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Anion and Cation Migration at 2D/3D Halide Perovskites Interfaces

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7 **Abstract**

8 This study explores the ionic dynamics in 2D/3D perovskite solar cells, which are known for
9 their improved efficiency and stability. The focus is on the impact of halide choice in 3D
10 perovskites treated with phenethylammonium halide salts (PEAI and PEABr). Our findings
11 reveal that light and heat drive ionic migration in these structures, with PEA⁺ species diffusing
12 into the 3D film in PEABr-treated samples. Mixed-halide 3D perovskites show halide
13 interdiffusion, with bromine migrating to the surface and iodine diffusing into the film.
14 Cathodoluminescence microscopy reveals localized 2D phases on the 3D perovskite, which
15 become more evenly distributed after thermal treatment. Both PEAX salts enhance the
16 performance of photovoltaic devices. This improvement is attributed to the passivation
17 capabilities of the salts themselves and their respective RP phases. Annealed PEAi-treated
18 devices show a better balance between efficiency and statistical distribution of photovoltaic
19 parameters.

20

21 **Introduction**

22 Perovskite-based solar cells (PSCs) with general structure ABX₃, where A = a monovalent
23 cation such as Cs⁺, formamidinium (FA⁺), and methylammonium (MA⁺); B = Sn²⁺ or Pb²⁺; and X
24 = Cl⁻, Br⁻, and I⁻, have seen rapid development in terms of power conversion efficiency (PCE) but
25 are lacking device stability.¹⁻⁴ A common way to improve the ambient stability of devices is the
26 passivation of the active 3D perovskite layer with more stable two-dimensional (2D)
27 perovskites.^{1,5} By simply coating the 3D perovskite film with a solution of an alkyl or
28 arylammonium cation salt, a 2D or *quasi*-2D layered perovskite phase is formed on the surface
29 of the film.^{6,7} The chemical composition of such layered materials is L₂A_{n-1}B_nX_{3n+1}, where L is the
30 organic cation and *n* is the number of inorganic layers of [BX₆]⁴⁻ octahedra (slabs).⁸ In the cases
31 of phenethylammonium iodide or bromide (PEAX) salts, the formed layered perovskites are
32 known as Ruddlesden-Popper (RP) phases. Currently, RP phases are used in most highly-
33 efficient 2D/3D PSCs.^{5,9-12} Because the molecular backbones of bulky L-site organic cations are
34 mostly hydrophobic, the ambient stability of the films towards humidity improves.¹³

35 Although the coating of 3D perovskites with RP phases improves the efficiency and room
36 temperature stability of solar cell devices, several studies have shown that the 2D/3D interface
37 transforms under exposure to light and/or prolonged heat.^{14,15} This contrast is due to the dynamic
38 nature of the 2D/3D perovskite interface. For instance, Sutanto and colleagues showed that

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1 prolonged thermal annealing (50 °C for more than 2 h) of 2D/3D perovskites based on PEAI and
 2 2-thiophene methylammonium iodide (TMAI) on (CsMAFA)Pb(I,Br)₃ caused constant changes
 3 in the structure of the RP phases: in general they degraded over time, and more $n \geq 2$ RP
 4 phases formed ($n \leq 2$ for PEAI and $n \geq 2$ for TMAI).^{6,15} In this case, the device
 5 performance decreased slightly. In contrast, Perini and co-workers showed a faster degradation
 6 of the 2D/3D PSCs (based on (CsMAFA)Pb(I,Br)₃) treated with PEAI and octylammonium
 7 bromide (OABr), compared to the untreated devices; the stability test was performed under inert
 8 atmosphere, 1 Sun irradiance, and 55 °C for 1000 h.¹⁴ The authors correlated this lower stability
 9 with changes caused by ionic migration at the 2D/3D interface during thermal annealing and
 10 device operation, including higher halide heterogeneity on the PEAI-treated surfaces.
 11 Furthermore, some reports have shown that the type of halide used in 2D-forming salts can
 12 significantly affect the interface between 2D and 3D materials, and therefore the performance of
 13 the device. One study by Liu *et al.* looked at how neo-pentylammonium halides (neoPAX) affect
 14 FA_{1-x}MA_xPbI₃ perovskites.¹⁶ They discovered that the presence of Cl⁻ anions in neoPACl greatly
 15 affects the crystallization of the RP phases, leading to a mix of $n = 1$ and $n = 2$ phases. The
 16 authors attributed this difference to an interplay between electronegativity and coordination
 17 ability of halide anions. The use of neoPACl led to improved device performance due to effective
 18 chlorine passivation and the creation of an $n = 2$ RP phase, which enhances charge transfer at the
 19 2D/3D interface. On the other hand, Sutanto *et al.* reported that TMA⁺ cations tend to form $n = 2$
 20 RP phases regardless of the halide used, highlighting the influence of the cations' chemical
 21 structure on the formation of RP phases.¹⁷ They demonstrated that treating
 22 [(FAPbI₃)_{0.87}(MAPbBr₃)_{0.13}]_{0.92}(CsPbI₃)_{0.08} perovskites with TMABr resulted in favorable band
 23 bending at the interface, better charge transport, and reduced voltage loss, thus improving device
 24 performance compared to treatments with TMAI and TMAcI.

25 The complexity of 2D/3D interfaces in halide perovskite solar cells and interdependence on
 26 the nature of the L-site cation, halide selection, and 3D perovskite composition, leaves many
 27 unanswered questions. Here we sought to answer the question: “how does the halide choice
 28 impact the 2D/3D interface formation, ion migration, and device performance?” We evaluate the
 29 formation of 2D layered perovskites using PEAI and PEABr salts on top of iodine-only and
 30 mixed-halide 3D perovskites (Cs_{0.17}FA_{0.83}PbI₃ and Cs_{0.17}FA_{0.83}PbI_{2.5}Br_{0.5}, respectively). The
 31 resulting 2D/3D interfaces were investigated before, during, and after prolonged annealing at 65
 32 °C for 30 minutes. A prolonged annealing is selected on purpose, to drive interface dynamics
 33 that could occur during solar cell operation where module temperatures can easily reach 65 °C.¹⁴
 34 *In-situ* photoluminescence (PL) measurements during annealing showed mostly iodine-rich
 35 PEA₂PbI_{4-x}Br_x 2D perovskite with both PEAX salts, although treatment with PEABr led to a
 36 higher Br content in the RP phases. X-ray photoelectron spectroscopy (XPS) showed clear ionic
 37 migration of PEA⁺ and FA⁺ cations and both halide species (Br⁻ and I⁻). Notably, the diffusion of
 38 PEA⁺ cations is exclusive to PEABr-treated samples. FA⁺ cations migrate to the surface of the
 39 2D films upon annealing, causing the formation of $n = 2$ RP phase. In 3D mixed-halide

1 perovskites with PEABr and PEAI treatment, we found an interdiffusion between iodine and
 2 bromine anions, with Br⁻ moving to the surface and I⁻ diffusing into the film. Finally, we
 3 summarized the work with an assessment of solar cell devices based on these 2D/3D films. The
 4 best performance and reproducibility were achieved with annealed PEAI-treated devices for both
 5 pure-iodine and mixed-halide 3D perovskites.

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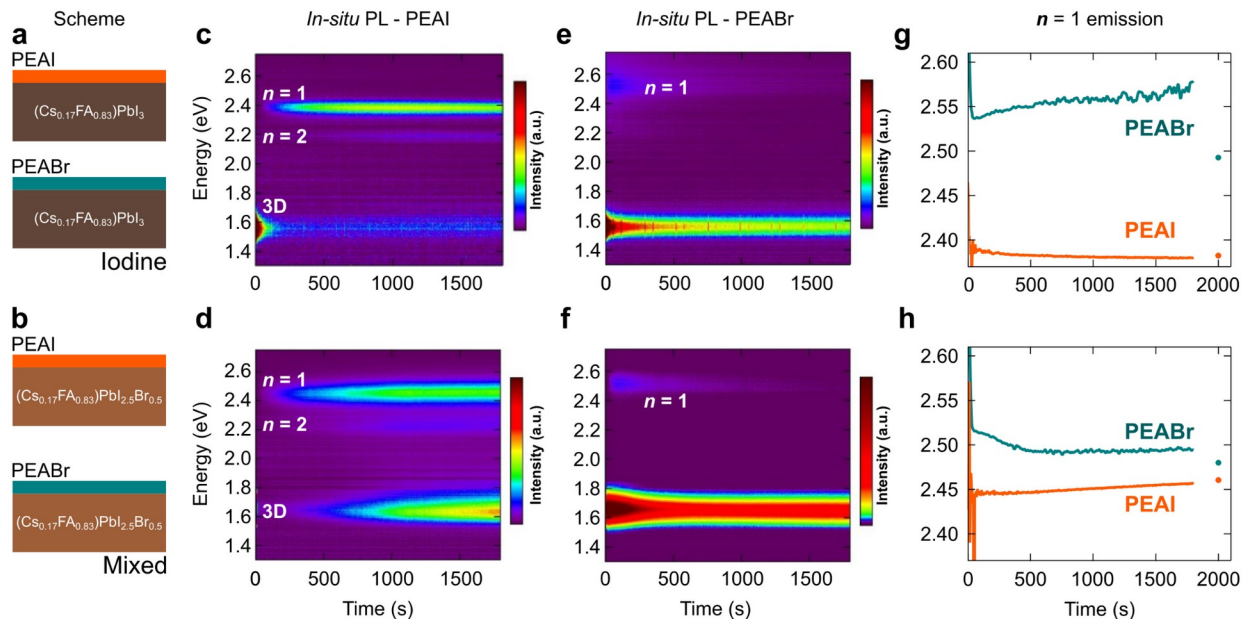
7 **In situ Optical & Hyperspectral Photoluminescence and Structural** 8 **Characterization**

