Shape Evolution and Single Particle Luminescence of Organometal Halide Perovskite Nanocrystals

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ABSTRACT Organometalic halide perovskites CH₃NH₃PbX₃ (X = I, Br, Cl) have quickly become one of the most promising semiconductors for solar cells, with photovoltaics made of these materials reaching power conversion efficiencies of near 20%. Improving our ability to harness the full potential of organometal halide perovskites will require more controllable syntheses that permit a detailed understanding of their fundamental chemistry and photophysics. In this manuscript, we systematically synthesize CH₃NH₃PbX₃ (X = I, Br) nanocrystals with different morphologies (dots, rods, plates or sheets) by using different solvents and capping ligands. CH₃NH₃PbI₃ nanowires and nanorods capped with octylammonium halides show relatively higher photoluminescence (PL) quantum yields and long PL lifetimes. CH₃NH₃PbI₃ nanowires monitored at the single particle level show shape-correlated PL emission across whole particles, with little photobleaching observed and very few off periods. This work highlights the potential of low-dimensional organometal halide perovskite semiconductors in constructing new porous and nanostructured solar cell architectures, as well as in applying these materials to other fields such as light-emitting devices and single particle imaging and tracking.

KEYWORDS: organometal halide perovskites · nanocrystals · preferred orientation · morphology control · size control · single particle photoluminescence

Organometal halide perovskites of general composition CH₃NH₃PbX₃ (X = Cl, Br, I) are attractive semiconductor materials due to their low cost and outstanding photovoltaic performance.¹−⁷ Organometal perovskites can be substituted with a variety of small organic cations and main group metals (in lieu of CH₃NH₃⁺ and Pb²⁺, respectively), leading in some cases to improved light harvesting capabilities.⁸−¹⁴ The organolead perovskites CH₃NH₃PbBr₃ and CH₃NH₃PbI₃ have bandgaps of 2.3 eV (540 nm) and 1.5 eV (820 nm),¹⁵,¹⁶ respectively, as well as high absorption coefficients. Critically, organolead perovskites have very long electron—hole carrier diffusion lengths, exceeding 1 µm in CH₃NH₃PbI₃ and 100 nm in CH₃NH₃PbBr₃, which in principle allows for the development of several solar cell architectures including perovskite-sensitized solar cells, planar heterojunction solar cells, and meso- and nanostructured solar cells.¹⁷ Building on the dramatic improvement of solar cell performance using the solid hole conductor spiro-OMeTAD instead of a liquid electrolyte (spiro-OMeTAD stands for 2,2’-7,7’-tetrakis(N,N-di-p-methoxy-phenylamine)-9,9’-spirobifluororene),¹⁸ the energy conversion efficiency of photovoltaics made from these intensely absorbing, visible-active semiconductors has risen from 3.8% to near 20% in only four years.¹⁹,²⁰ Photovoltaic performance depends critically on perovskite composition, crystallinity and morphology.²¹−²³ Higher perovskite film uniformity leads to lower recombination rates in planar heterojunction solar cells.²⁴,²⁵ Film uniformity is affected by factors such as precursor composition, annealing temperature and, if applicable, solvent used during the vapor-assisted or spin coating deposition process.⁶,²⁴−²⁶,³² Highly efficient mesostructured solar cells are produced by a two-step deposition process.³³−³⁵ Vapor-assisted methods and additives provide the means...

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to control the extent of perovskite crystallization.\textsuperscript{11,12,23–38} Highly crystalline perovskite domains maximize sunlight harvesting because they have few undesirable grain boundaries, providing well-defined CH$_3$NH$_3$PbI$_3$/spiro-OMeTAD interfaces and thus decreasing surface recombination.\textsuperscript{26,29,36} Photovoltaic device performance also depends on the relative absence of perovskite bulk defects and surface states.\textsuperscript{39–41}

On the basis of their very long electron–hole diffusion lengths (see above), perovskite borne photogenerated carriers should be able to very easily hop across grain boundaries from one single crystalline domain to another, enabling the incorporation of preformed, low-dimensional (nanosized) perovskites into the photoactive layer. Already, some of the best perovskite solar cells reported to date use a photoactive perovskite active layer. Already, some of the best perovskite solar cells reported to date use a photoactive perovskite layer deposited within mesoscopic conductive scaffolds. Nevertheless, preparation methods that could achieve more controllable and favorable crystallinity and morphology are lacking.\textsuperscript{47} Development of more thorough and extensive understanding of the crystallization process lengths (see above), perovskite borne photogen-

