

Intaglio transfer printing enables intrinsically stretchable high-definition quantum dot displays

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This file contains all reviewer reports in order by version, followed by all author rebuttals in order by version.

Version 0:

Reviewer comments:

Reviewer #1

(Comments for the Author)

This manuscript presents a strategy for fabricating stretchable quantum-dot LEDs (QLEDs). However, the work suffers from a fundamental contradiction that severely limits its impact and relevance to the field of stretchable optoelectronics. The impressive performance metrics are critically dependent on the use of a brittle ZnO electron transport layer and are predominantly demonstrated on rigid substrates. This undermines the core claim of presenting a robust platform for intrinsically stretchable displays.

The reported high luminance and efficiency for a “fully stretchable” device are not supported by a genuinely stretchable architecture, as the ZnO layer remains a known point of mechanical failure. Consequently, the manuscript does not offer a sufficient or practical advance for the field. Compounding this issue are unresolved questions regarding the proposed interfacial ligand exchange mechanism, which lacks rigorous experimental validation, and insufficient evidence for the mechanical reliability of the patterned pixels under operational strain.

For these reasons, the work does not meet the high standard of conceptual insight or practical engineering required for publication in Nature Nanotechnology.

Major Criticisms:

1. As claimed by the authors, ligand engineering at the interface is a pivotal advance of their LIFT process. They propose that 3-mercaptopropionic acid (MPA) selectively displaces the elastomer matrix specifically at the surface of the quantum dot (QD) nanocomposite. While the authors employ multiple techniques—including FT-IR spectroscopy, GISAXS, cyclic voltammetry, Mott-Schottky analysis, single-carrier devices, and final QLED performance—to support this mechanism and its role in enhancing hole injection, the presented evidence lacks sufficient rigor to conclusively demonstrate the proposed selective ligand exchange.

2. The emergence of a C=O stretching band in the FT-IR spectra of the treated film is presented as a key piece of evidence. However, this assertion is problematic because C=O stretching vibrations are inherent to all organic ligands used in the system, including the SEBS-g-MA elastomer, the original oleic acid on the QDs, and the introduced MPA. Therefore, the appearance of this band cannot be uniquely attributed to the successful binding of MPA, as it may simply reflect the presence of the other organic components.

3. Furthermore, the preserved interparticle spacing observed via GISAXS before and after MPA treatment is used to argue that only surface QDs are modified. This interpretation is not valid. Given the minuscule size of MPA molecules compared to the QD nanoparticles and the polymer matrix, any ligand exchange would not be expected to cause a detectable shift in the average center-to-center distance between QDs measured by GISAXS. Consequently, this data cannot be used to support the claim that ligand displacement is confined exclusively to the interface and does not penetrate the bulk of the nanocomposite.

4. To convincingly establish the effect of MPA, critical control experiments are absent. The study lacks a control treatment using pure methanol to isolate the effects of the solvent itself on the nanocomposite's morphology and electronic properties. Additionally, to demonstrate the specificity of the thiol group interaction, a comparison with other short-chain molecules

containing different terminal functional groups (e.g., amines, carboxylic acids without thiols) is necessary. Without these controls, the specific role of the thiol moiety in the purported interfacial engineering remains ambiguous and anecdotal.

