Colloidal PbS quantum dots (QDs) have provoked a revolution in the field of optoelectronic devices owing to their low-cost fabrication processing and excellent physical properties. Recently, the fabrication of nanostructured PbS QD photovoltaic (PV) devices based on zinc oxide (ZnO) nanowire array appears as an effective strategy for improving the overall device performance. Despite its potentially strong impact on the device performance, the role of nanowire areal density on photon absorption and exciton dynamics has not yet been studied and still remains unexplored. Here, for the first time, the areal density of ZnO nanowires is tuned through controlling the precursor concentration and its impact on PbS QD PV performance is studied. It is found that the device with optimized ZnO nanowire areal density yields significantly increased power conversion efficiency (PCE) (10.1% vs 8.5% of control nanowire-based device) due to improved antireflection effect and reduced surface recombination states. To further improve the photovoltaic performance, the ZnO nanowire surface is treated with hydrogen plasma. Transient photovoltage (TPV) measurement reveals that this passivation process noticeably reduces the nonradiative charge recombination yielding a champion device with a remarkable PCE of 10.8%.

1. Introduction

Colloidal quantum dots (CQDs) utilized for various optoelectronics devices\[1–5\] have attracted considerable attention as an emerging photovoltaic (PV) technology thanks to their tunable bandgaps, high stability, and solution processability\[6–14\]. Among them, lead sulfide (PbS) QDs devices\[15–21\] are particularly interesting due to their stability in air and size-tunable bandgaps with certified reported power conversion efficiency (PCE) up to 12%\[22\]. However, in these device structures, the optimum thickness of PbS QDs layer for absorbing all of the incoming photons is larger than the carrier diffusion length, leading to reduced carrier collection. In order to enhance the light absorption and carrier diffusion length without compromising collection efficiency, the interface engineering approaches such as utilization of functional organic ligands to passivate the surface states of QDs\[23,24\] or dot-to-dot surface coupled self-passivation schemes have been extensively studied\[25\]. However, the diffusion length still falls short of the optimum thickness for efficiently harvesting all near-infrared photons\[26\]. As an alternative approach, the ordered bulk heterojunction architectures such as ZnO nanowires embedded within PbS QDs PV structure have been utilized to decouple the direction of photon absorption and charge collection\[27–32\]. This method allows to use thick QDs active layer and greatly improves the charge extraction and short-circuit current ($J_{SC}$). Employing ZnO nanowires within PbS QDs PV has resulted in remarkably high $J_{SC}$ values of 31 mA cm$^{-2}$ and PCE of 9.6\%\[30\]. Moreover, very recently, record-high $J_{SC}$ values exceeding 33 mA cm$^{-2}$ with PCEs of 10.6\% have been demonstrated for PbS QDs PV with embedded ZnO nanowires utilizing a luminescence downshifting strategy\[31\]. Despite the increased value of $J_{SC}$, the reported highly dense ZnO nanowire arrays impose some limitations on the device performance. This mostly stems from the presence of large densities of surface states due
to the large surface-to-volume ratio of ZnO nanowires, which is detrimental to PV performance. In addition, closely packed ZnO nanowire arrays decrease the interfacial area between the absorber layer and nanowires leading to impeded carrier extraction. Thus, the ZnO nanowires should be sufficiently “wet” with QDs to reduce the path that the photogenerated carriers need to travel before they are extracted. Moreover, the light scattering and thereby optical absorption property of the QDs devices with densely packed ZnO nanowire arrays need improvement to enhance optical absorption inside the device structure. Despite its strong impact on the performance of QDs PVs, the role of nanowire area density still remains to be studied. It is also expected that the high density of surface states of ZnO nanowires adversely impact the photovoltaic performance of such devices. Therefore, surface passivation strategies that help to reduce the density of ZnO nanowire surface states are expected to enhance the charge carrier collection and overall PV performance of PbS QDs devices. The photoluminescence (PL) studies of ZnO nanowires have commonly revealed the presence of green emission, in addition to the main near band emission, which is mostly attributed to the presence of nanowire surface states.[32] Therefore, different surface treatments such as AlOx coating,[33] Ar ion milling,[34] and polymer covering[35] have been performed to suppress this deep level emission. Recently, hydrogen plasma (H-plasma) treatment was used to enhance the ZnO optical property by effectively quenching the green emission peak.[36] Here, for the first time, we demonstrate that by tailoring the areal density of grown ZnO nanowires and thereby the density of nanowire surface states and optical transmittance, the carrier collection and overall performance of PbS QDs devices are improved. Also they could benefit other PV devices such as polymer[37] and perovskite solar cells.[38–40] In this work, we use the precursor concentration as a handle to tailor the ZnO nanowire growth, which results in PCE increase from 8.5% for the device with highest ZnO nanowire areal density to 10% for the optimal ZnO nanowire areal density. Further passivation of ZnO nanowires using hydrogen plasma (H-plasma) successfully suppresses the nonradiative recombination leading to devices with a maximum PCE of 10.8%.