**RESULTS AND DISCUSSION**

\textbf{CH$_3$NH$_2$PbI$_3$: Synthesis and Solvent Effects.} A simple solution phase method to synthesize organometal perovskites starts by codissolving lead(II)- and methylammonium-halides (PbX$_2$ and CH$_3$NH$_3$X) in a polar solvent such as N,N-dimethylformamide ($\epsilon = 38.25$), $\gamma$-butyrolactone ($\epsilon = 39.1$), or acetonitrile ($\epsilon = 36.64$).\textsuperscript{48} An excess of methylammonium halide is often required in order to increase the solubility of lead dihalide, likely due to the formation of a yet unidentified adduct. This mixed precursor solution is then added to a lower polarity solvent such as ethyl acetate ($\epsilon = 6.0$), chloroform ($\epsilon = 4.81$) or toluene ($\epsilon = 2.38$), causing precipitation or “crashing” of crystalline perovskite.

When prepared by the general method above, the dark brown-colored (1.5 eV band gap)\textsuperscript{15} perovskite CH$_3$NH$_3$PbI$_3$ crystallizes with a tetragonal unit cell (Figure 1).\textsuperscript{34–36} Interestingly, the specific combination of precursor (polar) and crashing (nonpolar) solvents has a large effect on the relative diffraction peak intensities observed for this material by XRD (Figure 2). For example, dissolving the methylammonium and lead halide precursors in $\gamma$-butyrolactone, followed by crashing with either chloroform or toluene, yields a product with relatively strong (110) and (220) diffraction peaks and relatively weak (222) and (224) diffraction peaks (Figure 2). This situation indicates a preferred orientation and becomes much more dramatic when the precursors are dissolved in acetonitrile, followed by crashing with toluene; in this case, most diffraction peaks become barely visible in comparison to the main (110) and (220) XRD peaks (Figure 2).

\textbf{CH$_3$NH$_3$PbI$_3$: Synthesis and Cation Effects.} In a perfectly packed, classical AM$_3$ perovskite (where A = cation, M = central cation, and X = anion having ionic radii $R_A$, $R_M$, and $R_X$, respectively), the tolerance factor (t), defined as $t = (R_A + R_X)/\sqrt{2} \times (R_M + R_X)$, must remain close to unity ($t \approx 1$).\textsuperscript{49,50} In organometal halide perovskites, only small organic cations consisting of, at most, three C or N atoms are able to fit within the APbX$_3$ structure and fulfill this requirement (for example, when the cation is CH$_3$NH$_3$\textsuperscript{+}).\textsuperscript{15,51–58} The addition of larger organic cations is expected to either frustrate or terminate the lattice or both, potentially leading to interesting morphology- and size-controlled crystals (for example, nanosized, anisotropic, or layered perovskites).\textsuperscript{16} To explore this possibility, we repeated the procedure outlined above while adding a large n-octylammonium cation source such as CH$_3$(CH$_2$)$_7$NH$_3$ and using acetonitrile and toluene as the precursor and crashing solvents, respectively (Scheme 1).

Increasing the initial concentration of CH$_3$(CH$_2$)$_7$NH$_3$ (z in Scheme 1) in the precursor solution while keeping the concentrations of PbI$_2$ and CH$_3$NH$_3$I constant (x and y,
respectively) leads to a progressive increase in the relative intensity of the (110) and (220) XRD diffraction peaks (Figure 3). Similar to what is observed for certain solvents (see above), this strongly indicates that the individual perovskite crystals either pack or grow (i.e., orient) themselves preferentially along the [110] and (related) [220] directions (see below). Critically, none of the other known and related (CH₃NH₃)ₙPbIₖₘ (n, m = 2, 4; 3, 5; 4, 6) phases are observed by XRD (see Supporting Information available).

CH₃NH₃PbI₃: Preferred Orientation and Direction of Growth.