9 3D perovskite films, composed of Cs_{0.17}FA_{0.83}PbI₃ and Cs_{0.17}FA_{0.83}PbI_{2.5}Br_{0.5}, were prepared,
 10 and details of their preparation can be found in the **Methods** section. We will refer to the pure-
 11 iodine 3D perovskite as “*iodine*” and the mixed-halide 3D perovskite as “*mixed*” (scheme of
 12 **Figure 1a,b**). These films were coated with solutions of 2D-forming PEAX salts (X = I or Br) in
 13 isopropanol (IPA). Non-annealed samples are labeled with 0, while the annealed samples have
 14 the number 30 to indicate the 30-minute annealing at 65 °C. Previous results with PEACl
 15 treatment revealed poorer performance of devices, and we disconsidered this salt in this work.
 16 The effect of chlorine and PEACl treatments can be found elsewhere.^{16,18}

17 First, *in-situ* PL ($\lambda_{exc} = 405$ nm, 3.06 eV) measurements during annealing at 65 °C were
 18 performed to track the emission evolution of the RP phases and 3D perovskites (scheme in
 19 **Figure S1**).^{7,19} Around 1.55 eV and 1.65 eV, there is the emission of the bulk 3D *iodine* and
 20 *mixed* perovskites, respectively (**Figure 1c,d** and **Figure S2a,b**). At energies above 2 eV, the
 21 emission of the RP phases is observed. According to literature, $n = 1$ and $n = 2$ emission maxima
 22 for PEA₂A_{n-1}Pb_nI_{3n+1} are at approximately 2.4 eV²⁰ and 2.2 eV,^{6,7} respectively, and for the
 23 bromine analogs at 3.0 eV²¹ and 2.8 eV.²² The *iodine*-PEAI sample presents peaks for both $n = 1$
 24 and $n = 2$ RP phases with emission peak positions in agreement with literature (**Figure 1c**). The
 25 *mixed*-PEAI sample shows the same RP phases with blue-shifted PL peaks (**Figure 1d**); a clear
 26 indication of halide interdiffusion, with bromine incorporation into the surficial RP phases. In
 27 both samples, the PL signal of the RP phases becomes visible in the false color maps after ~120 s
 28 of thermal annealing for $n = 1$ phase and after 500 s for $n = 2$ phase. This delay suggests that
 29 most of the RP phases form upon annealing, rather than during spin-coating, in agreement with a
 30 recent report.⁷ For the *iodine*-PEABr and *mixed*-PEABr samples, the PL signal for the $n = 1$ RP
 31 phase is visible from the beginning of the annealing, but suffers a drastic intensity quench before
 32 500 s (**Figure 1e,f**). The $n = 2$ RP phase is not visible in the *in-situ* PL of both samples. Also, the
 33 emission peaks from the $n = 1$ RP phases are blue-shifted compared to the *iodine*-PEAI analog.
 34 This blue shift is due to the halide diffusion at the interface, leading to the formation of mixed-
 35 halide RP phases with wider bandgap.²³⁻²⁵ We disregard the presence of pure-bromine RP-phases
 36 for reasons that will be clearer further below.

37 Next, hyperspectral PL (HyPL) at $\lambda_{exc} = 405$ nm was carried out, and the hyperspectral
 38 images (**Figure S3**) were integrated to generate the HyPL spectra for each sample (**Figure S4**).

1 With the integrated HyPL, we obtain an averaged emission profile from the whole sample. All
 2 samples show highly homogeneous PL, with less than 1% intensity variation and less than 1 nm
 3 peak position variation for both 3D perovskites (**Figure S3**). Most spectra show three emission
 4 peaks stemming from the bulk 3D perovskite and the two RP phases. The *iodine*-PEAI and -
 5 PEABr samples present peaks for both $n = 1$ and $n = 2$ RP phases with emission peak positions
 6 in agreement with *in-situ* PL (**Figure S4a**). The relative intensity of the $n = 1$ phase increases
 7 after annealing for *iodine*-PEAI. In the *mixed*-PEAI sample, the HyPL intensity of the RP phases
 8 is low, and it is not visible for *mixed*-PEABr (**Figure S4b**). HyPL also shows blue-shifted RP
 9 emission peaks, compared to the *iodine*-PEAI. While the $n = 2$ RP phase is not detectable by *in-*
 10 *situ* PL for the *iodine*-PEABr sample (**Figure 1c**), it is visible in the HyPL data of **Figure S4a**.
 11 This is likely due to HyPL's excitation power being ten times higher than what was used for *in-*
 12 *situ* PL measurements (see **Methods**). It is noted that any emission above 3.06 eV cannot be
 13 accessed with the laser excitation energy used for HyPL and *in-situ* PL measurements. In
 14 addition, HyPL was performed in an ambient atmosphere, in stark contrast to *in-situ* PL, which
 15 was performed on fresh samples in the glovebox (see extended note on the effect of humidity and
 16 oxygen on the PL signal in the **Supporting Information**).



17 **Figure 1. a, b**) Scheme showing the four combinations of 2D/3D perovskites investigated in this work (left side) -
 18 *iodine* samples are at the top row and *mixed* samples at the bottom row. *In-situ* PL color maps taken during
 19 annealing for **c, d**) PEAI-treated and **e, f**) PEABr-treated samples. Gaussian fit to the $n = 1$ RP phase emissions of **g**)
 20 *iodine*-PEAI (orange) and *iodine*-PEABr (dark cyan) and **h**) *mixed*-PEAI (orange) and *mixed*-PEABr (dark cyan).
 21 Separate dots in panels **g**) and **h**) are PL spectra collected during annealing at a fresh spot.
 22

23

24 To gain more insights into the dynamics of the 2D/3D interfaces and laser illumination
 25 effects, we extracted the PL peak positions from the *in-situ* PL data using Gaussian peak fitting

1 of the $n = 1$ RP phases and 3D perovskites (**Figure 1g,h** and **Figure S2c,d**). We observe laser-
2 induced degradation and laser-induced ionic migration in different proportions, depending on the
3 sample's composition. It is well reported that 3D metal halide perovskites with diverse
4 compositions may suffer from several irradiation effects such as halide segregation²⁶⁻³⁰ and
5 photochemical reactions,^{31,32} which can increase halide vacancies and the density of defects.³³
6 Note that these effects can be enhanced or accelerated by the elevated temperature during
7 annealing. Examples of laser effects are given in the **Supplementary Text** and reference 7.

8 Looking at the peak positions of the $n = 1$ RP phases, the *iodine*- and *mixed*-PEAI samples
9 show emissions from PEA_2PbI_4 , at approximately 2.38 eV, and from $\text{PEA}_2\text{PbI}_{4-x}\text{Br}_x$, at 2.46 eV,
10 respectively (**Figure 1g,h**, orange curves). The 0.08 eV increase in the PL peak of the *mixed*-
11 PEAI sample confirms that some bromine from the *mixed* 3D perovskite diffuses into the RP
12 phases (**Figure 1g**). For both *iodine*- and *mixed*-PEAI samples, we observe a nearly constant PL
13 peak position of the RP phases, even on a fresh sample spot during annealing (**Figure 1g,h**,
14 orange dots); this indicates no significant laser effect during measurement, which aligns with the
15 steady PL of the RP phases observed in the color maps of **Figure 1c,d**. The PL position of the
16 RP phases of the *iodine*- and *mixed*-PEABr samples are depicted in **Figure 1g,h**, cyan curves, at
17 approximately 2.55 and 2.50 eV, respectively. The position of their PL peaks also evidences the
18 formation of mixed-halide $n = 1$ RP phases with composition $\text{PEA}_2\text{PbI}_{4-x}\text{Br}_x$ ($x \leq 1.5$). A
19 correlation between the PL emission peak and halide content is possible by taking literature data
20 and extrapolating the bromine content over the reported emission energies (**Figure S5**).^{23,25} Using
21 this correlation we estimate the bromide content to be $0.5 < x < 1.5$. The driving force for this
22 alloying is likely the more negative formation energy of the alloyed phases compared to the
23 pure-halide analogs (*i.e.*, PEA_2PbI_4 and $\text{PEA}_2\text{PbBr}_4$).²⁴ Lastly, the *iodine*- and *mixed*-PEABr
24 samples seem to exhibit an inhomogeneous distribution of bromine within the RP phases. This is
25 highlighted by the shift in the PL peak position when measurements are taken from a new sample
26 spot, as depicted by the cyan dots in **Figure 1g,h**. This inhomogeneity explains the fluctuations
27 in the PL peak maxima in *iodine*-PEABr and *mixed*-PEABr samples. Laser-induced halide
28 migration may also contribute to these differences.

29 Structural characterization through XRD shows a predominance of $n = 1$ RP phases on top of
30 all the samples, with peak position at $2\theta \cong 5.2^\circ$ for Cu K_α radiation (**Figure S6**). Because of
31 the similarity in the peak position for all $n = 1$ RP phases in our XRD data, we cannot tell apart
32 iodine- or bromine-rich phases and, for that purpose, we used the PL peak positions. Another
33 important aspect revealed by XRD is the presence of unreacted salt on the surface of the PEAI-
34 treated samples (**Figure S7**), as previously reported.^{7,11,34} This is consistent with the delayed
35 formation of the RP phases observed in the PL color maps (**Figure 1c,d**). However, there is no
36 evidence of PEABr salt in the XRD from **Figure S6** (a diffraction peak should appear at $2\theta =$
37 5.51°).³⁵ This suggests that PEABr reacts more readily with the 3D perovskites and is nearly
38 completely consumed during the spin coating process which is in agreement with the *in-situ* PL
39 color maps (**Figure 1e,f**). Again, this higher reactivity can be justified by the more negative

1 formation energy of the alloyed RP phases. The difference in formation energy is -3 meV per
2 halide atom for the formation of mixed-halide RP phases, in comparison to their phase-pure
3 equivalents.²⁴ Accordingly, the intensity of the diffraction peaks of the RP phases from the
4 PEABr-treated samples shows higher amounts of the $n = 1$ RP phase on top of the films than for
5 PEAI-treated samples (note that the intensity in this case is comparable - see **Methods** for details
6 and **Figure S8**). Finally, we demonstrate that the RP phase layers are oriented in parallel to the
7 3D perovskite film surface (**Figures S8**). This orientation is also supported by literature.^{16,36} The
8 organic cation bilayer, separating the 3D and RP phases, likely mitigates most strain that could
9 result from the 3D and 2D lattice mismatch.

