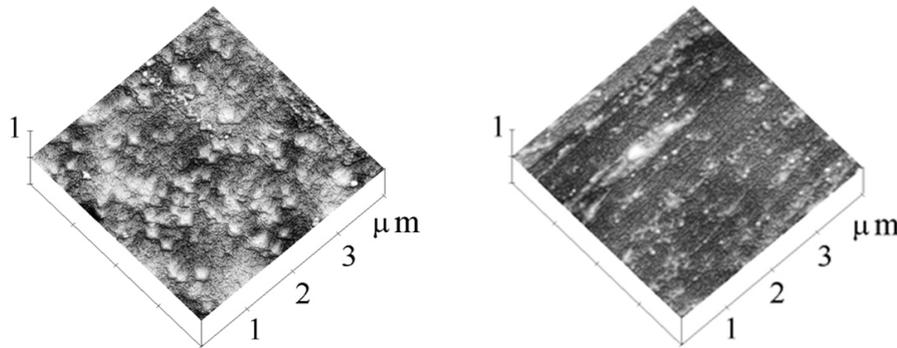


Figure 1. AFM images for surface of as-grown and electron irradiated sample with CdTe/ZnTe QDs grown on (100) GaAs substrate.

## Results and discussion

Figure 2 shows the typical PL spectra for as-grown CdTe ultrathin layers with thickness from about 1 ML ( $\approx 3\text{E}$ ) to 4 MLs ( $\approx 12\text{E}$ ) [7]. These spectra exhibit two lines in the excitonic region of ZnTe buffer layer [14]: line  $I_{FX}^{hh}$  with energy position 2.369 eV (2.367 eV) and  $I_{FX}^{lh} = I_2^{Ga} = 2.3601$  eV (2.359 eV) and accompanied by a strong PL line formed by quantum size CdTe structures which can be attributed to fragments of CdTe 1ML QWs or larger CdTe QDs with 2.2878 eV (1.990) samples with 1 MLs, respectively. These spectra have two lines in the excitonic region of the ZnTe buffer layer spectra [14]: the line of the high-hole component of free exciton ( $I_{FX}^{hh}$ ) as well as the line of exciton bound to neutral donor (probably Ga [16])

which overlaps with the light-hole component of free exciton ( $I_{FX}^{lh} = I_2^{Ga}$ ). The red shift of the excitonic lines of the ZnTe-related emission (in comparison with bulk ZnTe) is related to the presence of residual and inhomogeneous strain in the ZnTe buffer layer. The main origin of the residual strains is the lattice mismatch between the ZnTe epilayer and the GaAs substrate, but this shift is different from those observed in ZnTe/GaAs hetero structures with a 1.5 nm thick ZnTe [17]. Additional strains may occur when the ZnTe epilayers are covered by the CdTe ultrathin layers without sufficient strain relaxation. Figure 1 demonstrates a significant red shift and broadening of QDPL line for different values of CdTe coverage. In the structure (curves a) containing 1 ML CdTe ( $\langle L_z \rangle = 1$  ML) we can see an intensive PL line from QDs with a small spectral line width  $W \approx 3$  meV and a larger spectral width ( $W \approx 60$ ) meV for structures (curve b) containing 4 ML CdTe ( $\langle L_z \rangle = 4$  ML). It is known that both size and composition may be responsible for the QD spectra evolution. The PL intensity increased with the amount of deposited CdTe and reached its maximum at about 4 ML. The PL energy showed a minimum when  $\approx 1$  ML CdTe was deposited (in our experimental conditions). In our work we also studied the influence of irradiation on the properties of larger dots, it is connected with their higher PL intensity and better experimental possibilities for measurements. The temperature dependence of such QDs has not been fully explained now completely. Figure 3 shows the PL spectra from larger QDs. It is clear that the PL dependence on QD sizes distribution is lateral non uniform.