To better understand the microscopic basis for the preferred orientation behavior observed for CH₃NH₃PbI₃ by XRD, we performed additional characterization using Scanning and Transmission Electron Microscopies (SEM and TEM). “Bulk” CH₃NH₃PbI₃ made in the absence of any directing agents (without CH₃(CH₂)₇NH₃I) consists of relatively large 0.3–2 μm particles having cube-like or tetragonal morphology (Figure 4a and Table 1, entry 1). In contrast, CH₃NH₃PbI₃ made by reacting acetonitrile solutions of PbI₂, CH₃NH₃I, and CH₃(CH₂)₇NH₃I with toluene consists of high aspect ratio, elongated nanocrystals having wire- and rod-like morphologies (Figure 4, panels b and c, respectively). High-resolution transmission electron microscopy (HR-TEM) shows that the elongated perovskite single crystals grow along the [110] crystallographic direction; in other words, the main (long) axis of these perovskite nanowires and nanorods is perpendicular to the (220) set of lattice planes, which have a d spacing of 3.14 Å (Figure 5).

CH₃NH₃PbI₃: Morphology and Aspect Ratio Control.

We find that a convenient way of controlling the aspect ratio of CH₃NH₃PbI₃ nanorods and nanowires is by fine-tuning the rate at which the precursor solution is added to the crashing solvent. For example, single (fast) addition of an acetonitrile solution (4.6 mL) containing PbI₂ (0.435 mM), CH₃NH₃I (1.305 mM), and CH₃(CH₂)₇NH₃I (1.305 mM) to toluene (15 mL) produces relatively long and thin (1500 nm x 34 nm) CH₃NH₃PbI₃ nanowires,
along with a minor fraction of smaller (ca. 24 nm) CH$_3$NH$_3$PbI$_3$ nanocrystals (Figure 4b and Table 1, entry 2). When the precursor solution addition rate is slowed down to 0.01 mL/min, the wires become shorter and fatter (810 nm × 54 nm), with a much-reduced fraction of small nanocrystals formed as a byproduct (Figure 4c and Table 1, entry 3). Critically, only nanocrystals produced in the presence of CH$_3$(CH$_2$)$_7$NH$_3$I contain a strong asymmetric C/H stretching vibration at 2940 cm$^{-1}$, consistent with the presence of CH$_2$-containing C8 groups, likely on the surface of the particles (see Supporting Information). Interestingly, when 1,3-pentanediarnonium iodide ([H$_3$NCH$_2$CH$_2$CHNH$_3$CH$_2$CH$_3$]I$_2$) and γ-butyrolactone are used instead of n-octyl ammonium iodide and acetonitrile, XRD shows the formation of a crystalline product made up of much smaller single crystalline domains that lack the preferred orientation behavior observed above, as evidenced by the very broad diffraction peaks and the resemblance of their relative intensity compared to the standard XRD pattern, respectively (Figure 6 and Table 1, entry 4). TEM shows that the product of this reaction is composed of aggregates of relatively small (ca. 20 nm), isotropic nanodots (Figure 4d). These data suggest that the bidentate nature of the diammonium 1,3-pentanediarnonium cation has two effects: First, it behaves as a larger, more sterically encumbered organic dication, leading to more extensive capping and termination and thus to smaller crystalline domains. Second, because this dication has two ammonium headgroups, it can easily link neighboring nanocrystal surfaces, leading to the observed aggregation.

### Table 1. Methylammonium-Lead Iodide and Bromide Perovskites Synthesized under Various Conditions

<table>
<thead>
<tr>
<th>no.</th>
<th>sample</th>
<th>dimensions</th>
<th>band gap (eV)</th>
<th>PL max (nm)</th>
<th>quantum yield (%)</th>
<th>lifetime (ns)</th>
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<tr>
<td>1</td>
<td>CH$_3$NH$_3$PbI</td>
<td>0.3–2 μm</td>
<td>1.56</td>
<td>762</td>
<td>0.34</td>
<td>9</td>
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<td>CH$_3$NH$_3$PbI</td>
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<td>756</td>
<td>1.7</td>
<td>30</td>
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<tr>
<td>3</td>
<td>CH$_3$NH$_3$PbI</td>
<td>810 nm × 54 nm</td>
<td>1.57</td>
<td>760</td>
<td>1.4</td>
<td>66</td>
</tr>
<tr>
<td>4</td>
<td>CH$_3$NH$_3$PbI</td>
<td>Dots: 20 nm</td>
<td>1.62</td>
<td>744</td>
<td>0.025</td>
<td>9</td>
</tr>
<tr>
<td>5</td>
<td>CH$_3$NH$_3$PbBr</td>
<td>200–800 nm</td>
<td>2.26</td>
<td>540</td>
<td>1.2</td>
<td>9</td>
</tr>
<tr>
<td>6</td>
<td>CH$_3$NH$_3$PbBr</td>
<td>500 nm × 47 nm</td>
<td>2.30</td>
<td>531</td>
<td>13</td>
<td>9</td>
</tr>
<tr>
<td>7</td>
<td>CH$_3$NH$_3$PbBr</td>
<td>150 nm × 30 nm</td>
<td>2.30</td>
<td>520</td>
<td>0.43</td>
<td>9</td>
</tr>
</tbody>
</table>

