Experimental study of the size-dependent photoluminescence emission of CBD-grown PbSe nanocrystals on glass

Tahere Hemati © and Binbin Weng ©
School of Electrical and Computer Engineering, University of Oklahoma, Norman, Oklahoma 73019, United States of America
E-mail: binbinweng@ou.edu

Keywords: nanocrystals (NCs), photoluminescence (PL), chemical bath deposition (CBD), lead chalcogenide, PbSe, mid-infrared (MIR)

Supplementary material for this article is available online

Abstract

In this work, we study the size-dependent properties of Photoluminescence (PL) emissions of PbSe Nanocrystals (NCs) grown by Chemical Bath Deposition (CBD) method. In previous studies, PL emissions have been tuned by CBD-grown PbSe, and the growth mechanism was dependent on crystalized substrates such as GaAs. In this research, however, PL emissions are controlled over the mid-infrared (MIR) range, through PbSe NCs, which are deposited on glass as an amorphous material. This study proposes an alternative approach to control PL emissions, which provides us with more freedom to fabricate low-cost MIR light sources as crucial components in remote sensing and gas analysis. Moreover, in this study, the advantage of the post-thermal method to control the NCs size, compared to the growth temperature, is shown.

Introduction

Lead chalcogenides nanocrystals [1–3], with a large exciton Bohr radius [4], are of high interest for optoelectronic applications, such as light-emitting diodes [5], photodetectors [6], and photovoltaic devices [7, 8]. This interest arises mostly from the strong quantum confinement, and the broad size-tunability of the bandgap energies [9, 10]. Therefore, photoluminescence emissions of lead chalcogenides NCs can cover the near-infrared region entirely, and specifically lead selenide (PbSe) NCs cover PL emissions over the mid-infrared wavelengths [11, 12]. Efficient mid-infrared devices play an essential role in a variety of applications, including remote detection, environmental sensing, and gas analysis [13–15].

Colloidal PbSe NCs [16, 17], are well-known for providing a high tunability over the infrared range. However, the sensitivity of the colloidal-based sensors can be affected by environmental conditions, such as light and moisture [18]. Moreover, one of the challenges of colloidal NCs is their ability to couple onto a substrate [19–22]. Colloidal NCs that are synthesized in the solution possess an organic shell that acts as a protective layer between electrical charges in NCs and the environment. The presence of this shell prevents an effective coupling of colloidal NCs to a substrate [19]. Furthermore, the coupling of colloidal NCs onto a substrate may lead to the interaction between the NCs and the environment, and therefore affects the properties of the NCs on the substrate and make them different from the NCs in the solution [20, 21]. Uniform dispersion of NCs onto the substrates is another concern in the colloidal method [22].

Nevertheless, in non-colloidal methods, such as CBD, NCs grow directly from the substrate. Thus, all challenges of the coupling of NCs onto the substrate are not a concern anymore. Furthermore, the simplicity and cost-efficiency are augmented advantages of the CBD method [23–25]. Recently, Templeman et al [26] have provided a broad tunability over the MIR range by using the CBD method. They explored the oriented attachment growth mechanism where NCs can only grow from an oriented substrate such as GaAs. However, in this work, we show that in a cluster regime, and from a un-crystalized surface such as glass, through controlling the heat-treatment temperature, PL-tunability over the same range (MIR) is achievable.
In this work, instead of using crystalized substrates, PbSe NCs are synthesized on glass as a ubiquitous and inexpensive material, and PL emissions are tuned over the MIR range by using different heat-treatment temperatures. Furthermore, to indicate the efficiency of the heat-treatment method in tuning the PL emissions, the growth temperature is studied as another effective factor in controlling PL emissions.