5. On the point of the proposed mechanism for enhanced hole injection, the authors' explanation requires further scrutiny and stronger evidence. The argument rests on two pillars: the reduced physical barrier due to the short-chain MPA ligand and the elevated HOMO level of the quantum dots (QDs) facilitated by the electron-donating thiol group. However, several critical issues undermine this proposed mechanism. Firstly, the use of Cyclic Voltammetry (CV) to determine the HOMO energy levels of solid-state nanocomposite films is questionable. For such materials, Ultraviolet Photoelectron Spectroscopy (UPS) is the established and more direct method for measuring work functions and ionization energies (HOMO levels). CV measurements in solution may not accurately reflect the electronic structure at the critical solid-solid interface within the device. Secondly, according to the energy level diagram presented in Figure 2h, the hole injection barrier from the HTL (TFB, HOMO ~ -5.3 eV) to the QDs remains substantial. Even after MPA treatment, the HOMO level of the QD composite is reported at -5.89 eV (LIFT) or -5.80 eV (full MPA), implying a persistent injection barrier of approximately 0.5 eV to 0.6 eV. A shift of merely 0.07 eV to 0.16 eV seems inconsequential for effectively overcoming such a high barrier. The authors do not provide a convincing physical model for how such a minor shift could lead to the observed dramatic improvements in device performance. Thirdly, the data from hole-only (HoDs) and electron-only devices (EoDs) further complicate the narrative. The manuscript states that the enhanced hole current in the LIFT-processed HoDs remains "markedly" lower than the electron current in the EoDs, which is largely unchanged. If the hole injection is still the severely limited factor, it is difficult to reconcile this with the achievement of a high EQE of 20.1% in the balanced full QLED structure. Such high efficiency typically requires nearly balanced charge injection, which the presented HoD/EoD data do not convincingly demonstrate. In light of these points, the claimed mechanism of improved charge balance via enhanced hole injection appears insufficiently supported. Therefore, the authors must provide further and more rigorous experimental evidence to clarify the origin of the efficiency enhancement.

6. The authors propose a key mechanism for the LIFT process wherein mechanical strain is locally concentrated at the trench edges via thermal expansion of the PDMS stamp. They support this claim with finite element simulations, suggesting that this localized strain facilitates clean cleavage at the trench boundaries during patterning. This is presented as acting synergistically with MPA-induced interfacial adhesion enhancement to achieve high-definition patterning. However, I have significant concerns regarding the substantiation of this claim and its practical relevance to the final stretchable displays: Firstly, While Figure 3e convincingly demonstrates the superior patterning yield of the LIFT technique on a rigid substrate, this advantage appears severely compromised in the actual stretchable displays shown in Figure 4e-4l. The operational devices exhibit noticeable edge blurring, non-uniform emission, and apparent pixel defects under strain. This observable degradation critically weakens the authors' central claim that the LIFT process enables high-resolution patterning for stretchable displays. The data suggest that the process, while effective for creating patterns on a static, rigid surface, may not fully maintain this precision when integrated into a multi-layer, mechanically dynamic device architecture. To objectively validate their comparison and exclude potential selection bias in the presented images, the authors should provide video evidence demonstrating the patterning process and the resulting pixel integrity across a large array, both before and after deformation.

Secondly, the authors rely solely on a finite element analysis of strain distribution to support the core premise that thermal expansion induces localized cleavage. This analysis lacks critical details and experimental validation. The model is incomplete without a corresponding simulation of the temperature field distribution. The thermal expansion is driven by heat applied via a hot plate; therefore, the thermal conduction through the substrate, the QD composite layer, and the PDMS stamp is fundamental to the proposed mechanism. The absence of this thermal data leaves the mechanical boundary conditions of the strain simulation unjustified. The claim that thermal expansion is the primary driver requires direct experimental evidence. All materials involved (substrate, QD composite, PDMS) will expand upon heating. The critical question is the differential thermal expansion and the resulting stress fields that lead to fracture. The argument is not fully convincing without quantitative, experimental measurements of the stress or strain evolution during the heating process, for instance using techniques like in-situ stress monitoring or digital image correlation. Furthermore, the influence of temperature should be systematically analyzed. A comprehensive analysis of how different temperatures affect the thermal expansion, resultant stress fields, and ultimately the patterning fidelity is absent. Such a study is crucial to establish a reliable process window and to fully validate the proposed thermally assisted cleavage mechanism.

Minor Points and Questions:

1. In Figure 1a, a schematic illustration outlines the LIFT strategy, detailing its four key fabrication steps. To allow readers to fully comprehend the process, this schematic must be substantiated with corresponding photographic or microscopic evidence. The following visual data would be essential for clarity: Optical or PL images, alongside SEM micrographs, of the PDMS stamp and the ODTS-treated donor substrate during and after the QD composite pickup step; PL images of the PDMS/QD composite sample following the interfacial ligand engineering with MPA, demonstrating any changes in surface properties; PL and SEM images of the residual QD composite remaining on the PDMS stamp after the thermally assisted pattern formation on the intaglio trench; PL and SEM images of the PDMS stamp after the final pattern release, confirming the successful and clean transfer of the QD patterns.