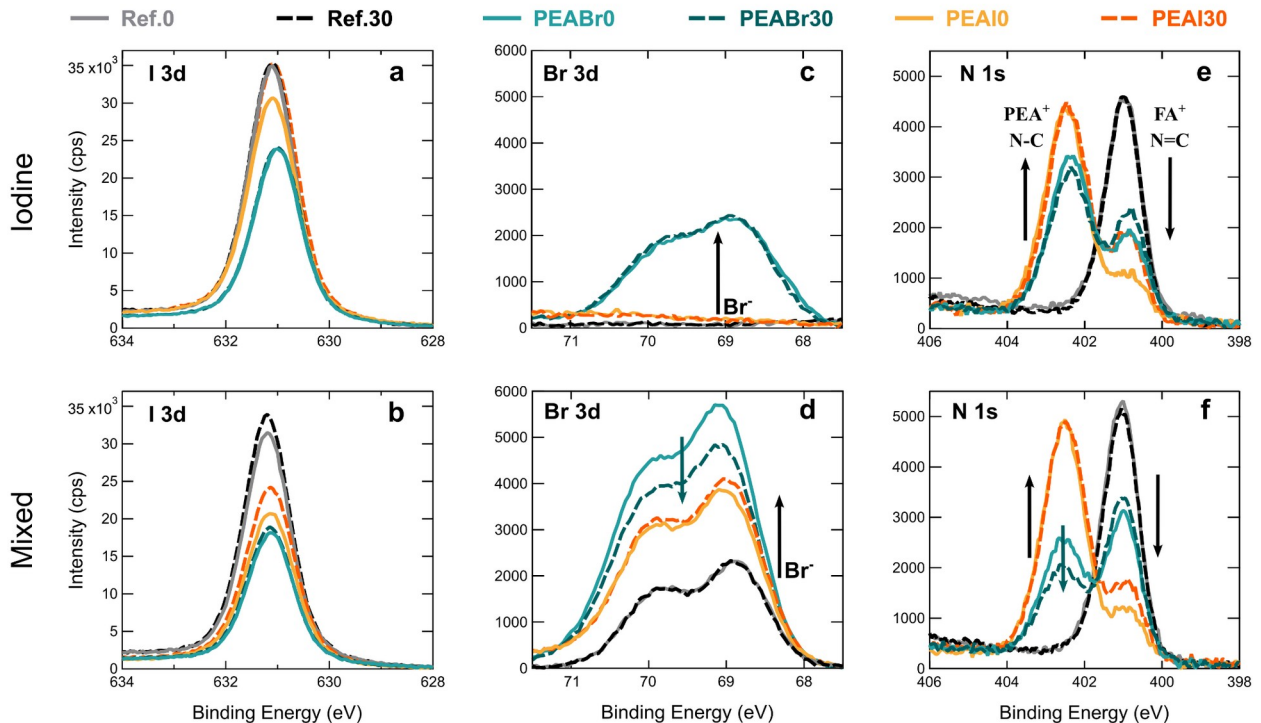
10 The intense diffractions of $n = 1$ RP phase on top of PEABr-treated samples and the low
11 intensity of its emission in the *in-situ* PL maps reveal a difference between XRD and optical
12 characterizations. A possible explanation for this may originate in the optical properties of the
13 mixed-halide, $n = 1$ RP phases. It has been shown that the photoluminescence quantum-yield
14 (PLQY) decreases significantly in tin-based, mixed-halide 2D perovskites.³⁷ This is likely true
15 for lead-based analogs,²² as increased Stokes-shift and FWHM of the emissions are observed in
16 their mixed-halide phases^{23,24} (compare also panels in **Figure 1c** and **1d**). The halide
17 inhomogeneities in these materials seem to increase the non-radiative recombinations and
18 coupling with phonons, leading to a decrease in PLQY.²³ These effects are more pronounced at
19 higher temperatures, which explains the strong thermal quench of the emission observed in
20 **Figure 1c,d**. In addition, there is nearly no signal of the emission from the mixed-halide RP
21 phases in the HyPL spectra (**Figure S4b**). Contrarily, XRD measurements are less sensitive to
22 small amounts or small domains of other phases such as the $n = 2$ RP phases. Although *in-situ*
23 and HyPL show clear emission signals from the $n = 2$ RP phase for basically all samples, there is
24 no evidence of this phase in the XRD (**Figure S6** and **Figure S7** - its (001) and (002) peaks
25 should be close to $2\theta = 4^\circ$ and 8° , respectively).^{38,39} We attribute this divergence to the higher
26 sensitivity of optical measurements to very small amounts of these luminescent materials⁷ and to
27 different penetration depths of X-rays versus visible light. Therefore, optical measurements will
28 be more sensitive to the surface, while X-rays will carry more information from the bulk of the
29 material. The detection sensitivity is even more pronounced in this case due to charge funneling
30 among the RP phases, where even small domains of the lower-bandgap $n \geq 2$ RP phases may
31 function as radiative recombination centers.^{21,40-42} Thus, optical or XRD measurements alone can
32 be misleading to quantify and compare the RP phases on the surface of 2D/3D perovskites.

33 **XPS Measurements**

34 Surface-sensitive XPS measurements were conducted on the reference 3D perovskites,
35 without RP phases, as well as on the 2D/3D stacks before and after annealing to assess the
36 sample composition at the surface (**Figures 2, S9**). **Table S1** summarizes the relative elemental
37 composition on the surface of the films compared to Pb assuming that Pb is the least likely to
38

1 migrate or diffuse in the system under study. After coating the 3D perovskites with PEAX salts,
 2 the intensity of Pb 4f peaks decrease, implying a lower Pb surface content compared to the
 3 reference samples due to the deposition of the salt and formation of the RP phases (**Figure**
 4 **S9a,b**). Except for the PEAI treated samples, the Pb 4f intensity is essentially constant before
 5 and after annealing. Similar decrease in intensity after the PEAX deposition is observed for the
 6 Cs 3d_{5/2} signal (**Figure S9c-d**).

7 The I 3d_{3/2} peak (**Figure 2a**) has an intensity similar to the peaks from the reference *iodine*
 8 perovskite, while the Pb signal decreases (**Figure S9a**). Thus, when PEAI is added, the I/Pb ratio
 9 increases by up to 25% (**Table S1**). Notably, the I/Pb ratio reveals a consistent iodine deficiency
 10 of about 20%, relative to the stoichiometry of both the 3D perovskite (APbI₃, reference) and the
 11 *n* = 1 RP phase (PEA₂PbI₄, *iodine*-PEAI). The treatment with PEAI also leads to a noticeable
 12 increase in the concentration of the carbon signal from PEA⁺ cations, along with a shoulder at
 13 approximately 287 eV associated with the C-N species (**Figure S9e**).^{31,43} This increase is
 14 observed for both PEAX salts. In the case of *iodine*-PEABr, there is a reduction in the intensity
 15 of the I 3d_{3/2} peak compared to other treatment conditions for the *iodine* perovskite. The I/Pb
 16 ratio for these samples indicates an approximate 40% iodine deficiency, assuming a PEA₂PbI₄
 17 _xBr_x stoichiometry. This higher iodine deficiency compared to *iodine*-PEAI samples corroborates
 18 the formation of mixed-halide RP phases.



1 **Figure 2.** Surface sensitive XPS spectra of specific elements before (solid lines) and after (dashed lines) annealing
 2 for 30 min. Annealed and non-annealed samples have the numbers 30 and 0 in their notations, respectively. *Iodine*
 3 samples are at the top row and *mixed* samples at the bottom row. PEAI- and PEABr-treated samples are represented
 4 in orange and cyan colors, respectively. **a,b)** I 3d_{3/2} peaks **c,d)** Br 3d peaks, and **e,f)** N 1s peaks. Ref. denotes the
 5 pristine *iodine* or *mixed* perovskites, depending on the row. Black arrows show the main compositional variation of
 6 each species with respect to the references, and dark cyan arrows emphasize a particular observation in a given
 7 panel.

8

9 For the *mixed* perovskite, we observe a general reduction in I 3d_{3/2} peak intensity following
 10 treatment with both PEAX salts (**Figure 2b**). Notably, the I/Pb ratio for the *mixed*-PEAI samples
 11 shows a slight increase of about 10%, relative to the reference *mixed* perovskite. This aligns with
 12 the iodine migration into the crystalline structure of the *mixed* 3D perovskite, while bromine
 13 migrates to the surface and RP phases. In other words, this halide interdiffusion is the reason for
 14 the 10% rise in surface iodine concentration, compared to a 25% increase in its *iodine*-PEAI
 15 counterpart. The iodine diffusion into the *mixed* 3D perovskite is linked to the red shift observed
 16 in *in-situ* PL (**Figure S2d**). Note that, in the case of the *iodine* samples, the perovskite core levels
 17 exhibit a small shift of roughly 0.1 eV towards lower binding energies after the deposition of the
 18 PEAX salts with respect to the core level position of the perovskite layer without PEAX
 19 treatment. While such a rigid peak shift of all core levels can be indicative of a change Fermi
 20 level position in the gap (*e.g.*, due to the passivation), the absolute value of the shifts is on the
 21 order of magnitude as the measurement error and can hence not be considered as statistically
 22 significant.