2. Results and Discussion

Figure 1a,b shows the schematic illustration of the PbS QDs device structure and energy band alignment of the indium-doped tin oxide (ITO), ZnO/ZnO nanowire, PbS-tetrabutylammonium iodide (TBAI), and PbS-1,2-ethanediethiol (EDT) layers at equilibrium. A 40 nm thick textured polycrystalline seed film of ZnO was initially deposited on prepatterned ITO substrates by a sol–gel process (for details see the Experimental Section). The ZnO nanowires were grown on the ZnO seed layer at 90 °C for 40 min using solution containing equal volume of zinc nitrate.

Figure 1. a) Schematic illustration of the PbS QDs PV device structure. b) Schematic qualitatively demonstrating the band alignment of the ZnO/ZnO nanowire, PbS-TBAI, and PbS-EDT layers at equilibrium, with typical conduction and valence band energy levels relative to vacuum as measured using ultraviolet photoelectron spectroscopy (UPS) and reported in refs. [24,30] c) Absorption and PL spectra of PbS QDs. d) FIB-milled cross-section SEM image of PbS QDs PV. Scale bar is 200 nm.
hexahydrate and hexamethylenetetramine in deionized water (DI) water (25 mL). The concentrations of precursor solution used for ZnO nanowire growth were selected to be 0.05, 0.033, 0.016, and 0.008 M, which we further refer to as A, B, C, and D, respectively. For reference, a planar sample with only ZnO seed layer was also prepared. Next, PbS QDs with the first absorption peak at \( \approx 950 \text{ nm} \) in solution (Figure 1c) were deposited using a layer-by-layer deposition method. To replace the native oleic acid ligands in the first ten layers, a solution of TBAI in methanol was used. The ligand exchange for the last two layers was performed by using a solution of EDT in acetonitrile. Au electrode was deposited by thermal evaporation through a shadow mask. The focused ion beam (FIB)-milled scanning electron microscopy (SEM) image of the PbS QDs/ZnO nanowire device is shown in Figure 1d.

The SEM images of the ZnO nanowire arrays grown with different concentrations show that the nanowire areal density is reduced by decreasing the concentration (Figure 2a). Figure 2b,c demonstrates the changes of height and diameter of ZnO nanowires and nanowire density with respect to precursor concentration, respectively. It is found that either nanowire dimension or length decreases with reducing the concentration. However, the decreased in the nanowire diameter is more pronounced, reducing from \( \approx 30 \text{ nm} \) for sample A to \( \approx 23 \text{ nm} \) for sample D. These results are consistent with the previously reported studies on ZnO nanowires growth at different concentrations.