2. Figure 1i shows PL images of high-resolution full-colour QD pixel arrays under planar and convex deformations. The image under convex deformation reveals localized pixel failure, notably in the vicinity of the scale bar, which compromises the authors claim about the LIFT strategy.

3. The description of the samples used for the full-colour QD pixel arrays presented in Figure 1j is critically insufficient for a proper evaluation of the results. Specifically, the manuscript fails to provide essential fabrication and structural details, such

as: the precise thickness of the red, green, and blue QD nanocomposite films that constitute the subpixels; the nature and material properties of the substrate onto which these patterned arrays are transferred and imaged.

4. Based on the data presented in Figures 2a-2c, Extended Data Figure 3, and Extended Data Figure 4, the authors claim that the chemical bonding between SEBS-g-MA and the QDs enhances mechanical stretchability. While this conclusion may be valid, the current evidence raises significant concerns regarding its reliability. In Extended Data Figure 3b, the FT-IR spectra comparison is made only between pristine QDs and the QD/SEBS-g-MA composite. The critical control—a spectrum of the QD/SEBS composite—is absent. This omission makes it impossible to definitively attribute the observed spectral changes to the specific chemical bonding between QDs and SEBS-g-MA, as opposed to general physical interactions present in both composites. The AFM images in Extended Data Figure 4b and 4c reveal that the QD/SEBS and QD/SEBS-g-MA composites exhibit fundamentally different surface morphologies even at 0% strain. This suggests a significant disparity in the initial dispersion and distribution of QDs within the two polymer matrices. Consequently, the superior mechanical performance of the QD/SEBS-g-MA composite cannot be ascribed solely to chemical bonding; it may be significantly influenced by this pre-existing difference in nanocomposite homogeneity. To substantiate the authors' claim, Extended Data Figure 5 should include a direct comparison between the SEBS/QD and SEBS-g-MA/QD composites.

5. A significant point of concern is the handling of the “positive aging effect” commonly observed in QLEDs utilizing ZnO electron transport layers. The device efficiency and stability are highly dependent on whether measurements are taken immediately after fabrication or after a stabilization period. The authors must explicitly state the timing of their J-V-L and EQE measurements relative to the device fabrication. Furthermore, to substantiate the performance claims, it is imperative that they provide data characterizing this aging process, such as the evolution of efficiency and luminance over time under constant operating conditions.

Additionally, the manuscript fails to specify the substrate used for the core EL performance evaluation (e.g., Figure 2I-o, Extended Data Fig. 6). The properties of the substrate (e.g., rigid glass vs. flexible/stretchable polymer) fundamentally influence the device structure, processing, and ultimately, the extracted performance metrics. The absence of this fundamental information is a serious omission. This lack of transparency could be perceived as obscuring critical experimental conditions, thereby misleading readers in their interpretation of the device's capabilities and making it difficult to reproduce the work.

6. The authors attribute the superior performance of the LIFT-processed devices, in part, to reduced current leakage compared to spin-coated films (Extended Data Fig. 6d-f). However, this conclusion may not be warranted. Spin-coating is a well-established method for fabricating high-performance QLEDs, with literature reports regularly demonstrating EQEs exceeding 30% and minimal leakage. The significantly poorer performance of the spin-coated control device in this work suggests it may not have been optimally fabricated. A more rigorous comparison would require the authors to demonstrate that their spin-coated device represents a state-of-the-art control, rather than a sub-optimal one. Without this, the claim that spin-coating inherently causes serious leakage in this context is not fully supported and weakens the credibility of the comparative analysis.

7. Extended Data Fig. 8's analysis of the composite's mechanical stretchability, which employs a rigid ITO electrode on a non-stretchable substrate, appears misaligned with the study's objectives. To properly evaluate stretchability, the test structure should mirror the proposed flexible device architecture from Figure 4. Additionally, the experimental details regarding the integration of the pre-stretched EML into the rigid device are unclear. If the EML is pre-strained on PDMS (e.g., at 0%, 50%, and 100%) before device fabrication, this approach does not accurately represent the operational stability of a stretchable device and could undermine the reported advantages.