**Experimental section**

**Materials and chemicals**
Sodium sulfite (Aldrich, BioUltra, anhydrous, ≥98% (RT)), selenium powder (Aldrich, 100 mesh, 99.99% trace metals basis), lead (II) acetate trihydrate (Aldrich, ≥99.99% trace metals basis), and sodium hydroxide (pellets EMPLURA) were used without further purification. A stock of sodium selenosulfate (Na₂SeSO₃), 0.2 M, was prepared by refluxing 0.2 M of selenium powder with 0.5 M of sodium sulfite (stirring at 90 °C for 6 h). The sodium selenosulfate solution was filtered in order to remove nonreacted selenium powder and was stored at 4 °C. Rectangular microscope slides (3 × 1") were used as substrates, cleaned by solvent (acetone 55 °C, methanol 25 °C) and RCA bath (a mixture of 5 parts DI water, 1 part ammonia hydroxide and 1 part hydrogen peroxide, 70 °C/15 min). In order to prevent large particles from adhering to the growing film, films were deposited on the bottom face of the substrates at an angle of ~70° with respect to the air–solution interface (figure 1). The deposition solution was prepared by diluting the 0.2 M lead acetate solution with water and adjusting the pH with NaOH (5 M) to ~11.7. Finally, 48 mM of Na₂SeSO₃ was added to a 50 ml Pyrex beaker.

**Structural, chemical, and optical characterization**

**X-ray diffraction (XRD)**
The crystallographic phase and texture of the films were studied by Rigaku Ultima IV diffractometer. Data were collected in the 2θ/θ geometry using Cu Kα radiation (λ = 1.5405 Å) at 40 kV and 44 mA. Diffraction scans were taken during 120 min over 2θ range of 10° – 70° with a step size of ~0.2°.

**Field emission scanning electron microscopy (FE-SEM)**
The morphology of the films was observed using an ultrahigh-resolution Ziess Neon EsB. The acceleration voltage was 15 kV.
Photoluminescence measurement

PL emissions were measured by a Bruker Invenio-R Fourier transform infrared (FTIR) spectrometer in the MIR range with a 16 cm⁻¹ resolution. InSb high gain detector was cooled by liquid nitrogen. To decrease the thermal background, the PL module was combined with an external SR-830 lock-in amplifier. Excitation was achieved by a 980 nm laser.

Results and discussion

Figure 2(a) shows Scanning Electron Microscopy (SEM) images of the PbSe NCs, grown at 40 °C for 2 h without any post-treatment, and figure 2(b) shows SEM image of this sample under 300 °C heat-treatment for 5 min. In these figures, the scale bars are 200 nm. From these images, it is apparent that in the as-grown sample (figure 2(a)), small grains are connected and make larger clusters. Hence, there is no clear border between these small grains. However, in figure 2(b), grains are isolated, and the border of grains are distinguishable.

Figure 3 shows x-ray diffraction (XRD) patterns of the as-grown and the heat-treated samples. Orthorhombic PbSe is indicated by the gray columns, the cubic phase is shown by the pink columns, and Pb(HSeO₃)₂ is displayed by the hatched column.

Figure 2. (a). An SEM image of a thin layer of PbSe without any heat-treatment (b). An SEM image of this sample after 5 min of heat-treatment (300 °C). Scale bars are 200 nm. Insets are the magnified SEM image of the as-grown and the heat-treated sample.

Figure 3. X-ray diffraction patterns of the as-grown (red) and the heat-treated (blue) samples. Orthorhombic PbSe is indicated by the gray columns, the cubic phase is shown by the pink columns, and Pb(HSeO₃)₂ is displayed by the hatched column.
According to the ICCD, three peaks at 38.3°, 44.26°, and 65.72° attribute to (102), (202), and (222) phases of orthorhombic PbSe, respectively. Moreover, peaks that are located at 29.31°, and 42.2° have a good match with (200) and (220) faces of cubic crystal of PbSe in the ICCD. These peaks are depicted by the pink columns in figure 3. The peak located at 21.6° matches with Pb(HSeO₃)₂ in the ICCD. Pb(HSeO₃)₂ peak is displayed by the hatched column.