23 The Br 3d XPS peaks in the *iodine*-PEABr sample show an obvious increase in bromine
 24 content because of the PEABr treatment (**Figure 2c**). Also, a general increase in the bromine
 25 content is observed with the deposition of both PEAX salts on the *mixed* perovskite (**Figure 2d**
 26 and **Table S1**). This increase corresponds with the trends observed in I 3d_{3/2} XPS (**Figure 2b**),
 27 with bromine migration from the underlying *mixed* perovskite to the surface and iodine diffusing
 28 into the *mixed* perovskite. Similar halide interdiffusion has been reported for mixed-halide
 29 perovskites under visible light illumination.⁴⁴ Intriguingly, despite the higher bromine
 30 concentration, the Br/Pb ratio on the surface of the *mixed*-PEABr decreases by approximately
 31 12% after annealing (see **Table S1** and the dark-cyan arrow in **Figure 2d**).

32 **Figure 2e,f** show two distinct XPS signals for the N 1s level: one, higher in energy, is
 33 associated with the N-C single bond from PEA⁺ cations, which is positively charged (+1), and a
 34 second one related to the N-C double bond (N=C) in FA⁺ cations, where the positive charge is
 35 shared between the two nitrogen atoms (+1/2).^{14,31,43} Interestingly, the PEA⁺ content on the surface
 36 of the PEABr-treated samples is not as high as for PEAI treatment and also not constant upon
 37 annealing. In particular, a decrease in the PEA/Pb ratio by 25% is observed after the annealing of
 38 the *mixed*-PEABr sample (**Table S1**), which suggests a penetration of this species into the film
 39 (**Figure 2f**, dark-cyan arrow). It is well known that anions and cations move in halide
 40 perovskites.^{45,46} For 2D/3D perovskite architectures explicitly, it was found that the bulky cation

1 PEA⁺ moves into the 3D perovskite film.^{14,47} These bulky cations can possibly migrate along
2 grain boundaries, assisted by mobile A-site point defects, for example.^{48,49} A transmission
3 electron microscopy TEM study by Lee *et al.* supports the migration of PEA⁺ species across
4 grain boundaries.⁴⁸

5 As we see a 12% decrease in the Br/Pb ratio in **Figure 2d**, we also speculate that Br is
6 migrating into the 3D film. This migration possibly also occurs through the grain boundaries, as
7 we do not observe an increase in the bandgap of the *mixed* 3D perovskites (**Figure S2d**). This
8 hypothesis is supported by the work of deQuilettes *et al.*, where the authors show higher
9 concentration of bromine at the grain boundaries upon hexylammonium bromide treatment, as
10 revealed by scanning-TEM energy-dispersive X-ray elemental maps.⁵⁰ Migration at the grain
11 boundaries have been reported to be the dominant mechanism in metal halide perovskites due to
12 a much lower activation energy compared to the migration in grain interiors.^{51,52} The migration of
13 Br through the grain boundaries may also be assisted by X-site point defects. Note that the Pb
14 content is constant before and after annealing, as seen in both ratios PEA/Pb and Br/Pb (**Figure**
15 **S9b**). Carbon 1s and Cs 3d_{5/2} XPS signals corroborate this interpretation.

16 In agreement with the N 1s peak, the C 1s XPS signal for PEABr-treated samples does not
17 increase as much as for PEAI-treated ones after deposition of the salts (**Figure S9e,f** and **Table**
18 **S1**). It also shows a decrease in the surface concentration after annealing (dark-cyan arrow in
19 **Figure S9f**). In addition, the deposition of PEABr does not impact the XPS signal intensity of
20 the A-site cations (FA⁺ and Cs⁺) to the same extent as the PEAI deposition does (**Figure 2e,f** and
21 **Figure S9c,d**, respectively). In fact, the FA⁺ XPS signal intensity is even higher than the PEA⁺
22 signal in the *mixed*-PEABr sample (**Figure 2f**). These observations support the hypothesis of
23 diffusibility of PEA⁺ into the 3D layer through the grain boundaries of the perovskite films. This
24 phenomenon is only observed for PEABr-treated samples and is more evident on the *mixed* 3D
25 perovskite.

26 Finally, the N 1s XPS also provides insights into the dynamics of the FA⁺ cations. Initially,
27 the FA/Pb ratio decreases due to the deposition of the PEAX salts. However, after annealing, its
28 intensity is partially restored in all cases (see last two columns in **Table S1**). This restoration of
29 FA⁺ on the surface corroborates the formation of *n* = 2 RP phase (*i.e.*, PEA₂APb₂X₇, A = FA⁺ and
30 Cs⁺) which is observed in the *in-situ* and HyPL data (**Figure 1c,d** and **Figure S4a,b**,
31 respectively). Furthermore, the distinct formation times of the *n* = 1 and *n* = 2 RP phases indicate
32 that the *n* = 1 phase forms initially and, subsequently, the A-site cations migrate from the 3D
33 perovskite to form the *n* = 2 RP phase. Cs⁺ may also be restored to the surface (**Figure S9c,d**)
34 and can be partially incorporated in the *n* = 2 RP phases. With the Cs incorporation, the
35 formation of the mixed-cation PEA₂FA_{1-x}Cs_xPb₂X₇ is more likely due to the low concentration of
36 Cs relative to FA cations in the 3D perovskites. For the *mixed*-PEAI, we only detect this *n* = 2
37 phase in its HyPL spectrum with low intensity (**Figure S4b**). Similar diffusion of FA⁺ cations has
38 been previously reported and associated to partial substitution of PEA⁺ with FA⁺ cations in

1 PEA₂APb₂X₇ RP phases.¹⁴ This is also possible in our case, for both $n = 1$ and $n = 2$ RP phases,
2 and it is likely happening concomitantly with the $n = 2$ RP phase formation.

3 Literature reports discuss A- and L-site cation diffusion in 2D/3D perovskites. Kamat and
4 colleagues provide evidence of A-site cation migration from 3D to 2D perovskites.^{53,54} Even in
5 physically paired films, the $n = 1$ RP phase film partially converts to $n = 2$ RP phase via A-site
6 cation migration.⁵³ This migration is influenced by the L-site cation structure, with
7 butylammonium (BA⁺) showing a more favorable migration barrier than PEA⁺. L-site cation
8 migration was unclear, but Seetharaman *et al.* reported evidence of such migration from RP
9 phases in Sn-based 3D perovskites, also in physically paired films.⁵⁵ They also observed less
10 favorable migration of PEA⁺ cations compared to alkylammonium analogs, with PEA⁺ migration
11 occurring only under thermal and irradiation conditions due to increased steric hindrance. L-site
12 migration in Pb-based 3D perovskites may be more difficult due to its lower thermal expansion
13 and less active Pb 6s² lone pair, responsible for local and long-range disorder in metal halide
14 perovskites.^{56,57}

15 In summary, our XPS data for PEABr treated samples suggests PEA⁺ diffusion into the
16 underlying 3D perovskite film. Diffusion of these species has been demonstrated in several
17 reports and for different cations such as octylammonium and butylammonium.^{14,47,53,54,58} The lack
18 of such diffusion in PEAI-treated samples, therefore, is an exception to what is generally
19 observed. The reason behind this different behavior of PEAI- and PEABr-treated samples is
20 possibly the lower reactivity of PEAI on the surface and its tendency to crystallize as unreacted
21 salt (**Figure S7**). In addition, as shown in the XPS data, the iodine deficiency on the surface of
22 the films may contribute to a prompt fixation of PEAI salt and related RP phase on the surface.