8. The authors' successful optimization of the EML's stretchability stands in stark contrast to the neglect of other critical device components, most notably the electron transport layer. This represents a fundamental limitation of the presented strategy. The reported high luminance and EQE are achieved in a device architecture that remains critically dependent on the ZnO nanoparticle layer, which is intrinsically brittle and lacks any meaningful stretchability. Consequently, the superior performance metrics are primarily demonstrated on rigid glass substrates, which severely undermines the practical relevance and claimed applicability of this research for genuinely stretchable LEDs.

Reviewer #2

(Comments for the Author)

The proposed LIFT strategy is innovative, combining ligand engineering with thermal transfer printing. This approach presents a well-conceived solution to address current limitations in pixelation and performance of stretchable quantum dot displays, likely to have a positive impact on the development of next-generation display technologies. However, several critical points need to be addressed before the manuscript can be considered for publication. I recommend the following revisions:

1. It is suggested that the authors include cross-sectional SEM images of the intrinsically stretchable QLEDs employing AgNWs as the anode. These types of images will help readers better understand device architecture.
2. The authors attribute the enhanced mechanical robustness and stable resistance during stretching to partial alloying at the Ag/Au/LM interface. However, the conclusions drawn from a single tensile test are insufficient. It is recommended to include cyclic tensile stability tests to evaluate the long-term mechanical reliability of the device under repeated deformation.
3. In Figure 1i and Extended Data Table 1, the authors compare the performance of their device with other intrinsically stretchable light-emitting devices. However, PeLEDs are not depicted in Figure 1i but instead illustrated in Extended Data Table 1, which creates confusion. Additionally, a more thorough review of relevant literature should be provided to contextualize the significance of this work.

4. In Figure 4d, the strain-luminance-current density curves demonstrate trends that differ significantly. The luminance improves by more than 1.5 times with increasing strain but decreases to 0.6 times at a strain of 70%. This behavior requires a detailed explanation in the manuscript. Furthermore, it would be valuable to report the changes in luminance after strain release to assess the reversibility of performance.
5. To confirm the mechanical stability of the device at each strain level, it is recommended to include data on mechanical reliability under repeated cycles of deformation.
6. The authors report fabricating monochromatic pixel arrays with resolutions of up to 16,000 pixels per inch and RGB subpixel arrays with 2,565 pixels per inch. However, these values are similar to those reported in their previous work (Nature Photonics, 18(10), 1105-1112), where RGB pixelated patterns with 2,565 pixels per inch and monochromatic QD patterns with approximately 20,526 pixels per inch were achieved. Is it feasible to achieve the same or higher pixel density? If not, what are the associated challenges?

Reviewer #3

(Comments for the Author)

In this work, Yoo et al. report high-performance, high-resolution, intrinsically stretchable QLEDs using QD/elastomer composites as the emissive layer and intaglio-transfer printing for patterning. The patterning resolution exceeds 10,000 ppi for single-color QD stripes and >2,000 ppi for full-color QD patterns. Key device metrics—turn-on voltage, luminance, and EQE—surpass those reported for stretchable OLEDs, OLECs, and QLEDs. As the authors argue, their materials design and patterning strategy address a central challenge in intrinsically stretchable LEDs: the long-standing trade-off between high brightness and high patterning resolution.

In 2024, some of the same authors published a related study in Nat. Electron. (2024, 7, 365) employing a similar materials combination (QDs and SEBS-g-MA). In my view, the major differences between the present submission and the Nature Electronics (2024) work are as follows. (1) Materials/composite design. Here, the authors perform a post-deposition ligand exchange, replacing SEBS-g-MA at the QD film surface with mercaptopropionic acid (MPA). Substituting the long SEBS-g-MA chains with short MPA ligands at the interface lowers the hole-injection barrier and supports the improved QLED performance. By contrast, the Nature Electronics (2024) study employed a ternary QDs/SEBS-g-MA/TFB composite to tackle a similar hole-injection issue. (2) Patterning method and resolution. MPA at the QD interface appears to enhance adhesion of intrinsically stretchable QD films to SiO₂ trenches—likely due to polarity matching—and, combined with localized strain during thermal-assisted transfer printing, enables high-definition patterning of stretchable QD layers. The earlier Nature Electronics (2024) work used conventional transfer printing, which may be less suited to pattern elastic QD/polymer composites at very high resolution. (3) Device performance. The present stretchable QLEDs achieve record-level luminance and current efficiency among stretchable electroluminescent devices based on both inorganic and organic emitters.