The PV metrics and current density–voltage (J–V) characteristics of planar reference and PbS QDs/ZnO nanowire devices were measured under 100 mW cm\(^{-2}\) simulated AM 1.5MG illumination and are shown in Figure 3a,b and Table S1 in the Supporting Information. The planar reference device shows the highest \( V_{OC} \) in compared to all ZnO nanowire based PbS QDs devices. The \( V_{OC} \) drops by about 7% for the device with the highest ZnO nanowire areal density (device A). As expected, the resulting PbS QDs/ZnO nanowire devices show noticeable increase in \( J_{SC} \) in comparison to planar reference device. The highest improvement in \( J_{SC} \) was observed for the device B and device C, which resulted in 27% and 30% PCE improvement compared to planar reference device, respectively. The
improvement in $V_{OC}$ of devices C and D (low concentration) can be attributed to the decreased areal density of ZnO nanowires, which considerably reduces the total number of nanowire surface states that would otherwise act as nonradiative recombination centers (vide infra). The champion device C yielded a $V_{OC}$ of 595 mV and PCE of 10.1% under 100 mW cm$^{-2}$ AM1.5G illumination.

The comparison of the semi-log dark $J$–$V$ characteristics of representative planar and ZnO nanowire PbS QDs devices shows that device A exhibits higher current leakage (Figure S2, Supporting Information) which can be attributed to the significantly enhanced carrier recombination at the heterointerface. To further evaluate the carrier recombination in PbS QDs/ZnO nanowire devices, we studied the variations of their $V_{OC}$ as a function of light intensity and extracted their diode ideality factors ($n$). For PV devices with significant trap-assisted charge recombination, the value of $n$ is larger than 1 ($1 < n < 2$)[43] and PbS QDs devices usually have ideality factor in the range of 1.5–2.[44–46] The light-intensity dependence of the $V_{OC}$ was studied according to the relation $V_{OC} \propto n kT/q$, where $n$ is the diode ideality factor, $T$ is the temperature, $k$ is the Boltzmann constant, and $q$ is elementary charge.[16,47] Fitting the $V_{OC}$ versus light intensity of resulting PbS QDs devices yielded the values of $n$ as 1.3, 1.9, and 1.45, for the planar device, device A and device C, respectively (Figure 3c). The value of $n$ for the devices B and D are determined to be 1.76 and 1.35, respectively, as shown in Figure S3 in the Supporting Information. Since the only difference between these devices is the degree of areal density of incorporated ZnO nanowires, these results strongly suggest that the trap-assisted charge recombination at the ZnO nanowire/QDs interface was monotonically suppressed by reducing the areal density of ZnO nanowires.

As mentioned above, the $J_{SC}$ of the ZnO nanowire based PbS QDs devices is noticeably higher than that of planar device due to improved charge collection. External quantum efficiency (EQE) spectra were measured to further confirm the obtained photocurrent values and shown in Figure 3d. The corresponding integrated photocurrent densities were also calculated by integrating the AM1.5G spectrum with the EQE spectra. The $J_{SC}$ values obtained from the $J$–$V$ curves are ≈1 mA cm$^{-2}$ higher than those of the calculated $J_{SC}$ from EQE results. This difference is due to the spectral mismatch between the spectra of the utilized solar simulator and that of AM1.5G. The enhancement in $J_{SC}$ for PbS QDs devices with ZnO nanowires grown at lower concentration (devices C and D) can be attributed to the increased optical transmittance and improved light
scattering. However, the value of $J_{SC}$ for device D is similar to that of device A (high concentration) likely due to the presence of small ZnO nanowires areal density of this sample to effectively contribute in photogenerated carrier collection.