CH$_3$NH$_3$PbBr$_3$: Morphology and Aspect Ratio Control. The yellow-colored (2.3 eV band gap)\textsuperscript{15,16} CH$_3$NH$_3$PbBr$_3$ perovskite crystallizes with a cubic unit cell (Figure 7). On the basis of the aforementioned effects of solvent, cation and precursor addition rate on the synthesis of crystalline CH$_3$NH$_3$PbI$_3$, we sought to use a similar approach to control the size and shape of CH$_3$NH$_3$PbBr$_3$ nanocrystals. We specifically used lead(II)-, methylammonium-, and $n$-octylammonium-bromide as precursors, and a combination of N,N-dimethylformamide (DMF) and toluene as the precursor and crashing solvents, respectively (Scheme 2).

The XRD pattern of “bulk” CH$_3$NH$_3$PbBr$_3$ obtained in the absence of large cation sources closely resembles the standard pattern reported for cubic CH$_3$NH$_3$PbBr$_3$, indicating that there is no preferred orientation in this sample (Figure 8).\textsuperscript{17} In contrast, addition of $n$-C$_8$H$_{17}$NH$_3$Br to the precursor solution during synthesis causes a very significant increase in the relative intensities of the (001) and (002) diffractions in comparison to the rest of the peaks in the standard pattern of this material; this observation strongly indicates that crystalline CH$_3$NH$_3$PbBr$_3$ made in this way preferentially grows along the [001] crystallographic direction (Figure 8).

In agreement with our XRD observations, TEM shows that CH$_3$NH$_3$PbBr$_3$ made in the absence of large organic cations (without CH$_3$(CH$_2$)$_7$NH$_3$Br) consists of relatively large 0.2–0.8 μm particles (Figure 9a and Table 1, entry 5). In contrast, CH$_3$NH$_3$PbBr$_3$ made by reacting DMF solutions of PbBr$_2$, CH$_3$NH$_3$Br and CH$_3$(CH$_2$)$_7$NH$_3$Br with toluene consists of much smaller and higher aspect ratio, anisotropic nanocrystals having wire- and plate-like morphologies (Figure 9, panels b and c, respectively). In this case, a convenient way of controlling the specific morphology of CH$_3$NH$_3$PbBr$_3$ nanocrystals is to vary the initial concentration of methyl- and $n$-octyl-ammonium (C1 and C8) bromides.
in the DMF precursor solution. For example, single addition of a DMF solution (1.5 mL) containing PbBr$_2$ (2.7 mM), CH$_3$NH$_3$Br (2.7 mM) and CH$_3$(CH$_2$)$_7$NH$_3$Br (5.4 mM) to toluene (15 mL) produces CH$_3$NH$_3$PbBr$_3$ nanowires (500 nm/C$^2_{47}$ nm) (Figure 9b and Table 1, entry 6). When this experiment is repeated with the same PbBr$_2$ concentration (2.7 mM) while decreasing the concentration of CH$_3$NH$_3$Br (0.68 mM) and CH$_3$(CH$_2$)$_7$NH$_3$Br (2.0 mM), the product is made of smaller elongated CH$_3$NH$_3$PbBr$_3$ nanocrystals (150 nm/C$^2_{30}$ nm) with plate-like morphology (Figure 9b and Table 1, entry 7).