On this basis, I consider the manuscript a significant advance for intrinsically stretchable QLEDs with implications for next-generation displays conformable to non-developable surfaces. That said, several issues should be addressed to strengthen the claims:

1. The manuscript posits chemical bonding between maleic anhydride/acid groups in SEBS-g-MA and free oleic acid via half-esterification. However, maleic anhydride is relatively stable. Its conversion to the diacid may not proceed readily in the solid state at room temperature. In Extended Data Fig. 3b, the QD/SEBS-g-MA film shows only a marginal increase near ~1200 cm⁻¹, assigned to additional C–O bonds. The authors need to be cautious about this. First, as depicted in Extended Data Fig. 3a, the total number of C–O bonds in the system would not change before or after reaction, if any. Second, the signal-to-noise ratio is low, and the apparent “bump” could reflect noise. Additionally, the authors used the same combination of QDs and SEBS-g-MA in their Nat. Electron. 2024 work without claiming covalent bonding at this interface.

This point does not detract from the demonstrated stretchability or device metrics, but it should be clarified. A simple control—mixing SEBS-g-MA with oleic acid in solution and monitoring by reliable spectroscopy—could test whether the reaction occurs under mild conditions. If it does not proceed in solution at room temperature, it is unlikely to occur efficiently in the solid state.

2. Similarly, FTIR data in Fig. 2e provides limited support for ligand exchange, even though MPA exchange is plausible and well documented. Because SEBS-g-MA contains two C=O groups (or three if it reacts with oleic acid to form the proposed complex in Fig. 1f) whereas MPA has one, the net C=O trend depends on the relative coverage of each species. The manuscript already includes QDs fully coated with MPA (Fig. S7). Using those and QDs coated solely with SEBS-g-MA as FTIR standards would strengthen the assignment and validate the exchange.

3. Fig. 1j is compelling. It is helpful to provide the patterning resolution in ppi in the figure or caption.

4. Extended Data Fig. 1 compares the pros and cons of several manufacturing methods. I The pros/cons ranking for inkjet printing and photolithography merits reconsideration. For example, with bank structures or pixel-defining matrices, inkjet-printed devices can also mitigate charge leakage. Photolithography uses spin-coated QD films whose packing density should be comparable to that achieved in transfer-printed films.

Version 1:

Reviewer comments:

Reviewer #1

(Comments for the Author)

The authors have revised the manuscript in accordance with my comments, and the manuscript is now acceptable for publication.

Reviewer #2

(Comments for the Author)

The revised manuscript is well prepared, and I recommend it for publication in its current form.

Reviewer #3

(Comments for the Author)

In the revised manuscript, the authors have made substantial changes to the work and clarified the issues about the chemistry and technique. The comments or concerns have been properly addressed, validating the key results of this work. This manuscript can be accepted by Nature Nanotechnology.

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Reviewer #1:

Summary Comments: *This manuscript presents a strategy for fabricating stretchable quantum-dot LEDs (QLEDs). However, the work suffers from a fundamental contradiction that severely limits its impact and relevance to the field of stretchable optoelectronics. The impressive performance metrics are critically dependent on the use of a brittle ZnO electron transport layer and are predominantly demonstrated on rigid substrates. This undermines the core claim of presenting a robust platform for intrinsically stretchable displays.*

The reported high luminance and efficiency for a “fully stretchable” device are not supported by a genuinely stretchable architecture, as the ZnO layer remains a known point of mechanical failure. Consequently, the manuscript does not offer a sufficient or practical advance for the field. Compounding this issue are unresolved questions regarding the proposed interfacial ligand exchange mechanism, which lacks rigorous experimental validation, and insufficient evidence for the mechanical reliability of the patterned pixels under operational strain.

For these reasons, the work does not meet the high standard of conceptual insight or practical engineering required for publication in Nature Nanotechnology.

Our response to summary comments: We sincerely appreciate the reviewer’s careful evaluation of our work. However, we respectfully address the concern that the study presents a fundamental contradiction or lacks practical relevance to stretchable optoelectronics.