Given the fact that the growing procedure and post-thermal process are performed at a non-vacuumed ambient, there is a possibility for the penetration of Oxygen and Hydrogen during the heat-treatment and the growing procedure. Thus, emerging of a Pb(HSeO₃)₂ peak can be attributed to the Oxygen and Hydrogen penetration. Moreover, emerging of a cubic PbSe after heat-treating can show how the orthorhombic phase, which is a less symmetric and ordered structure, has changed to a more symmetric cubic phase. The transition of the PbSe NCs from the orthorhombic phase to the cubic can be interpreted via the thermodynamic models. The most stable phase of any crystal has minimum free energy \[27\]. Besides, the minimum enthalpy and the maximum entropy are in favor of minimizing free energy (\(G(P,T) = H-TS\)). Compared to high-symmetric structures, like the cubic phase, less-symmetric structures (orthorhombic) show higher entropy, and therefore lower free energy \[28\]. However, due to the temperature dependency of enthalpy and entropy, the interpretation of the phase transition is not straightforward. However, the impact of enthalpy is investigated by GGA and LDA approximations in \[27\].

A comprehensive study on the phase diagram of cubic and orthorhombic PbSe is performed in \[29\]. In essence, thermodynamic laws suggest that, in our case, the entropy and enthalpy changes in a way that the minimum free energy is acquired by synthesizing orthorhombic PbSe at lower temperatures. Likewise, in \[30\] cubic CsSnI₃ is reported as a high-temperature phase while the orthorhombic is a low-temperature one. The difference in structures is manifested in the rather large enthalpy and entropy change associated with the transition \[30\].

Besides, the orthorhombic PbSe phases, which are even visible in the as-grown sample spectrum, became sharper after the heat-treatment, which can be interpreted as an improvement in the crystal quality. Figure 4 shows the PL spectra of the sample (40 °C/2 h) before and after heat-treatment (300 °C.5 min). The red dashed line is the PL spectrum of the as-grown sample, and the blue solid line is the PL spectrum of the heat-treated sample. According to figure 4, there is a broadband PL emission with a peak position at 3.75 μm before the heat-treatment. However, after heat-treating, we observed a significant blueshift (∼950 nm) in the peak position of the PL. In the following paragraph, the reason for this blueshift will be investigated in detail, using SEM and AFM photos and XRD analysis.

In 2017, Yang et al \[31\] suggested that the unusual PL blueshift after the heat-treatment can be attributed to the phase transition from orthorhombic to tetragonal, which results in the renormalization of a bandgap. Furthermore, recently, in 2019, the other work investigated the effect of phase transition on the bandgap shift \[32\]. In our study, as it is shown in figure 3, there is a phase transition. Therefore, the given blueshift can be attributed to the bandgap reconfiguration, results in the phase transition.

![Figure 4. The PL spectra of the sample (40 °C/2 h) before and after heat treatment (300 °C.5 min).](image)
Another explanation for the observed blueshift can be related to the size difference between the coalescence NCs in the as-grown sample, and the separated NCs in the heat-treated sample. The coalescence NCs, which are distinct in SEM photos (figure 2), are larger than separated NCs expanded due to the heat-treatment. To support this claim, an Atomic Force Microscopy (AFM) image of the as-grown sample is shown in figure S.2(a) is available online at stacks.iop.org/NANOX/1/010030/mmedia.

According to the AFM analysis, the average size of the coalescence NCs is estimated to be around 28 nm. Moreover, to provide further quantitative information, the absorption-band shifting according to NCs size is plotted in figure S.2(b) [33]. As figure S.2 (see Supporting Information) shows, the absorption edge of NCs of ~28 nm size is around 3.76 μm. Considering the stokes shift, there is a good agreement between the peak position of the PL emission and the estimated coalescence NCs size.