**Instability of CH$_3$NH$_3$PbBr$_3$ under a TEM Beam.** During structural characterization by TEM, we noticed that CH$_3$NH$_3$PbBr$_3$ nanocrystals are particularly unstable under an electron beam. For example, when imaging with a FEI-Tecnai 2-F20 STEM operating at 200 kV, CH$_3$NH$_3$PbBr$_3$ nanoplates quickly (3–5 min) “melt” away to form smaller dot-like particles of unknown composition (Figure 10). The particles produced in this way were not particularly crystalline, and thus far we were unable to identify their exact composition. This behavior is not unique to nanocrystalline CH$_3$NH$_3$PbBr$_3$, as bulk CH$_3$NH$_3$PbBr$_3$ decomposed just as fast under the electron beam (see Supporting Information). The close resemblance between this behavior and a recently reported synthesis of CH$_3$NH$_3$PbBr$_3$ nanodots raises questions about the true nature of such particles.$^{42}$

The thermal stability of perovskites has been explored using thermogravimetric analysis (TGA) and differential scanning calorimetry (DSC).$^{16,60}$ For CH$_3$NH$_3$PbI$_3$, a transition occurs from tetragonal to cubic phase between 55 and 57 °C. Consecutive HI (20%) and CH$_3$NH$_2$ (5–6%) losses are said to occur between 300 and 420 °C. An additional transition between 550 and 600 °C is in good agreement with PbI$_2$. In contrast to CH$_3$NH$_3$PbBr$_3$, however, CH$_3$NH$_3$PbI$_3$ did not display any sign of decomposition under the electron beam during TEM characterization.

**Ensemble Optical Properties of CH$_3$NH$_3$PbI$_3$ and CH$_3$NH$_3$PbBr$_3$.** Figures 11 and 12 show solid-film diffuse reflectance, solution-phase optical absorption, and photoluminescence spectra of several CH$_3$NH$_3$PbI$_3$ and CH$_3$NH$_3$PbBr$_3$ perovskites, respectively. The diffuse reflectance spectra (Figures 11a and 12a) are particularly helpful in determining the absorption onsets of these materials because, due to partial or incomplete solubility (particularly for large and bulk particles), solution-phase absorption spectra of these materials show a
significant amount of scattering at photon energies below their band gap (Figures 11b and 12b). Nevertheless, the solution-phase spectra possess features worth noting. A lack of obvious excitonic peaks in the solution-phase absorption spectra of CH$_3$NH$_3$PbI$_3$ can be attributed to its exciton binding energy (45 meV) being similar to $k_BT$ at room temperature (25 meV, $k_B =$ Boltzmann constant) (Figure 11b). In contrast, the solution-phase absorption spectra of CH$_3$NH$_3$PbBr$_3$ shows clear excitonic peaks for all different morphologies (Figure 12b), in agreement with its higher exciton binding energy (150 eV).52

As expected, bulk CH$_3$NH$_3$PbI$_3$ and CH$_3$NH$_3$PbBr$_3$ samples show absorption onsets of about 800 and 550 nm, respectively (Figures 11a and 12a). The absorption edge remains almost unchanged (<5 nm blue-shift) for CH$_3$NH$_3$PbI$_3$ nanowires compared to bulk CH$_3$NH$_3$PbI$_3$, but is significantly blue-shifted (by about 30 nm) for the aggregated CH$_3$NH$_3$PbI$_3$ dots (Figure 11). In turn, the absorption edge is blue-shifted by about 10 nm for CH$_3$NH$_3$PbBr$_3$ wires and plates compared to bulk CH$_3$NH$_3$PbBr$_3$ (Figure 12). The specific apparent band gaps, measured from the absorption data using Tauc plots ($(A\nu)^2 \text{vs } \nu$ for a direct band gap semiconductor, where $A =$ absorption coefficient, $\nu =$ energy of light) were 1.56 eV for bulk CH$_3$NH$_3$PbI$_3$, 1.57 eV for CH$_3$NH$_3$PbI$_3$ wires, and 1.62 eV for aggregated CH$_3$NH$_3$PbI$_3$ dots; 2.26 eV for bulk CH$_3$NH$_3$PbBr$_3$, 2.30 eV for CH$_3$NH$_3$PbBr$_3$ wires, and 2.30 eV for CH$_3$NH$_3$PbBr$_3$ plates. The mechanisms leading to these shifts are presently unknown; however, the shifts do not appear to originate from quantum confinement, because the crystalline domains (as determined by XRD and TEM, Table 1) are larger than the reported Bohr radii (2.2 nm and 2.0 nm for CH$_3$NH$_3$PbI$_3$ and CH$_3$NH$_3$PbBr$_3$, respectively).61 It is worth noting that some references have attributed a narrowing in the optical band gap of larger organo-metal halide perovskite crystals to changes in the degree of Pb–I bond stress.62