**Figure 4a** shows the transmittance spectra of the investigated ZnO nanowires films deposited on ITO/glass substrates. We note that all ZnO nanowire-based films show higher transmittance compared to planar reference film. As previously reported in the literature, nanowires work as an antireflection layer due to their optical scattering effect (out-coupling property), which can increase the transmittance and decrease the reflectance of ITO glass substrate as compared to planar ZnO. For light management in the nanostructures, one of the main parameters for nanowire array is their pitch size, defined as the distance between the cores of two adjacent nanowires. In order to effectively reduce the reflectance of nanowire array, the optimum value for pitch size is required. Previously, we studied the role of pitch size in a nanocone array (as an antireflection layer) on the reflectance spectra and device performance. Our finite-difference time-domain (FDTD) simulation results and reflectance measurement in that study revealed that 1 $\mu$m pitch size is the best condition for reducing the reflection in the device. Moreover, in the literature, many people studied the role of pitch size on optical properties of nanowire arrays. For example, Hua et al. found that by controlling the diameter of nanowire and the pitch size, they can drastically enhance the absorption of germanium (Ge) nanowire arrays (reduce its reflection), as they proved it by FDTD simulation as well. They found that 500 nm pitch size with 300 nm diameter for Ge nanowires show the best absorption (lowest reflection). Similar principle can be applied to our ZnO nanowire array study. By decreasing the areal density to $\approx$150 (per $\mu$m$^2$) for sample C, we will have a lowest reflection, due to the light scattering effect. We have included the measured reflectance spectra of different nanowire samples in Figure S4 in the Supporting Information. In the absence of optical absorption for ZnO at wavelengths $>$400 nm, the difference in transmittance of different nanowires is due to their different reflectance spectra, for instance, the lower transmittance of the nanowire sample with the highest areal density is due to its higher reflectance. The transmittances of films B and C are higher than that of film A, which well corroborate with their corresponding $J_{SC}$ values. Moreover, the transmittance for film D is lower than that found in other ZnO nanowire-based devices suggesting reduced optical absorption, which in fact can be a reason for its lower $J_{SC}$. The reduced areal density of the ZnO nanowires grown at lower concentration could potentially facilitate their infiltration with PbS QDs. In order to gain more information about carrier recombination inside the device structure, we employed electrochemical impedance spectroscopy (for the discussion and data see Figure S5, Supporting Information), which confirms the reduction of exciton recombination for the devices with lower areal density of ZnO nanowires.

The hydrogen plasma treatment (see the Experimental Section) was performed on sample with optimal precursor concentration (device C) to passivate ZnO nanowire surface states and further improves the photovoltaic performance. Figure 4b shows the room-temperature PL spectra of the as-prepared and H-plasma treated ZnO nanowires films. Both spectra exhibit the near band emission of ZnO at around 380 nm together with a significantly broader emission in the visible region. The
PL spectrum of the H-plasma treated ZnO nanowire shows a clear enhancement of the near band emission accompanied with a significant quenching of the visible emission peak. The enhancement of the near band emission can be attributed to the reduced nonradiative recombination centers as a result of ZnO nanowire surface passivation. Recent work revealed that increase of radiative recombination can be a result of H doping and might be also partially played a role here too.

X-ray electron spectroscopy (XPS) was performed to study the composition of ZnO nanowire surface before and after H-plasma treatment. The O 1s peak of the ZnO nanowire samples before and after H-passivation are shown in Figure 4c. Deconvolution of the spectra shows two main peaks located at 529.9 and 531.3 eV, which are attributed to Zn–O–Zn and hydroxyl groups, respectively. In contrast to the Zn–O–Zn peak, which is similar in intensity for both samples, the H-plasma treated film shows a noticeable increase of the hydroxyl group peak. This result suggests that H-plasma treatment passivated O-dangling bonds on the ZnO nanowire surface and thereby reduced the density of surface states.

Figure 5a shows the representative J–V characteristics of the optimized PbS QDs devices with and without H-plasma treatment. Compared with the nontreated device with a PCE of 9.9%, the H-plasma passivated device shows improvement in PV metrics leading to a maximum PCE of 10.8%. The EQE spectra of the devices and integrated current densities confirm the photocurrent improvement upon H-plasma treatment (Figure 5b). We believe that the small reverse saturation current together with the reduced charge recombination upon ZnO nanowire surface passivation are likely the reasons for improved values of $V_{OC}$ and fill factor. The light-intensity dependence of $V_{OC}$ was measured for passivated and unpassivated devices, as shown in Figure 5c. The $n$ value for the passivated device is extracted to be $\approx 1.33$, which is noticeably lower than that of the unpassivated device ($n = 1.45$) suggesting a reduced trap-assisted charge recombination. To confirm this result, we analyzed the devices using the transient photovoltage (TPV) measurements (at open-circuit condition with 1 sun light bias) (Figure 5d). The device with passivated ZnO nanowires exhibits a clearly slower voltage decay (1005 µs) compared to unpas-sivated device (830 µs), which further confirms the reduced density of trap states for the passivated device. In addition, the electroluminescence (EL) measurements at $V = 1$ (V) show stronger emission peak for the passivated device, which is
likely due to decreased nonradiative recombination centers at the nanowire/QDs interface (Figure 6a). The passivation process also reduces the reverse leakage current (Figure 6b) and recombination resistance indicating a lower charge recombination at the PbS QDs/ZnO nanowire interface (Figure 6c). The devices with and without passivation show good stability after storing in ambient environment over 7 d (Figure S6, Supporting Information).