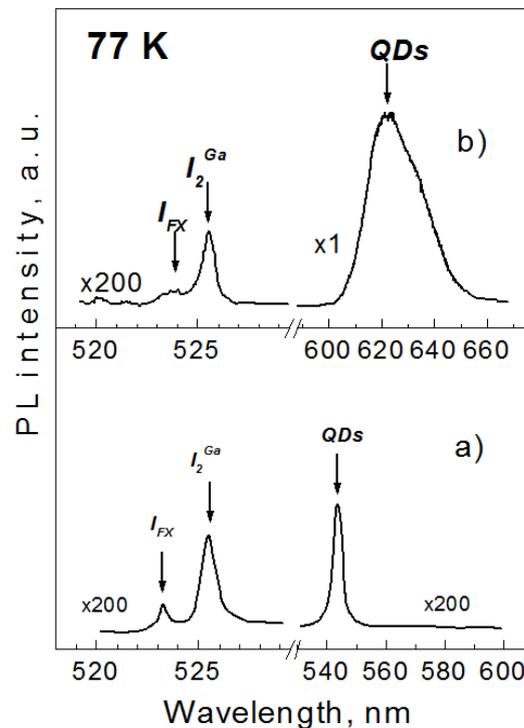


Figure 2. Photoluminescence spectra of ultrathin CdTe/ZnTe QWs with a well width 1 ML (a) and 4 ML (b). The spectra were obtained at 77 K, with  $\lambda_{exc} = 488$  nm,  $P_{exc} = 0.3$  W/cm<sup>2</sup>.

The size distribution of as-grown QDs was mainly bimodal (that suggests dot formation of two predominant sizes) but for some specimens a single mode distribution was obtained. For each temperature the QD PL spectra for bimodal distribution were fitted by two Gaussians. The multimodal size distribution of

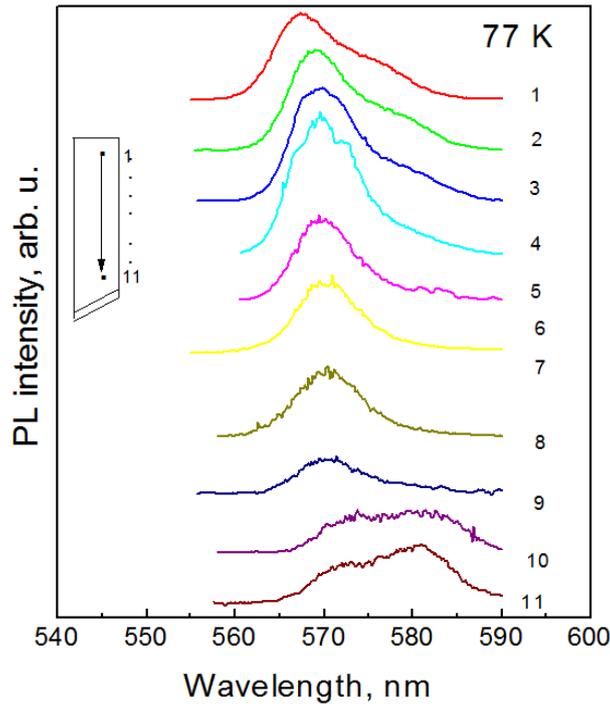


Figure 3. Position dependent PL spectra from the sample with CdTe/ZnTe QDs. The inset shows a sample with the arrow indicating the scanning direction.  $\lambda_{exc} = 514.5$  nm,  $P_{exc} = 2$  W/cm<sup>2</sup> (curves 1-7) and 12 W/cm<sup>2</sup> (curves 8-11).

CdTe QDs was qualitatively confirmed by AFM measurements. A detailed study of the 4.2 K PL spectra from quantum-size structures with different CdTe coverage  $L$  (see Figure 4a) confirmed the presence of the PL line connected with “scrap” of atomic plains in the excitonic region of the spectra too. Temperature dependencies of PL intensity,  $I_{PL}$ , emission energy,  $E_{PL}$ , and full width half-maximum, FWHM, from QDs were studied (see Figure 4b). FWHM remained almost unchanged with temperature increase until the critical value of temperature value  $T_C$  was reached. Below the  $T_C$  the QD PL intensity decreased considerably. Up to  $\approx 130$  K  $E_{PL}$  was almost invariable and monotonically shifted to the low energy region ( $\approx 10$  meV) in the range from 140 to 200 K with the rate lower than the ZnTe band gap change, which meant the lateral exciton localisation in our samples. Irradiation led to a decrease in intensity and spectral shift of PL bands connected both with QWs and QDs as well as with buffer ZnTe layers. It was shown that the red shift of the QW PL band is conditioned by the relaxation of the radiation enhanced stress and blue one is due to QWs disordering in the result of Cd/Zn interdiffusion. In addition to changes in the PL band line width and a decrease in its intensity, irradiated QDs had a shift to the short wavelength region. Both AFM and optical measurements indicated the changes in the bimodal QD shape. The change in the temperature dependence of the PL intensity, position and line width in temperature range from 4.2 to 250 K obtained from irradiated samples confirmed the change in the above considered above the mechanism of thermal quenching of QD radiative transition from. We explained such changes by the formation of radiation enhanced defects creation near QD interfaces with additional levels on the interface.