An interesting characteristic of organometal perovskite semiconductors is that they are good fluorophores and can be highly emissive. Given the ease with which these materials perform photoinduced charge separation in high performance photovoltaics, it could appear counterintuitive that they can perform radiative recombination. However, light emission is often observed in materials used for light-emitting diodes and solar cells, as this indicates that nonradiative relaxation pathways are relatively slow.63 For example, GaAs and PbS can exhibit high PL quantum yields and high photovoltages in solar cells.64 In addition, such emission properties may actually be useful in studying the structural and photophysical causes for radiative energy loss in photovoltaic devices, as well as in enabling different applications outside of the solar cell realm, for example in light emitting devices or in single particle imaging and tracking.

Among the different perovskites synthesized here, bulk CH$_3$NH$_3$PbI$_3$ and aggregated CH$_3$NH$_3$PbI$_3$ dots have low PL quantum yields (QYS) of 0.34% and 0.025%, respectively, indicating that nonradiative exciton decay is dominant in these samples. Because

Figure 11. Solid-film diffuse reflectance (a), solution-phase optical density (O.D. includes absorption and scattering) (b), and relative photoluminescence spectra (normalized by each sample’s O.D. at $\lambda_{\text{exc}} =$ 450 nm) (c) of CH$_3$NH$_3$PbI$_3$ perovskites with different size and morphology (a.u. = arbitrary units).

Figure 12. Solid-film diffuse reflectance (a), solution-phase optical density (O.D. includes absorption and scattering) (b), and relative photoluminescence spectra (normalized by each sample’s O.D. at $\lambda_{\text{exc}} =$ 350 nm) (c) of CH$_3$NH$_3$PbBr$_3$ perovskites with different size and morphology (a.u. = arbitrary units).
structural defects are known to play a crucial role in perovskite optical properties, this weak PL intensity could be attributed to surface defects. In agreement with this idea, CH$_3$NH$_3$PbI$_3$ wires and rods made in the presence of the surface passivating n-octylammonium cation have a higher PL QYs of 1.7/$\%$, and their PL lifetimes inversely correlate with their aspect and surface-to-volume ratios (Table 1, see also Supporting Information). In comparison to CH$_3$NH$_3$PbI$_3$, crystalline CH$_3$NH$_3$PbBr$_3$ samples display more intense emission, with PL QYs of 1.2%, 13%, and 0.43% for bulk, wires, and plates, respectively (Table 1).

Single Particle Fluorescence Microscopy: Shape-Correlated Single Particle Emission. To obtain a deeper insight into the photoluminescence behavior of perovskite crystals having different size and morphology, we utilized single particle fluorescence microscopy. To our surprise, different CH$_3$NH$_3$PbX$_3$ nanocrystals sustain synchronous photoluminescence at the single particle level under continuous photoexcitation above their band gap (Figure 13). More specifically, wire-, rod-, and dot-like CH$_3$NH$_3$PbI$_3$ nanocrystals show shape-correlated photoluminescence emission across whole particles, with little photobleaching or photobrightening observed in some cases, and with few off periods (see Supporting Information). This type of behavior, while preceded, is rare. Synchronous blinking (displaying both on and off photoluminescent periods) spanning the entire length of II–VI (CdSe) semiconductor quantum nanowires with 9 nm diameter and $>$5 $\mu$m length was described and explained by a dynamic, transient photolytic filling of surface-trap sites. More typically, 1D nanostructures such as GaAs/Ga$_{1-x}$In$_x$P (0.34 < x < 0.69) and GaAs/GaAsP nanowires with <20 nm diameter display low temperature emission from their tips (only from the nanowire ends). Here, CH$_3$NH$_3$PbI$_3$ nanocrystals with several distinct morphologies behave as nonblinking and photostable single particle emitters at room temperature. These results were corroborated by registered atomic force microscopy-photoluminescence (AFM-PL) experiments, which showed that CH$_3$NH$_3$PbI$_3$ nanowires under steady excitation continuously emit along their length (Figure 14). Interestingly, the single particle PL lifetime measured by AFM-PL is ca. 40 ns, which is roughly similar to the ensemble PL lifetime of 30 ns measured for the same CH$_3$NH$_3$PbI$_3$ nanowire sample (Figure 14 and Table 1, entry 2).