3. Conclusion

In summary, the impact of areal density of ZnO nanowires as an electron transporting layer (ETL) inside PbS QDs PV was studied and a direct correlation between nanowire density and PV performance was observed. The areal density of ZnO nanowires was precisely optimized by controlling the precursor concentration, resulting in enhanced optical transmittance of ZnO ETL. We found that areal density of \( \approx 150 \ \mu m^{-2} \) (corresponding to nanowire growth precursor concentration of \( 0.016 \) M) has the best effect on the absorption and device performance leading to a champion device with a high \( J_{SC} \) of \( 30 \) mA cm\(^{-2} \) and PCE of 10.1%. To further improve the device performance, the surface of ZnO nanowires was treated by H-plasma. We demonstrated that this process significantly reduces the nanowire surface states leading to a device with improved PCE up to 10.8%. Our results demonstrate that the precise control of ZnO nanowire growth together with H-plasma treatment can be utilized to manipulate the optical and electrical characteristics, and as a result improve the overall PV performance. The proposed ZnO nanowire growth and H-plasma passivation processes are not only limited to PbS QDs PV and could also benefit other PV devices such as polymer and perovskites solar cells.

4. Experimental Section

PbS QDs Device Fabrication and Plasma Treatment: After cleaning ITO substrates of 200 nm thickness in DI water and acetone, the substrates
were exposed to oxygen plasma (PDC 32C, Harrick Plasma) for 1 min. A solution of zinc acetate dihydrate (0.3 M) and ethanolamine (0.3 M) in 2-methoxyethanol was spin coated on the cleaned ITO substrates followed by an annealing step (200 °C for 15 min). PbS colloidal QDs with native oleic acid ligands and first exciton absorption peak in solution positioned at ~950 nm were synthesized as previously reported in literature.\textsuperscript{18,19} Layer by layer PbS QDs deposition onto the ZnO-coated ITO/glass substrates was performed by spin coating. A solution of PbS QDs with concentration of 50 mg mL\textsuperscript{−1} in octane was spin coated at 2000 rpm for 25 s. TBAI (Sigma-Aldrich) exchange for the native oleic acid capping ligands was performed using a TBAI solution in methanol with 10 mg mL\textsuperscript{−1} concentration and waiting 25 s followed by spin drying and two times rinsing with methanol. The QDs deposition steps with TBAI ligand exchange were repeated for ten layers. The ligand exchange for the last 2 layers was done using a solution of EDT in acetonitrile, waiting 25 s followed by spin drying and two times rinsing with acetonitrile. The concentration of EDT solution was 0.054 cm\textsuperscript{2}. The hydrogen rate of 1 Å s\textsuperscript{−1} was provided by a 150 W xenon arc-lamp (Newport 96000) with an AM1.5G filter. Optical transmittance and absorbance spectra were measured using a UV–vis–NIR spectrometer (Lambda 1050). EQE was measured by a 250 (W) tungsten halogen lamp source with an Oriel Cornerstone 130 monochromator. Integrated photocurrent densities were calculated by integrating the product of the EQE and the AM1.5G spectra over the wavelength range of 370–1100 nm. PL spectra were measured with a SpectraPro 300i spectrometer equipped with liquid nitrogen cooled InGaAs detector arrays (Princeton Instruments). EL spectra were collected on the same setup as for the PL measurements using a 2400 Keithley sourcemeter. XPS measurements were performed using an X-ray photoelectron spectrometer (ESCALAB 250Xi, Thermo Fisher SCIENTIFIC INC, USA) with Al K\textsubscript{α} radiation (hv = 1486.6 eV) as a source. The transient photovoltage decay measurements were done with a Tektronix TDS 3054B oscilloscope using a pulse laser light source emitting at 532 nm under white light bias provided by solar simulator. A series of neutral density filters were used between the light source and the device for light intensity dependence measurements. The SEM imaging was performed using an Field Electron and Ion Company (FEI) Helios NanoLab 600 at 5 kV. To prepare device cross-sections for SEM, Ga FIB milling at 30 kV was used on the same tool.