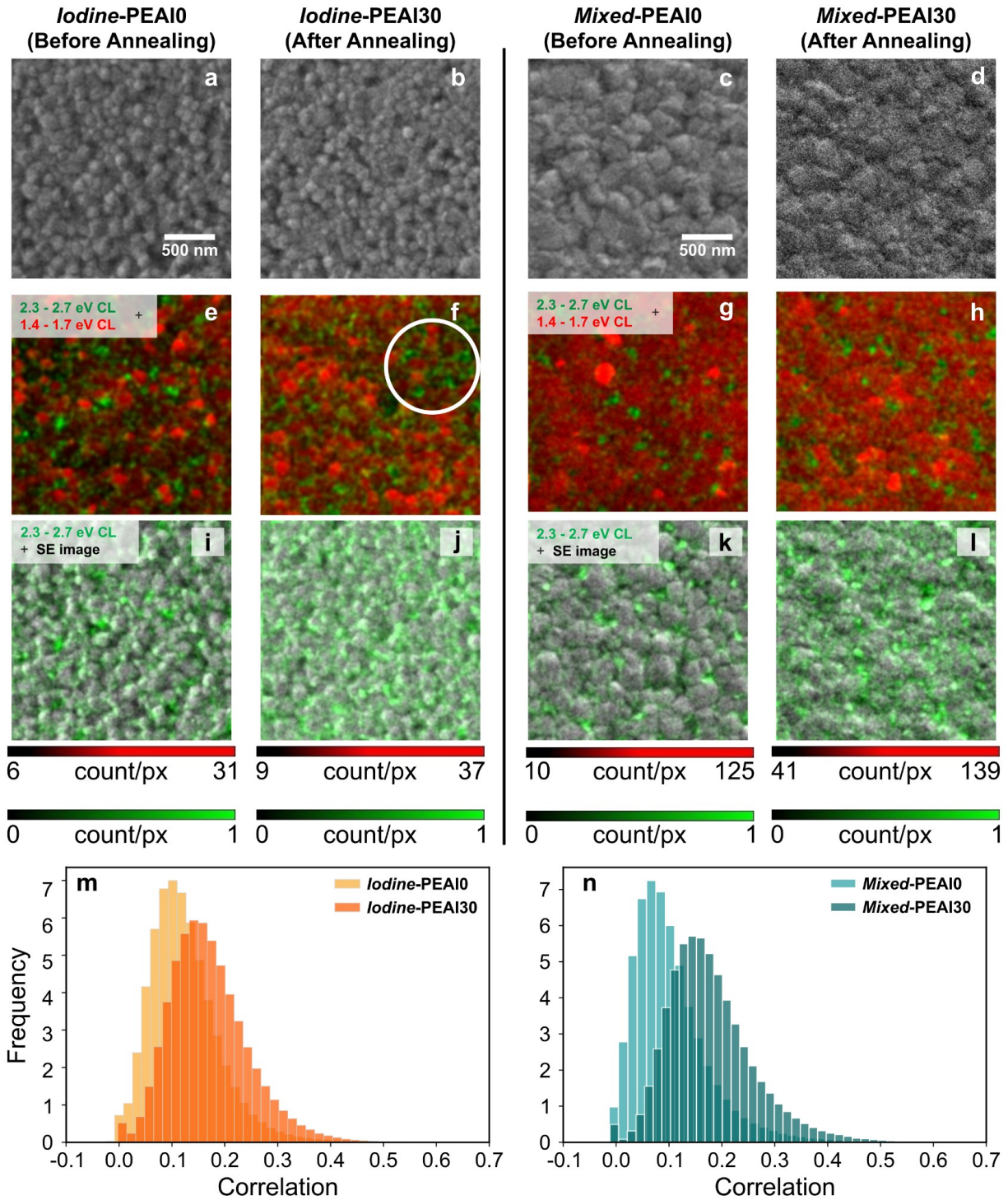
23

24 **Cathodoluminescence**

25 To analyze the radiative charge-carrier recombination behavior with sub-micrometer spatial
26 resolution we carried out cathodoluminescence (CL) experiments on two selected samples,
27 *iodine*-PEAI and *mixed*-PEAI, before and after annealing. The major difference between CL and
28 PL is that charge carriers are excited *via* electrons instead of photons. Two bandpass filters were
29 selected for the measurements: 2.3 - 2.7 eV (458 - 543 nm) for the RP phases and 1.4 - 1.7 eV
30 (735 - 900 nm) for the 3D perovskites. It is noted that the higher energy bandpass filter also
31 detects PbI₂ emission in addition to the RP phases' emission. The energy deposited by the
32 electron beam at 3 kV extends roughly 90 nm into the 2D/3D stack.⁵⁹ More details are provided
33 in the **Methods** section. The secondary electron (SE) images (**Figure 3a-d** and **Figure S10**)
34 show larger grain sizes for the *mixed* samples compared to the *iodine* samples. The morphology
35 and grain size do not change significantly during annealing. Representative CL images from the
36 2D/3D perovskites are depicted in **Figure 3e-l**, where red and green colors are assigned to the
37 emissions from the 3D perovskite and RP phases, respectively. It is speculated that the bright
38 green is related to PbI₂ emission, and the more diffuse green is related to RP emission.

1

1 Starting the discussion with the CL emission representative of the 3D perovskite films, it is
2 apparent that the *iodine*-PEAI sample shows significant intensity variation (shades of red) before
3 annealing and less variation after annealing (**Figure 3e,f** and **Figure S10**). These bright grains
4 are possibly caused by differences in crystal quality where bright domains correspond to



5

2

1 **Figure 3. a-d)** Scanning electron microscopy, secondary-electron images of the *iodine*-PEAI and *mixed*-PEAI
 2 samples before and after 30 min annealing at 65 °C. Overlaid cathodoluminescence images of the 2D/3D films with
 3 different bandpass filters: 2.3 - 2.7 eV for the RP phases (green) and 1.4 - 1.7 eV for the 3D perovskites (red): **e)**
 4 *iodine*-PEAI0; **f)** *iodine*-PEAI30 - white circle emphasizes a region with good coverage of RP phases; **g)** *mixed*-
 5 PEAI0; and **h)** *mixed*-PEAI30 samples. CL image overlaid with secondary electron (SE) image for **i)** *iodine*-PEAI0;
 6 **j)** *iodine*-PEAI30; **k)** *mixed*-PEAI0; and **l)** *mixed*-PEAI30 samples. Annealed and non-annealed samples have the
 7 numbers 30 and 0 in their notations, respectively. Image sizes: 2.5 x 2.5 μm^2 (128 x 128 px²). Pixel dwell time: 20
 8 μs . **m) - n)** LPIC histograms before and after annealing for *iodine*-PEAI and *mixed*-PEAI sample, respectively, with
 9 2.3 - 2.7 eV bandpass filter.

10

11 comparatively low nonradiative recombination rates.⁶⁰ Another explanation are inhomogeneities
 12 in the 3D perovskite composition due to some initial degree of phase segregation before
 13 annealing. For *mixed*-PEAI, halide segregation may be the reason for the red brighter spots, as
 14 the charge carriers are funneled to the iodine-richer phases with lower bandgap and recombine
 15 from these centers.^{29,30,61} After annealing, the emission of the 3D perovskites is more
 16 homogeneous in both cases, and the overall intensity of their emissions increases with annealing
 17 (**Figure 3f,h**). This observation corroborates the hypothesis of PEAX passivation, as the RP
 18 phases spread through the grain boundaries and surfaces (white circle in **Figure 3f**).

19 Next, CL data from **Figure 3i-l** depicts emission from the 2.3 - 2.7 eV bandpass filter
 20 corresponding to the RP and/or PbI_2 phases overlaid with the SE images. It is worth mentioning
 21 that the emission of PbI_2 is highly intense in CL and, thus, it contributes to the CL signal when
 22 using the band filters for the RP phases (PbI_2 bandgap = 2.34 eV).⁶² Additionally, all 3D films
 23 were fabricated with PbI_2 -rich precursors. Initially, the emission is more concentrated at the grain
 24 boundaries, suggesting a more appreciable reaction of the PEAX salts with the excess PbI_2 on the
 25 surface of the films or at the grain boundaries' dangling bonds (**Figure 3i,k**). After annealing, the
 26 emission of the RP phases becomes more homogeneously distributed (**Figure 3j,l**) but intense
 27 green spots are still visible, which we speculate to be from excess PbI_2 . This effect is more
 28 pronounced for *mixed*-PEAI. Here, it is important to differentiate microscopic heterogeneity
 29 (seen in CL) from macroscopic homogeneity (seen with HyPL). These findings are consistent
 30 with the mechanism extracted from *in-situ* PL: initially, we form $n = 1$ phases with the reaction
 31 between PEAI and PbI_2 , and then the reaction continues with the 3D perovskites, with A-site
 32 incorporation into the RP phases.

33 To perform a more quantitative analysis, we measured the CL signal in different regions of
 34 the sample and used a local pixel intensity correlation (LPIC) function (**Figure S11 and S12**).
 35 The LPIC function subtracts the intensity of the pixel from the average of the eight neighboring
 36 pixels to indicate how the intensity of each pixel correlates to its neighbors (see **Methods**). Thus,
 37 the LPIC function provides a quantitative metric for the degree of homogeneity of the overall
 38 emission in the form of a histogram as a function of the degree of nearest neighbor correlation.
 39 Histograms of the LPIC function are depicted in **Figure 3m,n**. In both cases, the LPIC function

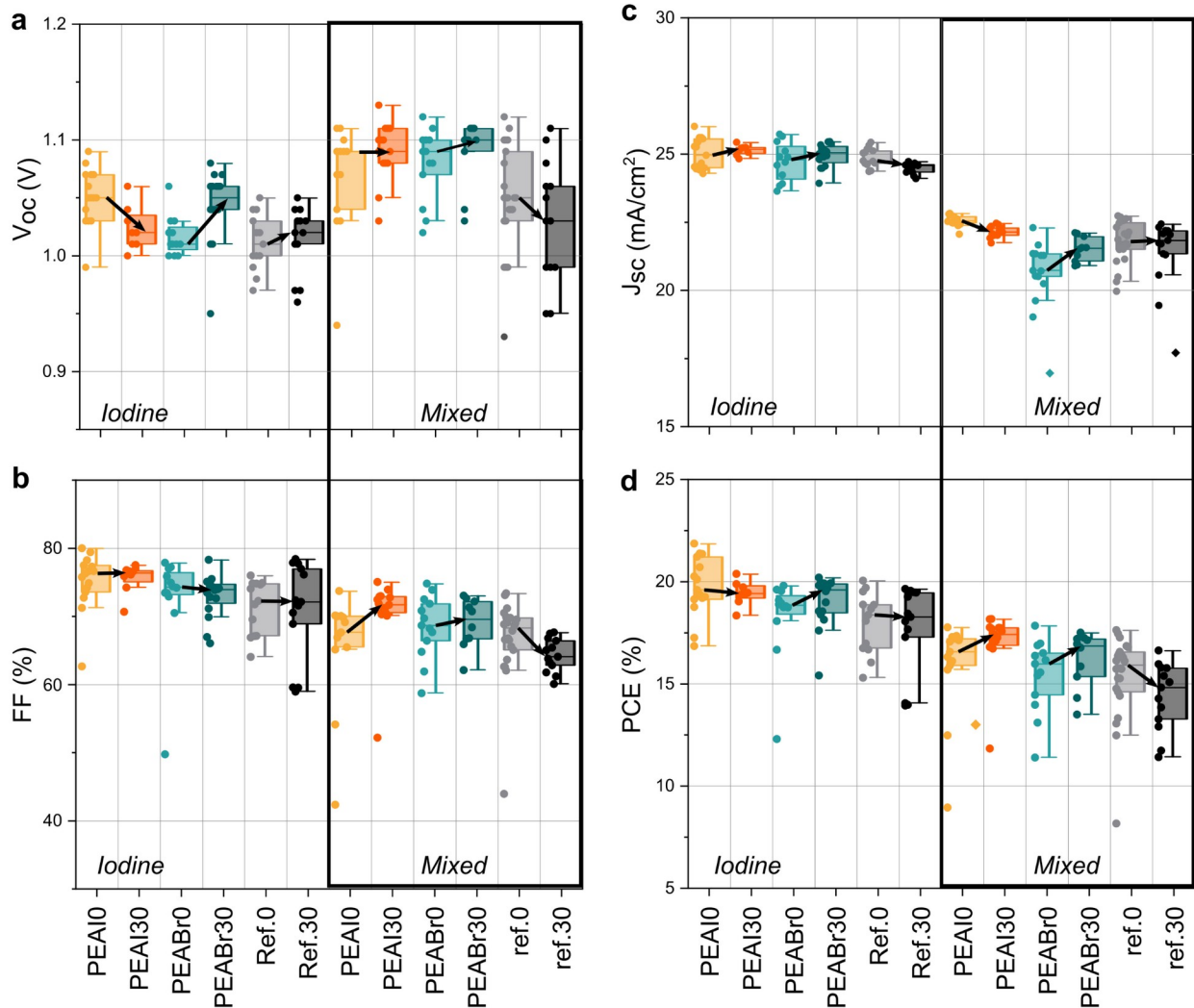
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1 confirms quantitatively what we observe qualitatively in the CL images of **Figure 3i,l**: upon
2 annealing, the emission from the RP phases is more homogeneously distributed over the films.