These coalescence NCs are made of individual crystals in a size of around 10 nm. However, the size of the individual expanded NCs in the heat-treated sample, which are separated now after the heat-treatment, is assessed to be around 17 nm. Thus, the PL peak position of the heat-treated sample at 300 °C, originates from individual NCs with a size of 17 nm, while the PL emission of the as-grown sample comes from coalescence NCs with a size of 28 nm. Given the fact that smaller NCs emit PL at shorter wavelengths, and the coalescence NCs act as a unit, the observed blueshift can be justified.

Furthermore, the XRD analysis of the as-grown sample (figure 3) suggests the existence of the orthorhombic PbSe, which has been observed in other lead chalcogenides [34] as well. Orthorhombic crystals tend to emerge in lower temperatures, compared to cubical structures [35]. Thus, the initial temperature of Pb and Se precursors had been sufficient to create orthorhombic PbSe crystals. However, this temperature is not enough to create cubic PbSe.

The emitting of another PL-peak in bulky-range (3.75 μm) after heat-treating can be related to the small fraction of grains, which has not been isolated completely. Furthermore, the XRD analysis (figure 3) after the heat-treatment shows the transition of orthorhombic PbSe to the cubic phase, which occurs at higher temperatures. Since the heat-treatment has not been performed under vacuum, it is possible that this large blueshift is more related to the effect of Oxygen rather than the phase transition to the cubic structure and isolating grains.

Therefore, to find out the net impact of Oxygen on shifting the peak-position of the PL emission, a sample was heat-treated under the same condition, once in an Argon ambient and once in a mixture of Argon and Oxygen. Compared to the Argon ambient, no PL shifting was observed in the mixture of Argon and Oxygen ambient (see Supporting Information), indicating that Oxygen has no effective impact on the PL shifting.

Moreover, the impact of heat-treatment temperature on PbSe NCs is studied through tuning the temperature from 300 °C to 425 °C with the intervals of 15 °C. Figures 5(a)–(f) show SEM images of the sample under different heat-treatment temperatures. These images indicate how increasing the temperature leads to the enlarging of the grain size.

![Figure 5](image-url)
Moreover, figure 6(a) shows the PL spectra of the sample (40 °C/2 h) under different heat-treatment temperatures. This figure indicates how the PL emissions change under different temperatures. To make a better vision of the PL-shifting, the normalized PL emissions are shown in figure 6(b). Moreover, figure 6(c) quantitatively indicates the changes of the PL intensity and peak-position over 300 °C–420 °C. As figure 6(c) shows, increasing the temperature leads to the enhancement of the PL and a redshift. However, the slope of the shift in peak-positions starts to flatten at temperatures higher than 375 °C. It should be mentioned that at temperatures higher than 425 °C, PbSe thin film starts to evaporate.

The redshift of PL emissions, due to the increased temperatures, can be explained through the size-dependent effective bandgap energy of NCs equation, \[ E_{\text{nano}} = E_{\text{bulk}} + \left( \frac{\pi h^2}{2 \mu R^2} \right) \]

Figure 6. (a). The PL spectra of the sample (40 °C/2 h) under different heat-treatment temperatures (300–425 °C, 5 min). (b). The normalized PL spectra under different heat-treatment temperatures. (c). Peak position and PL intensity changes by manipulating heat treatment temperatures.

Figures 5 and 6 show that increasing the temperature leads to larger structures and the PL peak redshifts. However, in figure 4, the PL peak of the heat-treated sample, compared to the as-grown sample, blueshifts. In the first glance, these two trends seem opposite. However, as mentioned this unusual PL blueshift can be attributed to the phase transition from orthorhombic to cubic, or size-dependent energy bandgap of NCs. Thus, the size of coalescence NCs in the as-grown sample is larger than individual NCs in the heat-treated sample at 300 °C. However, the increment of the heat-treatment temperature enlarges the NCs size and redshifts the PL peak.
According to figures 5 and 6, smaller NCs show weaker PL emissions. One possible explanation for this phenomenon can be the surface to volume ratio [37]. Since surface defects have a crucial role in the PL intensity, and smaller particles possess a higher surface to volume ratio, it seems that the effect of un-passivated surface defects can be more significant in smaller NCs. These surface defects can trap the photo-generated charge carriers and alleviate the PL of NCs, which leads to weaker PL emission for smaller NCs [37].