CONCLUSION

Crystalline CH$_3$NH$_3$PbX$_3$ (X = I, Br) particles were synthesized by adding lead(II) halide and methylammonium halide precursor solutions in polar solvents (CH$_3$CN, $\gamma$-butyrolactone, DMF) to a much less polar solvent (toluene, chloroform). CH$_3$NH$_3$PbI$_3$ nanocrystals are particularly unstable under an electron beam, “melting” away to form smaller dot-like particles of unknown composition during TEM analysis.
CH$_3$NH$_3$PbI$_3$ nanowires show PL quantum yields between 1.4 and 1.7% depending on their aspect ratio. CH$_3$NH$_3$PbI$_3$ wires show lifetimes of 30–60 ns, longer than bulk CH$_3$NH$_3$PbI$_3$, which we attribute to better surface passivation by the n-octylammonium cation. CH$_3$NH$_3$PbBr$_3$ nanowires show a PL quantum yield of 13%, significantly higher than that of the corresponding CH$_3$NH$_3$PbBr$_3$ bulk or aggregated dots. Under single particle fluorescence microscopy, wire-, rod-, and dot-like CH$_3$NH$_3$PbI$_3$ nanocrystals show coherent, shape-correlated photoluminescence emission across whole particles, with little photobleaching (or photobrightening) observed and very few off periods. The differently shaped CH$_3$NH$_3$PbI$_3$ and CH$_3$NH$_3$PbBr$_3$ nanocrystals may be of future use in the construction of porous solar cells, and in the study the effect of morphology and crystallization on solar cell performance.

METHODS

Materials. Lead(II) iodide (99%), lead(II) bromide (>98%), methylamine solution (33 wt % in absolute ethanol), N,N-dimethylformamide (DMF, anhydrous, 99.8%), 1,3-pentanediamine (Dytek EP diamine, 98%), γ-butyrolactone (≥99%), n-octylamine (99%), and 1-octadecene (tech., 90%) were purchased from Sigma-Aldrich. Hydroiodic acid (ACS, 55–58%), hydrobromic acid (ACS, 47–49.0%) and oleic acid (tech., 90%) were from Alfa-Aesar; acetonitrile (99.9%), toluene (99.9%), tetrahydrofuran (THF, 99.0%), and acetone (99.9%) were from Fisher. All chemicals were used as received unless specified otherwise.

Synthesis. Ammonium Halides. Hydrogen halides were prepared by a modified literature procedure. Briefly, hydroiodic acid (10.0 mL, 0.075 mol) or hydrobromic acid (8.6 mL, 0.075 mol) was added to a solution of excess methylamine (24 mL, 0.192 mol) in ethanol (100 mL) at 0 °C, and the mixture stirred for 2 h. The solution was concentrated under vacuum, and the resulting powder was dried under dynamic vacuum at 60 °C for 12 h, then recrystallized from ethanol. n-Octylammonium iodide ([CH$_3$(CH$_2$)$_7$NH$_3$]I), n-octylammonium bromide ([CH$_3$(CH$_2$)$_7$NH$_3$]Br), and 1,3-pentanediiodomethane were washed repeatedly with ethyl ether and dried under dynamic vacuum before use.

CH$_3$NH$_3$I was added to a solution of excess methylamine (24 mL, 0.192 mol) in water (240 mL) while stirring. For CH$_3$NH$_3$PbI$_3$ nanodots, the solution was dropped into hexane, THF, 1,4-butanediol (4 mL), and 1,3-pentanediiodomethane (9 mg, 0.025 mmol) were used. CH$_3$NH$_3$PbBr$_3$ nanocrystals were synthesized by dissolving PbI$_2$ (9.2 mg, 0.02 mmol) and CH$_3$NH$_3$I (9.6 mg, 0.025 mmol) in 1,3-pentanediiodomethane (3.2 mg, 0.02 mmol) and mixed with 1,3,5-triiodoacetone (40 mL). Another solution B was made by dissolving CH$_3$NH$_3$I (10.3 mg, 0.04 mmol) in acetonitrile (40 mL). For bulk CH$_3$NH$_3$PbI$_3$, solution A (4 mL) was added to toluene (15 mL) while stirring. For CH$_3$NH$_3$PbCl$_3$, the mixture was added into toluene (15 mL). For aggregated CH$_3$NH$_3$PbI$_3$, the mixture was added into toluene (15 mL). For aggregated CH$_3$NH$_3$PbI$_3$, the mixture was added into toluene (15 mL). For aggregated CH$_3$NH$_3$PbCl$_3$, the mixture was added into toluene (15 mL).