Device Characterization: Current–voltage (J–V) characteristics were measured in a nitrogen-filled glovebox by a semiconductor device analyzer (Agilent Technologies, B1500A). The solar simulator illumination of 100 mW cm\textsuperscript{−2} was provided by a 150 W xenon arc-lamp (Newport 96000) with an AM1.5G filter. Optical transmittance and absorbance spectra were measured using a UV–vis–NIR spectrometer (Lambda 1050). EQE was measured by a 250 (W) tungsten halogen lamp source with an Oriel Cornerstone 130 monochromator. Integrated photocurrent densities were calculated by integrating the product of the EQE and the AM1.5G spectra over the wavelength range of 370–1100 nm. PL spectra were measured with a SpectraPro 300i spectrometer equipped with liquid nitrogen cooled InGaAs detector arrays (Princeton Instruments). EL spectra were collected on the same setup as for the PL measurements using a 2400 Keithley sourcemeter. XPS measurements were performed using an X-ray photoelectron spectrometer (ESCALAB 250Xi, Thermo Fisher SCIENTIFIC INC, USA) with Al K\textsubscript{α} radiation (hv = 1486.6 eV) as a source. The transient photovoltage decay measurements were done with a Tektronix TDS 3054B oscilloscope using a pulse laser light source emitting at 532 nm under white light bias provided by solar simulator. A series of neutral density filters were used between the light source and the device for light intensity dependence measurements. The SEM imaging was performed using an Field Electron and Ion Company (FEI) Helios NanoLab 600 at 5 kV. To prepare device cross-sections for SEM, Ga FIB milling at 30 kV was used on the same tool.

Supporting Information
Supporting Information is available from the Wiley Online Library or from the author.

Acknowledgements
This work was in part supported by Natural Sciences and Engineering Research Council (NSERC) of Canada (Award No. PDF-487850-2016). D.P. acknowledges the financial support from the HOMING program of the Foundation for Polish Science cofinanced by the European Union under the European Regional Development Fund (POIR.04.04.00-00-SE7/18-00). M.M.T. would like to acknowledge Research Laboratory of Electronics (RLE) at Massachusetts Institute of Technology.

Conflict of Interest
The authors declare no conflict of interest.

Keywords
colloidal quantum dots, lead sulfide, nanowires, passivation, photovoltaics

Received: September 9, 2019
Revised: October 26, 2019
Published online: November 18, 2019