Manipulating the heat-treatment temperature led to a broad range of PL tunability over the MIR. However, to explicate the efficiency of the post-thermal method, we need to investigate the growth temperature, as another effective factor in controlling PL emissions. To achieve this goal, two samples, grown under two different temperatures: 4 °C (160 min), and 40 °C (120 min), are studied. To enhance the PL emission, both samples are heat-treated at 375 °C (at temperatures higher than 375 °C, the PL-shifting is not significant, and at temperatures lower than 375 °C the PL intensity is weak (figure 6(c)).

Figure 7(a) displays the SEM image of the samples grown at 4 °C, and figure 7(b), is the SEM image of the samples grown at 40 °C. Scale bars are 200 nm. Insets show the AFM images of 4°-grown and 40°-grown samples. The RMS roughness of the 4°-grown sample is 3.2 nm, while this value for 40°-grown is 16.5 nm. Figure 7 suggests that lower growth temperature leads to synthesizing of smaller NCs, and smoother thin PbSe films.

Furthermore, to provide an insight into crystal structures and orientations of these samples, XRD analysis was performed. In figure 8, x-ray diffraction patterns of the 4°-grown sample are shown by blue color, and x-ray diffraction patterns of the 40°-grown sample are shown by red color. Orthorhombic PbSe is indicated by the gray columns, the cubic phase is shown by the pink columns, and Pb(HSeO₃)₂ is displayed by the hatched column.

As figure 8 indicates, both samples show the same crystal orientation. However, the 4°-grown sample shows more intense peaks at 44 and 64 degrees. In a perfect crystal, all the intensity is concentrated at the node of the reciprocal lattice. However, in an imperfect crystal, a fraction of the intensity is eliminated from the node and scattered in the unit cell [38]. Thus, the intensity can be an indicator of crystal quality. Therefore, more intense peaks, which, according to the ICCD, are related to orthorhombic structure, suggest that lower growth temperatures can lead to synthesizing of orthorhombic PbSe with higher crystal quality.

This higher crystal quality can be attributed to the growth velocity, which is lower in the 4°-grown sample [39]. By slowing the growth velocity, the possibility of emerging defects decreases, resulting in higher quality crystals [40]. Figure 9 displays the PL spectra of the samples grown at 4 °C and 40 °C. In this figure, the blue dashed line shows the PL spectrum of the 4°-grown samples, and the red solid line shows PL spectrum of the 40°-grown sample. As you can see, the reduction of the growth temperature leads to enhanced PL emissions. Besides, the 4°-grown sample shows ~130 nm blueshift. This shifting can be justified through the size-dependent effective bandgap energy of the NCs equation. According to this equation, larger NCs emit PLs at longer wavelengths. On the other hand, figure 7 indicates that a higher growth temperature leads to synthesizing of larger NCs. Thus, the 4°-grown sample, due to the smaller NCs, emits at shorter wavelengths.

Moreover, figures 9 and 7 suggest that lower growth temperatures lead to stronger PL signals. By comparing the roughness and the grain size (figure 7), we expect that the 40°-grown sample emits stronger PL signals due to the larger grains and higher roughness. However, to drive this conclusion, we did not consider two important...
criteria: thickness and crystal quality. The thickness of the 40°C-grown sample is 166 nm, while the thickness of the 4°C-grown sample is 190 nm. Therefore, the 4°C-grown sample is 24 nm thicker than the 40°C-grown sample. Since the PL intensity is linearly depended on the film thickness [41–43], a small portion of this PL enhancement is due to the difference in the film thickness.