Thus, the following changes were observed after irradiation:

(I) small decrease of PL intensity from both QDs and buffer ZnTe layer and significant drop of PL intensity of 1 ML CdTe quantum well, QW, fragment;

(II) blue shift of QD peak position;  
 (III) change of QD PL line shape from bimodal to single mode resulted in line narrowing. Such effect was caused by quenching of low energy component (QDs of larger size);

(IV) absence of radiation enhanced strain relaxation in ZnTe buffer layer contrary to heterostructures containing QWs.

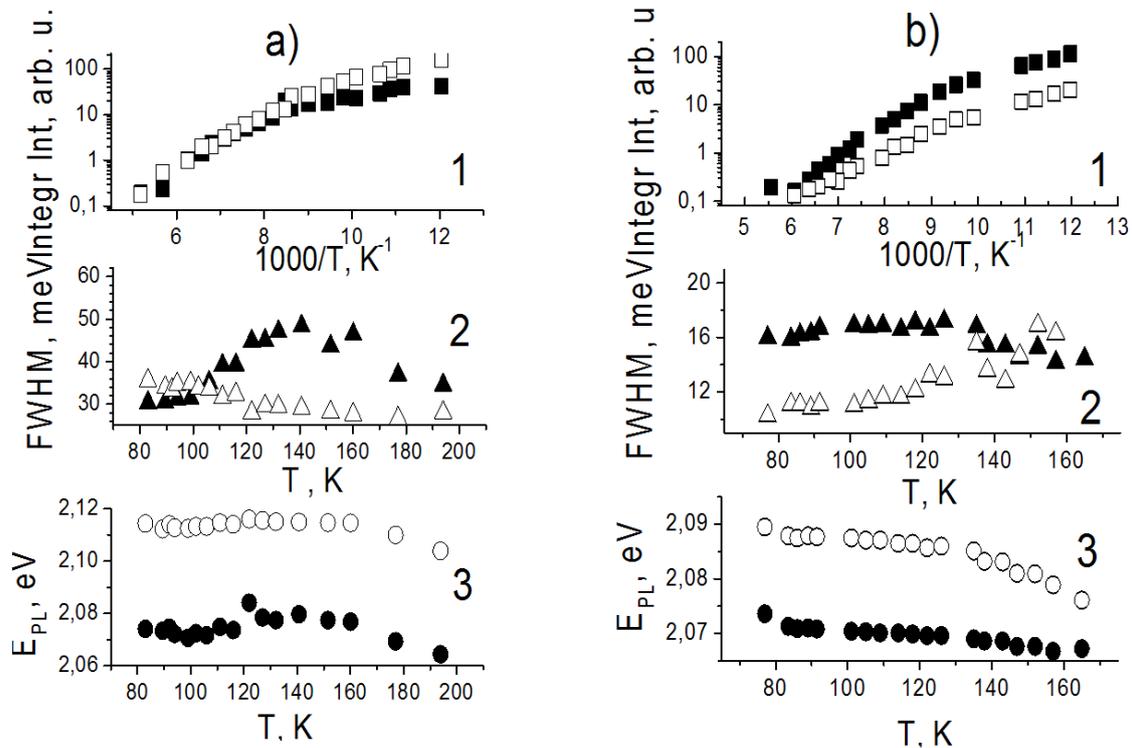


Figure 4. Temperature dependence of band parameters of as-grown QDs (a) and QDs after  $\gamma$ -irradiation (b). Open and solid circles correspond to two components of PL band deconvolution. (1)-PL integral intensity (2)-line width FWHM, (3)-peak position EPL.

## Conclusion

The obtained results show that under e-beam irradiation QDs are more stable than 1 ML QWs, which is in agreement with our previous investigations. Changes in QD PL spectra after irradiation are explained by predominant spreading of larger QDs. The reasons for differences in the influence of e-beam irradiation on the optical properties of QD and QW containing structures are discussed.

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