3

4 **2D/3D Perovskite Solar Cells**

5 We evaluated the impact of different 2D/3D interfaces on device performance by assembling
6 solar cells as outlined in the **Methods** section. The architecture of the devices was FTO|TiO₂-c|
7 TiO₂-mp|3D-Perovskite|RP-phases|Spiro|Au, with reference devices lacking the RP phase layers.
8 The device performance metrics are shown in **Figure 4a-d** (reverse) and **Figure S13** (reverse
9 and forward J-V scans).



10

2

14

1 **Figure 4. a)** Open-circuit voltage (V_{oc}); **b)** fill factor; **c)** short-circuit current density; and **d)** power conversion
 2 efficiency (PCE) of the 2D/3D and reference PSCs before and after 30 min of annealing at 65 °C. Annealed and
 3 non-annealed samples have the numbers 30 and 0 in their notations, respectively. Solar cell parameters are extracted
 4 from the reverse J-V scans. The shaded area in the boxplots includes the 25th to the 75th percentile of the data.
 5 Median data value is presented as a solid line in the plot. The black arrows show the change in the median values
 6 from devices without to devices with annealing.

7
 8 **Figure 4a** reveals an improvement in open-circuit voltage (V_{oc}) compared to the reference
 9 when treating the 3D perovskites with PEAX salts, especially in the *mixed* perovskite cases. This
 10 can be associated with passivation of near-band-edge defects. The passivation of
 11 monoammonium halide salts is well-reported in the literature.^{11,13,15,63} The most significant V_{oc}
 12 improvements are observed for annealed *mixed*-PEAI and *mixed*-PEABr devices, in comparison
 13 to the annealed reference (60 and 70 mV, respectively). This could potentially be due to the
 14 interdiffusion of halides, which, in conjunction with the formation of the 2D/3D interface, results
 15 in effective defect passivation or improvement of the band alignment. Higher diffusibility
 16 through grain boundaries or reactivity of PEA^+ /PEABr species may contribute to a more
 17 effective passivation of the annealed *mixed*-PEABr (the highest V_{oc} , 1.10 V).

18 Despite the effective passivation offered by both PEAX salts, **Figure 4d** shows higher PCE
 19 for PEAI-treated devices, both pre- and post-annealing, with best PCE of 21.86% for *iodine*-
 20 PEAI0 and 18.18% for *mixed*-PEAI30. This hints to a combined contribution of the unreacted
 21 PEAI salt and related RP phases to device efficiency, in agreement with previous reports on
 22 PEAI and PEA_2PbI_4 passivation.^{11,48,64} Annealed *iodine*- and *mixed*-PEAI samples also showed a
 23 narrower fill factor (FF) distribution and highest short-circuit current density (J_{sc}), indicating an
 24 improved charge extraction (**Figure 4b,c**). A decrease in the V_{oc} of *iodine*-PEAI30 devices is
 25 observed, despite the increase of the V_{oc} in *iodine*-PEAI0. This happens because there is a
 26 difference between passivation via unreacted PEAI salt and the RP phase. Before annealing, the
 27 passivation is mainly due to the unreacted PEAI salt.¹¹ Upon annealing, the passivation is due to
 28 the PEA_2PbI_4 RP phase. However, we also form some $n = 2$ RP phase in the case of the *iodine*-
 29 PEAI30 sample. The passivation of $n = 2$ phase is not as effective as for PEA_2PbI_4 , although it
 30 improves the FF and J_{sc} due to a more effective charge separation and extraction.⁷ Lastly, the $n =$
 31 2 RP phase is also less insulating compared to the $n = 1$.^{65,66} Although there is a complex
 32 interplay and multiple overlapping effects upon 2D passivation, in general we observe
 33 improvement in the device performance and statistical distribution of the photovoltaic
 34 parameters in all annealed PEAI-treated devices.

35 Devices treated with PEABr also showed higher PCE compared to the references,
 36 particularly in annealed samples. Interestingly, prolonged annealing worsened parameters in
 37 pristine *mixed* 3D perovskites devices, likely due to increased halide segregation and structural
 38 defects. PEABr treatment appears to mitigate this degradation, enhancing device performance.

39 In this study, annealed *iodine*- and *mixed*-PEAI devices presented the best trade-off between
 40 efficiency and reproducibility. Different halides and ammonium salts yielded varying results. For

1 instance, Sutanto *et al.*¹⁷ reported superior solar cells using bromide anion for TMA⁺ cations,
2 while Liu *et al.*¹⁶ found that treatment with neoPA⁺ and Cl⁻ outperformed devices treated with
3 neoPAI and neoPABr. Regardless, all 2D perovskites formed were predominantly, if not entirely,
4 iodine-based. Thus, the effect of the halide is influenced by the structure of the L-site cation and
5 the composition of the underlying 3D perovskites. Due to this variability, it is challenging to
6 establish general trends in 2D/3D perovskites fabricated with solution deposition of ammonium
7 salts.

8

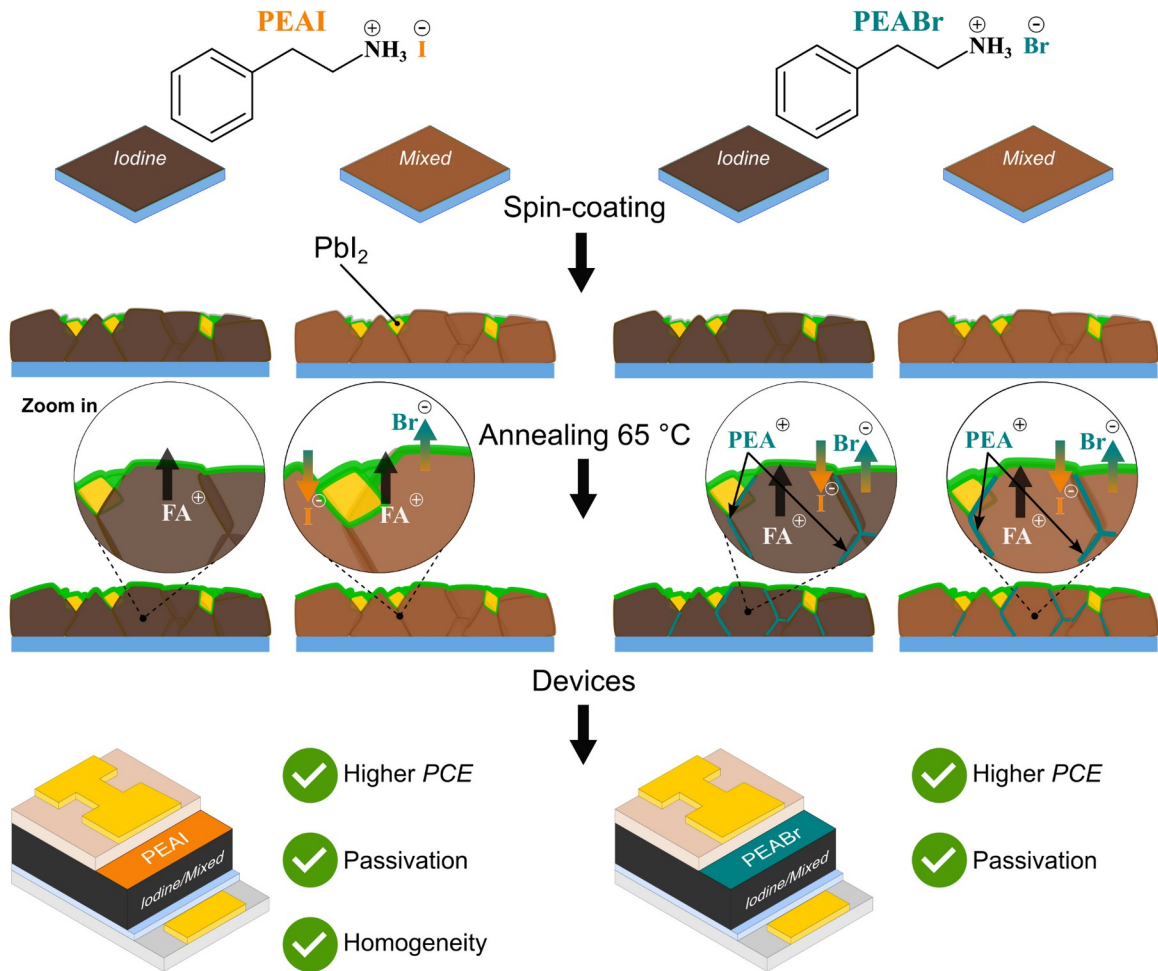
9 **Conclusions and Perspectives**

10 **Figure 5** summarizes our findings regarding the migration at the 2D/3D interfaces
11 investigated here. **(1)** Spin coating of the PEAX dissolved in IPA onto the 3D perovskite films
12 leads to incomplete and heterogeneous coverage of the RP phases in the microscopic scale,
13 although HyPL shows even coverage of the RP phases throughout the film (macroscopic scale).
14 In the case of PEAI, some of the salt remains unreacted even after annealing (**Figure S7**). XRD
15 suggests that the reaction with PEABr is more favorable towards the formation of the RP phases
16 in both *iodine* and *mixed* perovskites. This might be related to a lower formation energy of the
17 mixed-halide 2D perovskite compared to their phase-pure counterparts (PEA₂PbI₄ and
18 PEA₂PbBr₄). **(2)** Upon heat treatment the reaction between the PEAX salts, excess PbI₂/PbBr₂,
19 and the 3D perovskites continues. In all cases, the reaction results in $n = 1$ and $n = 2$ RP phases,
20 leading to a more homogeneous coverage of the 3D layers in the micrometric scale. **(3)** Three
21 migration trends are evident. XPS data reveals that upon annealing, FA⁺ and Cs⁺ cations partially
22 return to the film surface, triggering the formation of $n = 2$ RP phases. It also shows PEA⁺
23 cations penetrating 3D perovskites when films are coated with PEABr, a phenomenon not seen
24 with PEAI. Both, PEA⁺ and bromine appear to diffuse into the film infiltrate as a neutral salt,
25 possibly explaining the low bromine intake by the RP phases and the absence of PEABr salt
26 diffractions in the XRD. Stronger coulombic interactions between PEA⁺ and Br⁻ might play a role
27 in this exclusive PEABr behavior. An interplay between surface iodine deficiency and lower
28 reactivity of PEAI compared to PEAI may be the reason for the lack of PEA⁺ diffusion in PEAI-
29 treated samples.