XRD analysis (figure 8) displays that the highest peak of the 4°C-grown sample is four times more intense than the 40°C-grown sample. Due to the accumulation effect of equivalent crystal planes, or the multiplicity factor [44], the thickness increment of a sample leads to the enhancement of the XRD peak intensity. Thus, following the footsteps of previous works [45, 46], which are investigating the impact of the PbSe thickness on XRD peak intensity, we calculated that 24 nm increment in the thickness has enhanced the intensity of the highest peak around 1.4 times. However, the peak intensity of the highest peak in the 4°C-grown sample is 4 times higher than the 40°C-grown sample. Therefore, the thickness difference is not the only reason for the peak intensity difference between the 4°C-grown sample and the 40°C-grown sample. Moreover, crystal quality plays an important role. Thus, 1.4/4 (~35%) of the enhanced peak intensity is caused by the thickness difference between these two samples.
Conclusion

In this work, by using the CBD method, PbSe NCs were synthesized on glass as a less expensive material relative to the previously used crystallized substrates. The samples were heat-treated at 300°C, 325°C, 350°C, 375°C, 400°C, and 425°C for 5 min. We found that increasing heat-treatment temperature leads to a redshift of the PL spectra, resulted from the size-dependent effective bandgap energy. Moreover, we noted that higher growth temperatures showed weaker PL signals. Although it was expected that the 40°C-grown sample with higher roughness and larger grains indicates stronger signals, the 4°C-grown sample, due to its higher crystal quality and having a thicker layer of PbSe, showed stronger PL emissions. Overall, in this research, a broad range of roughness and larger grains indicates stronger signals, the size-dependent effective bandgap energy. Moreover, we noted that higher growth temperatures led to a redshift of the PL spectra.