CH$_3$NH$_3$PbBr$_3$ nanocrystals were synthesized by dissolving PbI$_2$ (23.1 mg, 0.05 mmol), CH$_3$NH$_3$I (4 mg, 0.003 mmol) and 1,3-pentanediiodomethane (9 mg, 0.025 mmol) were used. CH$_3$NH$_3$PbBr$_3$ nanocrystals were synthesized by dissolving PbI$_2$ (23.1 mg, 0.05 mmol), CH$_3$NH$_3$I (4 mg, 0.003 mmol) and 1,3-pentanediiodomethane (9 mg, 0.025 mmol) were used.

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Each movie was 25 s in duration and 5 movies were collected per sample. ImageJ was used to analyze the PL intensity versus time for the background and selected particles.

Registered Atomic Force Microscopy-Photoluminescence (AFM-PL) coverslips were mounted on the stage of an inverted optical microscope (Olympus IX71) equipped with a piezoelectric scanner (Physik Instrumente P773.3CD XYZ2) for positioning isolated wire or rod structures in the optical probe region of the microscope. Excitation was provided by a 485 nm pulsed diode laser (PicoQuant LDH-P-C 485B) operating at a pulse repetition rate of either 5 or 10 MHz and pulse width of ~100 ps. The laser light was focused onto the stage with a 1.4 NA/100× oil immersion microscope objective to form an approximately 0.5-μm-diameter optical probe region at the top surface of the coverslip. Average excitation powers were in the range of 2–10 mW. Emitted fluorescence was collected by the same microscope objective and directed onto a single-photon counting avalanche photodiode detector (APD) (PerkinElmer SPCM-AQR). The emission was spatially filtered with a 75-μm-diameter pinhole located in the image plane of the microscope and spatially filtered with a 160 nm bandpass filter centered at 791 nm (Semrock) before reaching the detector. For particularly bright structures, a neutral density (ND 1) filter was placed in front of the bandpass filter to attenuate the photoluminescence reaching the detector. The output of the APD was directed to a time-correlated single-photon counting photon counting module (PicoQuant Picolnarp 300) to record the photon data. The photon detection response was postprocessed with vendor-supplied software (Picoquant Symphotime) to obtain the photoluminescence images. An atomic force microscope (AFM) (Veeco Instruments Bioscope S2), mounted on the stage of the inverted optical microscope, was used to record nanometer scale topography images of the rod and wire structures imaged by the optical microscope. The AFM height images were obtained using Si nanoprobe operated in tapping mode with resonance frequency of approximately 300 kHz. Spatial alignment of the AFM tip with the optical probe region was accomplished by monitoring the excitation laser light scattered from the tip with a second APD. Correlated photoluminescence and AFM height images were obtained by first raster scanning the sample through the optical probe volume to record the photoluminescence image and then engaging and raster scanning the AFM probe to record the AFM image from the stationary sample.

Conflict of Interest: The authors declare no competing financial interest.

Supporting Information Available: Reported XRD patterns for (CH$_3$NH$_3$)$_2$PbI$_4$ and (CH$_3$NH$_3$)$_2$PbBr$_4$; (n = 2.4, 3.5, 4.6), time-resolved photoluminescence data for CH$_3$NH$_3$PbBr$_3$ single particle fluorescence microscopy movies and traces for CH$_3$NH$_3$PbI$_3$ dots, wires, plates), additional e-beam instability data for CH$_3$NH$_3$PbBr$_3$ and IR and XRD vs TEM particle size comparison for CH$_3$NH$_3$PbI$_3$. This material is available free of charge via the Internet at http://pubs.acs.org.

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REFERENCES AND NOTES