30 Based on our findings and literature reports on the dynamic nature of the 2D/3D interface,
31 two effects seem to affect the performance stability of devices treated with L-site cations:
32 diffusion of the L-site cations and reconstruction of the perovskite surface, with formation of
33 higher- n phases; and halide mixing at the interface, with diffusion from the 2D layer into the 3D
34 and vice versa. As such, we propose here few strategies that could help enhance the stability of
35 perovskite devices incorporating surface treatments with L-site cations: **(i)** engineering cations to
36 reduce their reactivity with the underlying perovskite;⁶⁵ **(ii)** use cations with mixed halide
37 counterions to avoid halide exchange between the 3D bulk and the 2D surface layer; **(iii)** develop
38 non-halide counterions to pair with the cations used for surface passivation; and **(iv)** develop

1

1 further the use of Dion-Jacobson 2D/3D perovskites, as the doubly-charged L-site cations in this
2 case can slow down their migration under prolonged thermal stress.⁶⁶



3

4 **Figure 5.** Schematics summarizing the migration pathways in 2D/3D perovskite stacks. Different halide
5 compositions, pure-iodine and mixed-halide, 3D perovskite films were combined with PEAI or PEABr (PEA =
6 phenethylammonium) to form the 2D/3D perovskite interfaces. V_A and V_X represent A-site cation and halide
7 vacancies, respectively. See text for details.

8

9 To conclude, we explored how the properties of 2D/3D perovskite interfaces change based
10 on the halide selection in PEAX salts and the halide composition of the underlying 3D
11 perovskite. We found that these interfaces exhibit complex ionic dynamics under thermal stress.
12 Initially, after coating the PEAX salts, the RP phases are distributed unevenly on the surface at a
13 microscopic level, with some residual unreacted salt persisting on the PEAI-treated samples.
14 However, annealing leads to a more uniform coverage, which indicates high ionic mobility at the
15 2D/3D interface. Three main migration behaviors were observed: i) bromine from *mixed*
16 perovskites diffuses to the film's surface while iodine migrates inwardly when treated with
17 PEAX salts; ii) PEA⁺ (associated to PEABr) and PEABr penetrates through the grain boundaries
18 upon annealing, which is not seen with PEAI; and iii) thermal treatment partially restores A-site

2

17

1 cations from the 3D perovskites to the surface, leading to the formation of minority $n = 2$ RP
2 phases. Despite the improved performance and stability of 2D/3D perovskite solar cells, in this
3 work and literature reports, our findings highlight the dynamic nature of these interfaces and
4 validates the importance of thermal annealing to achieve more homogeneous interfaces. Notably,
5 the mobility of PEA⁺ suggests that even bulky cations can diffuse within the 2D/3D interface,
6 potentially affecting the long term stability of devices. Therefore, while 2D/3D perovskites hold
7 promise for stable photovoltaics, optimizing these interfaces remains a challenge. The complex
8 migration behavior and compositional dependence described here suggest that generalized
9 statements about 2D/3D interfaces seem difficult at this point and more studies are needed.

10 Based on our findings and literature reports on the dynamic nature of the 2D/3D interface,
11 two effects seem to affect the performance stability of devices treated with L-site cations:
12 diffusion of the L-site cations and reconstruction of the perovskite surface, with formation of
13 higher- n phases; and halide mixing at the interface, with diffusion from the 2D layer into the 3D
14 and vice versa. As such, we propose here few strategies that could help enhance the stability of
15 perovskite devices incorporating surface treatments with L-site cations: (i) engineering cations to
16 reduce their reactivity with the underlying perovskite;⁶⁷ (ii) use cations with mixed halide
17 counterions to avoid halide exchange between the 3D bulk and the 2D surface layer; (iii) develop
18 non-halide counterions to pair with the cations used for surface passivation; and (iv) develop
19 further the use of Dion-Jacobson (DJ) 2D/3D perovskites, as the doubly-charged L-site cations in
20 this case can slow down their migration under thermal stress.⁶⁸

21 **Supporting information**

22 Schematic of *in-situ* PL measurements; HyPL images and integrated PL curves; XRD
23 measurements, fits, and detector images; additional XPS data; SEM and additional CL images
24 and data; forward and reverse photovoltaic parameters.

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38

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11

12 **Author contributions**

13 **Conceptualization**, C.A.R.P., J.-P.C.-B., C.M.S.-F.; **software/analysis codes**, T.K., S.A.;
 14 **experimentation**, R.F.M., C.A.R.P., T.K., A.K., F.B., N.H. **formal analysis**, C.A.R.P., T.K.,
 15 R.F.M., A.K. N.H., J.H.; **data curation**, C.A.R.P., T.K., N.H., J.H., A.K., R.F.M., C.M.S.-F.;
 16 **writing original draft**, R.F.M., C.M.S.-F.; **writing - review & editing**, all authors have read
 17 and contributed to the draft; **visualization**, R.F.M., C.A.R.P., T.K., C.M.S.-F., **supervision**,
 18 C.M.S.-F.; C.P.S., N.S.G., P.S, J.-P.C.-B.; **funding acquisition**, J.-P.C.-B., C.P.S, C.M.S.-F.

19

20 **Materials and Methods**

21 Phenethylammonium bromide ($\geq 98\%$), phenethylammonium iodide,
 22 anhydrous N,N'-dimethylformamide (99.8%), anhydrous dimethyl sulfoxide
 23 ($\geq 99.9\%$), anhydrous chlorobenzene (99.8%), anhydrous isopropanol
 24 (99.5%), acetylacetone (99%), titanium di-isopropoxide bis(acetylacetonate) (75% in 2-
 25 propanol), anhydrous ethanol (99.9%), TiO₂ paste, bis(trifluoromethane)sulfonimide lithium salt,
 26 FK209Co (98%), and acetonitrile (anhydrous) were purchased from MilliporeSigma and used as
 27 received. Lead iodide (99.99%, trace metals basis - for Perovskite precursor) and lead bromide
 28 ($> 98.0\%$ - for Perovskite precursor) were purchased from TCI America and used as received.
 29 Spiro-OMeTAD was purchased from 1-Material.

30

31 **Perovskite thin film fabrication**

32 Thin films were deposited on 1 in x 1 in, on soda lime glass substrates.

33 **Cs_{1/6}FA_{5/6}PbI₃**: The masses of PbI₂, FAI, CsI precursors were weighed in different vials to
 34 prepare a 1.2 M Cs_{1/6}FA_{5/6}PbI₃ solution with 5 % excess PbI₂. The PbI₂ solution was prepared
 35 first, in DMF:DMSO (2:1) ratio, and stirred at 85 °C for 1h. Once the PbI₂ was fully dissolved,
 36 the PbI₂ solution is transferred to the FAI vial. After the full solvation of FAI, the solution was
 37 transferred to the CsI vial, and the mixture was stirred until complete dissolution of the salt at
 38 room temperature. For the film preparation, 45 μ l of the as-prepared solution was used with a

1 two-step static spin process: 1000 rpm, 1000 rpm/s, for 10s; followed by 6000 rpm, 6000 rpm/s,
2 for 20s. An aliquot of 250 μ l chlorobenzene was dripped onto the spinning film at 28 s from the
3 start of the spin-coating, taking about 1 s to drip the antisolvent. The as-cast perovskite film was
4 subsequently annealed at 150 $^{\circ}$ C for 10 min.

5 **Cs_{1/6}FA_{5/6}PbI_{2.5}Br_{0.5}**: A similar procedure was carried out to prepare a 1.2 M
6 Cs_{1/6}FA_{5/6}PbI_{2.5}Br_{0.5} perovskite solution. The PbI₂, PbBr₂, FAI, and CsI precursors were weighed
7 in different vials. The precursors were dissolved in the following order: PbI₂ \rightarrow FAI \rightarrow PbBr₂ \rightarrow
8 CsI, using DMF:DMSO (2:1) and a 5% excess of PbI₂ and PbBr₂ (mol%). Each solution is
9 transferred to the next vial until complete dissolution of the salts. The same spin and annealing
10 recipe was used as for Cs_{1/6}FA_{5/6}PbI₃.

11

12 **Surface treated films**

13 PEAI and PEABr 20 mM in IPA solutions were prepared (5 mg/ml and 4.05 mg/ml,
14 respectively). 80 μ l were dynamically spun on the perovskite surface at 5000 rpm, 5000 rpm/s,
15 for 20 s (\sim 2 s after the beginning of the rotation), either without annealing, or with annealing for
16 30 min at 65 $^{\circ}$ C right after the deposition.