References

[1] Nienhaus L, Wu M, Bulović V, Baldo M A and Bawendi M G 2018 Using lead chalcogenide nanocrystals as spin mixers: a perspective on near-infrared-to-visible upconversion. Dalton Trans. 47 8509–16
[2] Spoore F C, M, Grimaldi G, Delerue C, Evers W H, Crisp R W, Geiregat P, Hens Z, Houtepelen J A and Siebbeles L D A 2018 Asymmetric optical transitions determine the onset of carrier multiplication in lead chalcogenide quantum confined and bulk crystals ACS Nano 12 4796–802
[3] Oh S, J, Strauss D B, Zhao T, Choi J-H, Lee S-W, Gaulling E A, Murray C B and Kagan C R 2017 Engineering the surface chemistry of lead chalcogenide nanocrystal solids to enhance carrier mobility and lifetime in optoelectronic devices. Chem. Commun. 53 728–31
[4] Wise F W 2000 Lead salt quantum dots: the limit of strong quantum confinement. Acc. Chem. Res. 33 773–80
[5] Hu W, Henderson K, Zhang Y, You G, Wei L, Bai Y, Wang J and Xu X 2012 Near-infrared quantum dot light emitting diodes employing electron transport nanocrystals in a layered architecture Nanotechnology 23 375202
[6] Ren Z, Sun J, Li H, Mao P, Wei Y, Zhong X, Hu J, Yang S and Wang J 2017 Bivler PbS quantum dots for high-performance photodetectors. Adv. Mater. 29 1702055
[7] Carey G H, Abdelhady A L, Ning Z, Thon S M, Bakr O M and Sargent E H 2015 Colloidal quantum dot solar cells. Chem. Rev. 115 12732–63
[8] Böhm M L, Jellicoe T C, Tabachnyk M, Davis N J I K, Wünsivevsky-Rocca-Rivara F, Ducati C, Ehrler B, Bakulin A A and Greenham N C 2015 Lead telluride quantum dot solar cells displaying external quantum efficiencies exceeding 120% Nano Lett. 15 7987–93
[9] Gorser S, Albou-Yaron A and Hodes G 1995 Quantum size effects in chemically deposited, nanocrystalline lead selenide films J. Phys. Chem. 99 16442–8
[10] Wang X, Xi G, Liu Y and Qian Y 2008 Controllable synthesis of PbSe nanostructures and growth mechanisms Cryst. Growth Des. 8 1406–11
[11] Lu, H, Carroll G M, Neale R N and Beard M C 2019 Infrared quantum dots: progress, challenges, and opportunities ACS Nano 13 939–53
[12] Dolatari M, Rostami A, Mathur S and Klein A 2019 Trap engineering in solution processed PbSe quantum dots for high-speed MID-infrared photodetectors J. Mater. Chem. C 7 5658–69
[13] Hemati T and Weng B 2018 The mid-infrared photonic crystals for gas sensing applications. Adv. Mater. 30 1703973
[14] Pietryga J M, Schaller R D, Werder D, Stewart M H, Klimov V I and Hollingsworth J A 2004 Pushing the band gap envelope: mid-infrared enhanced quantum dot photodetectors J. Appl. Phys. 105 053101
[15] Pietryga J M, Schaller R D, Werder D, Stewart M H, Klimov V I and Hollingsworth J A 2004 Pushing the band gap envelope: mid-infrared enhanced quantum dot photodetectors J. Appl. Phys. 105 053101
[16] Veksler A, Wu M, Bulović V, Baldo M A and Bawendi M G 2013 Lead chalcogenide nanocrystals through cation-exchange-mediated nucleation Angew. Chem. Int. Ed. 47 3029–33
[17] Zhao H and Rosei F 2013 Infrared quantum dot photodetectors. Adv. Mater. 25 329–38
[18] Acharya K P, Khon E, O'Connor T, Nemitz I, Klinkova A, Anzenbacher P and Zamkov M 2011 Heteroepitaxial growth of PbSe nanocrystals onto substrate films via hot-injection routes. ACS Nano 5 4935–64
[19] Litvin A P, Ushakov E V, Parfenov P S, Fedorov A V and Baranov A V 2014 FRET between close-packed quasi-monodispersed PbSe QDs in a porous matrix J. Phys. Chem. C 118 6531–5
[20] Ushakov E V, Chechovsk S A, Litvin A P, Parfenov P S, Zakharov V V, Vukasik A, Fedorov A V and Baranov A V 2016 Optical properties of ordered superstructures formed from cadmium and lead chalcogenide colloidal nanocrystals Opt. Express 24 A538–64
[21] Pomp P P, Martiradonna L, Della Torre A, Della Sala F, Manna L, De Vittorio M, Calabili F, Cingolani R and Rinaldi R 2006 Metal-enhanced fluorescence of colloidal nanocrystals with nanoscale control Nat. Nanotechnol. 1 126–30
[22] Shandalov M and Golan Y 2004 Microstructure and morphology evolution in chemical solution deposited semiconductor films 2. PbSe on as face of GaAs (111) The European Physical Journal-Applied Physics 28 51–7