[1] J. Gao, A. F. Fidler, V. I. Klimov, Nat. Commun. 2015, 6, 8185.
[2] G. Konstantatos, I. Howard, A. Fischer, S. Hoogland, J. Clifford, E. Klem, L. Evina, E. H. Sargent, Nature 2006, 442, 180.
[3] A. F. Fidler, J. Gao, V. I. Klimov, Nat. Phys. 2017, 13, 604.
[4] Z. Ren, J. Sun, H. Li, P. Mao, Y. Wei, X. Zhong, J. Hu, S. Yang, J. Wang, Adv. Mater. 2017, 29, 1702035.
[5] Y. Shirasaki, G. J. Supran, M. G. Bawendi, V. Bulović, Nat. Photonics 2013, 7, 13.
[6] M. Yuan, M. Liu, E. H. Sargent, Nat. Energy 2016, 1, 16016.
[7] M. L. Böhm, T. C. Jellicoe, M. Tabachnyk, N. J. L. K. Davis, W. Xu, F. Tan, Q. Liu, X. Liu, Q. Jiang, L. Wei, W. Zhang, Z. Wang, Nat. Photonics 2015, 9, 1507.
[8] H. Aqoma, S.-Y. Jang, Energy Environ. Sci. 2018, 11, 1603.
[9] G. H. Carey, A. L. Abdelhady, Z. Ning, S. M. Thon, O. M. Bakulin, N. C. Greenham, Nano Lett. 2015, 15, 7987.
[10] A. Swarnkar, A. R. Marshall, E. M. Sanehira, B. D. Chernomordik, D. T. Moore, J. A. Christians, T. Chakraborti, J. M. Luther, Science 2016, 354, 92.
[11] J. Yuan, X. Ling, D. Yang, F. Li, S. Zhou, J. Shi, Y. Qian, J. Hu, Y. Sun, Y. Yang, X. Gao, S. Duhm, Q. Zhang, W. Ma, Joule 2018, 2, 2450.
[12] X. Ling, S. Zhou, J. Yuan, J. Shi, Y. Qian, B. W. Larson, Q. Zhao, C. Qin, F. Li, G. Shi, C. Steward, J. Hu, X. Zhang, J. M. Luther, S. Duhm, W. Ma, Adv. Energy Mater. 2019, 9, 1900721.
[13] F. Li, S. Zhou, J. Yuan, C. Qin, Y. Yang, J. Shi, X. Ling, Y. Li, W. Ma, ACS Energy Lett. 2019, 4, 2571.
[14] M. Liu, O. Voznyy, R. Sabatini, F. P. García de Arquer, R. Munir, A. H. Balawi, X. Lan, F. Fan, G. Walters, A. R. Kirmani, S. Hoogland, F. Laquai, A. Amassian, E. H. Sargent, Nat. Mater. 2017, 16, 258.
[15] H. T. Dastjerdi, R. Tavakoli, P. Yadav, D. Prochowicz, M. Saliba, M. M. Tavakoli, ACS Appl. Mater. Interfaces 2019, 11, 26047.
[16] W. Xu, F. Tan, Q. Liu, X. Liu, J. Liang, W. Zhang, Z. Wang, S. Qu, Z. Wang, Sol. Energy Mater. Sol. Cells 2017, 159, 503.
[17] X. Lan, O. Voznyy, F. P. García de Arquer, M. Liu, J. Xu, A. H. Proppe, G. Walters, F. Fan, H. Tan, M. Liu, Z. Yang, S. Hoogland, E. H. Sargent, Nano Lett. 2016, 16, 4630.
[18] K. Lu, Y. Wang, Z. Liu, L. Han, G. Shi, H. Fang, J. Chen, X. Ye, S. Chen, F. Yang, A. G. Shulga, T. Wu, M. Gu, S. Zhou, J. Fan, M. A. Loi, W. Ma, Adv. Mater. 2018, 30, 1707572.
[19] M. M. Tavakoli, A. Simchi, Z. Fan, H. Aashuri, ChemComm 2016, 52, 323.
[20] M. M. Tavakoli, H. Aashuri, A. Simchi, Z. Fan, Phys. Chem. Chem. Phys. 2015, 17, 24412.
[21] J. Xu, O. Voznyy, M. Liu, A. R. Kirmani, G. Walters, R. Munir, M. Abdelsamie, A. H. Proppe, A. Sarkar, F. P. García de Arquer, M. Wei, B. Sun, M. Liu, O. Ouellette, R. Quintero-Bermudez, J. Li, J. Fan, L. Quan, P. Todorovic, H. Tan, S. Hoogland, S. O. Kelley, M. Stefik, A. Amassian, E. H. Sargent, Nat. Nanotechnol. 2018, 13, 456.