17

18 ***In situ* PL**

19 The in situ PL measurements were carried out using a home-built setup in an N₂-filled glove
20 box. Excitation was performed using a laser diode emitting at 405 nm and the PL emission was
21 collected using an optical fiber coupled with an Ocean Optics spectrometer (“Flame”) calibrated
22 by the manufacturer. A Jacobian correction was applied to the data transforming them from
23 wavelength- to energy-space and a linear background was removed before fitting the spectra
24 using a combination of one to three Gaussians, depending on the presence of RP phases. The
25 integration time for each spectrum was 500 ms during annealing, with a laser power of
26 approximately 5 mW/cm². Immediately after finishing the surface treatment of the films with
27 PEAX salts, the films were annealed for 30 min at 65 $^{\circ}$ C. Before placing them on the hot plate,
28 the measurements were initialized, and the films were placed into the focal point of the laser on
29 the hotplate.

30

31 **Hyperspectral PL**

32 The hyperspectral PL setup consists of a hyperspectral imager Grand-EOS from PhotonEtc
33 equipped with a CMOS camera (Orca-Flash 4.0 v3 from Hamamatsu) which allows for acquiring
34 large-size images (15x15 cm²) with a resolution of 133 μ m per pixel. Spectral acquisitions were
35 done in the range of 500-1000 nm with a spectral resolution of 2 nm. The samples are uniformly
36 illuminated with LEDs at a wavelength of 405 nm under a fluence of 70 mW/cm². The
37 integration time for the measurements was 1 s. Experimental data were acquired using
38 PHySpecV2 software resulting in a three-dimensional (x,y,l) datacube with both spatial and

1 spectral features. For the images shown in **Figure S3**, intensity maps at the respective
2 wavelength (resolution 2 nm) were extracted.

3

4 **X-ray diffraction (XRD)**

5 XRD measurements were performed in a Bruker AXS D8-Discover X-ray Diffractometer.
6 The source used was a Co K α radiation ($\lambda = 1.7902 \text{ \AA}$). All the diffractograms were reported
7 converting the wavelength to the more commonly used Cu K α radiation ($\lambda = 1.5406 \text{ \AA}$). The
8 measurements were carried out with a Bragg-Brentano geometry with an area detector. The
9 initial incidence and detection angles were 5 and 15°, respectively. The whole measurement was
10 performed with three frames where source and detector increased by 5°/frame each, with ending
11 positions at 15 and 25°, respectively. Because the 2D materials are oriented on the surface of the
12 samples (planes parallel to the film), the integrated XRD profile corresponds to 100% of their
13 diffraction intensity, and so we can roughly say that the amount of RP phases on top of the film is
14 proportional to the diffraction intensity. For this experiment, we set the sample stage to oscillate
15 in a way that the films moved 4 mm in x and 4 mm in y directions while measuring. This way we
16 could have an averaged signal of a larger area of the films. The integration time for each frame
17 was 300 s (total of 900 s for the whole measurement).

18

19 **X-ray photoelectron spectroscopy (XPS)**

20 XPS spectra are measured within a Kratos Axis Ultra system using a monochromatic Al K α
21 source and a hemispherical electron energy analyzer (3×10^{-8} torr). Spectra of various core levels
22 with 0.05 eV resolution and survey scans with 1 eV resolution were acquired. Depending on the
23 signal to noise ratio four to eight sweeps are averaged. Samples were grounded for
24 measurements and did not show charging effects. Slight mismatches in binding energies were
25 corrected using the Pb peak position.

26

27 **Scanning electron microscopy (SEM) and Cathodoluminescence (CL) Imaging**

28 Secondary electron images and Cathodoluminescence (CL) images are acquired with a
29 home-modified Zeiss Gemini SUPRA 55 S2 Scanning Electron Microscope (SEM) at the
30 Molecular Foundry. An aluminum parabolic reflector is positioned above the sample in order to
31 couple a 1.3π sr solid angle of emission into photomultiplier tubes (Hamamatsu, H7360 and
32 H7421-40) using a 458 nm and 543 nm dichroics and subsequent 735 nm long pass filter to
33 separate the color channels into ranges 458 to 543 nm and 735 nm to beyond. All CL images
34 were acquired on a freshly exposed area, by first blanking the electron beam and moving to the
35 region of interest before recording an image. All CL images were acquired at 3 kV and at a beam
36 current of ~ 10 pA with 512 x 512 pixels and a scanning rate of 20 $\mu\text{s}/\text{pixel}$. The total image
37 acquisition per region of interest took 5.24 s.

38

1 **Analysis of CL images**

2 SE images obtained in tandem with CL images collected were analyzed using open-source
3 ImageJ. First, the raw CL images and SE images were Gaussian blurred (2 pixels for 458-543 nm
4 CL channel and 1 pixel for both SE and 735-900 nm CL channel images). The intensity of each
5 image is auto adjusted using imageJ, followed by creating a multichannel composite to combine
6 the adjusted images to provide a comprehensive visualization of the intensity distributions. For
7 detailed CL image analysis, Gaussian filtered 32 bit CL images are used. The pixel intensity
8 histograms and local pixel intensity correlation map were obtained using custom Python scripts.
9 The local pixel intensity correlation map shows the correlation between central pixels and their
10 eight adjacent pixels in a 3 by 3 neighborhood, by subtracting eight times the central pixel
11 intensity from the sum of 8 neighboring pixels. This shows how each pixel's value correlates
12 with its neighbors, capturing sort-range intensity inhomogeneity. The correlation of the pixels on
13 the image boundaries are set to 0.

14

15 **Photovoltaic device fabrication**

16 Substrates of glass measuring 1 in x 1 in with a patterned thin film of FTO (sheet resistance
17 of $7 \Omega \text{ sq}^{-1}$) were meticulously cleaned through a series of steps. The cleaning process included
18 a 15-minute sonication in a 2% Hellmanex solution in deionized-water, followed by sequential
19 rinses with deionized water, acetone, and isopropanol for 10 minutes each in the sonicator . The
20 cleaned substrates were rapidly dried with a nitrogen gun and placed onto a hotplate in
21 preparation for the subsequent deposition of the electron-transport layer. To isolate the active
22 area of the pixels, glass slides were placed along the edges of each substrate, leaving them
23 exposed. The hotplate temperature was set to 450 °C. A layer of compact TiO_2 (referred to as
24 $\text{TiO}_2\text{-c}$) was then applied using spray pyrolysis. This involved a solution made up of 480 μL
25 acetylacetone (Sigma-Aldrich), 720 μL titanium di-isopropoxide bis(acetylacetonate) (75% in 2-
26 propanol, Sigma-Aldrich), and 10.8 mL 99.9% pure anhydrous ethanol (Sigma-Aldrich). Oxygen
27 flowed at a rate of 3.5 L min^{-1} as the carrier gas for the spray. A Sparmax spray gun was utilized
28 for the process, with approximately 10-second intervals between cycles and a 30-second delay in
29 between. The spraying continued until the solution was exhausted. Substrates were then
30 maintained at 450 °C for 30 minutes and allowed to cool down to room temperature.

31 For the mesoporous TiO_2 film (referred to as $\text{TiO}_2\text{-mp}$), a solution of TiO_2 paste (consisting
32 of 30 nm nanoparticles, GreatSolar) in 99.9% pure anhydrous ethanol (Sigma-Aldrich) was spin-
33 coated onto the substrates. Spin-coating was performed at 2000 rpm, with an acceleration of
34 2000 rpm s^{-1} , for a duration of 10 seconds. Magic tape was applied to prevent deposition on the
35 contacts during the spin-coating process. The coated substrates were swiftly transferred to a 100
36 °C hotplate for drying, lasting more than 10 minutes. The mesoporous film was subsequently
37 sintered by gradually raising the temperature to 450 °C and maintaining it for 30 minutes. After a
38 partial cooldown, while the substrates were still around 150 °C, they were moved to a glovebox
39 with oxygen and water concentrations below 10 ppm. The deposition of the perovskite and

1 passivation layer was performed on TiO₂-covered substrates, as described in the section
2 '*Perovskite thin film fabrication*'.

3 A spiro-OMeTAD solution in CB was prepared right before it was deposited to minimize
4 aggregation in solution. This solution contained 28.4 mg of spiro-OMeTAD, 14.5 μL of Co (II)
5 salt (FK209, Sigma-Aldrich, 300 mg mL⁻¹ stock solution in acetonitrile), 8.8 μL of Li-TFSI
6 (Sigma-Aldrich, 520 mg mL⁻¹ stock solution in acetonitrile), and 14.4 μL of tert-butylpyridine
7 (Sigma-Aldrich). The spin-coating parameters were set at 3000 rpm, with an acceleration of
8 3000 rpm s⁻¹, for a duration of 30 seconds. Around 80 μL of the spiro-OMeTAD solution was
9 dripped onto the substrate at approximately 1 second after the spin-coating began. Prior to Au
10 deposition, any CsFAMA and spiro-OMeTAD films present on the substrate edges were
11 removed using cotton swabs moistened with DMF and acetonitrile, respectively. Finally, a layer
12 of gold, approximately 60 nm thick, was thermally evaporated to serve as the top contact for the
13 device.

14

15 **Photovoltaic device characterization**

16 The photovoltaic performance assessment was conducted utilizing a Fluxim Litos Lite
17 arrangement, which featured a Wavelabs Sinus-70 AAA solar simulator that produced an AM1.5
18 spectrum for excitation purposes. The current-voltage (*J-V*) attributes were obtained through
19 both forward and reverse scans, each performed at a scan rate of 50 mV s⁻¹. Stabilized power
20 output measurements were completed utilizing a Maximum Power Point (MPP) tracking
21 algorithm over a span of 60 seconds. Prior to the measurement, the devices were not subjected to
22 any pre-conditioning. To define a pixel area of 0.0625 cm², masking was employed during the
23 measurements. Nitrogen gas was introduced into the measurement chamber throughout the
24 characterization process, and no temperature regulation was implemented.

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