[23] Gholibadi N and Hatam E G 2015 Surface studies, structural characterization and quantity determination of PbSe nanocrystals deposited by chemical bath deposition technique. J. Cryst. Growth 418 111–4
[25] Moreno O P, Gutiérrez Pérez R, Chávez Portillo M, Hernández Téllez G, Rubio Rosas E, Cruz Cruz S and Moreno Rodríguez A 2016 Synthesis, morphological, optical and structural properties of PbSSe2– nanocrystals Optik 127 8341–9
[26] Templeman T, Sengupta S, Maman N, Bar-Or E, Shandalov M, Ezersky V, Yahel E, Sarusi G, Visoly-Fisher I and Golan Y 2018 Oriented attachment: a path to columnar morphology in chemical bath deposited PbSe thin films Crystal Growth & Design 18 1227–35
[27] Bhambhani P, Kabra K, Sharma B K and Sharma G 2014 High pressure study of structural and electronic properties of PbSe Journal of Solid-State Physics 2014 Online
[28] Schatschneider B, Liang J-J, Jezowski S and Tkatchenko A 2012 Phase transition between cubic and monoclinic polymorphs of the tetracyanoethylene crystal: the role of temperature and kinetics CrystEngComm 14 4656–63
[29] Wang S et al 2015 Revisit of pressure-induced phase transition in PbSe: crystal structure, and thermoelastic and electrical properties Inorg. Chem. 54 4981–9
[30] Wang R, Novendra N and Navrotsky A 2019 Energetics, structures, and phase transitions of cubic and orthorhombic cesium lead iodide (CsPbI3) polymorphs JACS 141 14501–4
[31] Yang L, Wei K, Xu Z, Li F, Chen R, Zheng X, Cheng X and Jiang T 2018 Nonlinear absorption and temperature-dependent fluorescence of perovskite FAPbBr3 nanocrystal Opt. Lett. 43 122–5
[32] Whitcher T J et al 2019 Dual phases of crystalline and electronic structures in the nanocrystalline perovskite CsPbBr3, NPG Asia Mater. 11 1–12
[33] Wageh S, Badr M H, Khalil M H and Eid A S 2009 Strong confinement of PbSe nanocrystals in phosphate glass Physica E 41 1157–63
[34] Khan A H, Brescia R, Polovitsyn A, Angeloni I, Martín-García B and Moreels I 2017 Near-infrared emitting colloidal PbS nanoplatelets: lateral size control and optical spectroscopy Chem. Mater. 29 2883–9
[35] Wang Q and Wu W 2018 Temperature and excitation wavelength-dependent photoluminescence of CH3NH3PbBr3 crystal Opt. Lett. 43 4923–6
[36] Dai Q, Zhang Y, Wang Y, Hu M Z, Zou B, Wang Y and Yu W W 2010 Size-dependent temperature effects on PbSe nanocrystals Langmuir 26 11435–40
[37] Yin Q, Liu C, Wang J, Zhao Z, Heo J, Xie J and Han J 2019 Excitation-wavelength- and size-dependent photo-darkening and photo-brightening of photoluminescence from PbS quantum dots in glasses Opt. Mater. Express 9 504–15
[38] Guinier A 1994 Fundamentals of X-ray diffraction theory X-ray Diffraction in Crystals, Imperfect Crystals, and Amorphous Bodies (New York: Dover Publications) 04806680118
[39] David JW 1996 Effect of temperature on crystal growth and crystal properties of paracetamol J. Chem. Soc., Faraday Trans. 92 439–44
[40] Dhanaraj G et al (ed) 2010 Springer Handbook of Crystal Growth, (Heidelberg: Springer Science & Business Media) 978-3-540-74761-1
[41] Hone F G and Dejene F B 2017 Tuning the optical band gap and stoichiometric ratio of chemically synthesized lead selenide thin films by controlling film thickness, J. Mater. Sci., Mater. Electron. 28 5979–89
[42] Ruppert C, Burak A view O and Heinz T F 2014 Optical properties and band gap of single- and few-layer MoTe2 crystals Nano Lett. 14 6231–6
[43] Geng D, Lozano G, Calvo M E, Núñez N O, Becerro A I, Ocaña M and Mínguez H 2017 Photonic tuning of the emission color of nanoporphor films processed at high temperature Adv. Opt. Mater. 5 1700099
[44] Zhu J, Koltynin Y and Gedanken A 2000 General sonochemical method for the preparation of nanophasic selenides: synthesis of ZnSe nanoparticles Chem. Mater. 12 73–8
[45] Feng W, Zhou H and Chen F 2015 Impact of thickness on crystal structure and optical properties for thermally evaporated PbSe thin films Vacuum 114 82–5
[46] Osherov A, Ezersky V and Golan Y 2007 Microstructure and morphology evolution in chemically deposited semiconductor films: 4. From isolated nanoparticles to monocrystalline PbS thin films on GaAs (100) substrates. Eur. Phys. J. Appl. Phys. 37 39–47