The key performance metrics reported in this work were obtained from intrinsically stretchable QLED architectures and benchmarked against previously reported fully stretchable light-emitting devices (**Figure 1i, Extended Data Table 1**). The rigid device configuration (**Fig. 2i-o**) was used only to evaluate the intrinsic optoelectronic and mechanical properties of the QD nanocomposite fabricated via the LIFT process. This approach allowed us to isolate the material-level characteristics of the QD nanocomposite from additional variables associated with the stretchable device configuration, which is a common strategy for establishing fundamental material performance. **For a more detailed discussion, please refer to our response to minor comment #5,6,7.**

Importantly, the fully stretchable devices did not rely on a brittle ZnO nanoparticle layer. Instead, a mechanically compliant composite electron transport layer (ETL) consisting of ZnO nanoparticles blended with PFN-Br was employed to ensure mechanical deformability and device stability under strain. While bare ZnO films exhibit pronounced tearing, the ZnO/PFN-Br hybrid ETL shows mild wrinkling and localised crack under comparable deformation (**Supplementary Fig. 31**). We apologize for any lack of clarity in the original manuscript and have revised the text to explicitly clarify the device architecture. **For a more detailed discussion, please refer to our response to minor comment #8.**

We further note that the reviewer’s concerns regarding the interfacial ligand engineering mechanism and the mechanical reliability of the patterned pixels have been carefully addressed through additional experiments and expanded discussions in the revised manuscript.

Overall, this work establishes material and process platform that simultaneously enables high efficiency, high luminance, and high-resolution patterning in an intrinsically stretchable QD nanocomposite system. To the best of our knowledge, this study demonstrates the highest-patterning resolution reported for stretchable light-emitting devices, achieving sub-micrometer feature sizes of 700 nm, while also reaching the highest brightness among fully stretchable light-emitting devices reported to date (**Figure 3j**, **Figure 1i**, **Extended Data Table 1**). Achieving record-high brightness and sub-micrometer resolution concurrently within a fully stretchable architecture represents a significant conceptual and engineering advance toward practical stretchable displays. The manuscript has been revised throughout to clarify these points and to better highlight the conceptual contribution and practical implications of this work.

***Comment #1:** As claimed by the authors, ligand engineering at the interface is a pivotal advance of their LIFT process. They propose that 3-mercaptopropionic acid (MPA) selectively displaces the elastomer matrix specifically at the surface of the quantum dot (QD) nanocomposite. While the authors employ multiple techniques—including FT-IR spectroscopy, GISAXS, cyclic voltammetry, Mott-Schottky analysis, single-carrier devices, and final QLED performance—to support this mechanism and its role in enhancing hole injection, the presented evidence lacks sufficient rigor to conclusively demonstrate the proposed selective ligand exchange.*

***Comment #2:** The emergence of a C=O stretching band in the FT-IR spectra of the treated film is presented as a key piece of evidence. However, this assertion is problematic because C=O stretching vibrations are inherent to all organic ligands used in the system, including the SEBS-g-MA elastomer, the original oleic acid on the QDs, and the introduced MPA. Therefore, the appearance of this band cannot be uniquely attributed to the successful binding of MPA, as it may simply reflect the presence of the other organic components.*

Our response to comment #1 and 2: We thank the reviewer for the insightful comments regarding the rigor of the evidence supporting the proposed selective ligand engineering using 3-mercaptopropionic acid (MPA). In Fourier transform infrared spectroscopy (FT-IR), the newly emerged peak at $\sim 1700\text{ cm}^{-1}$ after MPA treatment is assigned to the C=O stretching vibration of protonated carboxylic acid groups in MPA, which is distinct from the carboxylate-related stretching bands (~ 1540 and $\sim 1400\text{ cm}^{-1}$) associated with surface-passivating ligands, such as native oleic acid or those newly formed in SEBS chains (**Extended Data Fig. 6b**).

To provide more direct evidence for selective ligand exchange using MPA, we performed additional solid-state ^{13}C nuclear magnetic resonance (NMR) and X-ray photoelectron spectroscopy (XPS) analyses. During the selective ligand exchange process, the thiol group of MPA strongly coordinates with the surface Zn atoms of the QDs, forming Zn–S–CH₂ bonds at the QD interface, which is a well-established binding motif for thiol-functional ligands on Zn-rich QD surfaces. Consistent with this mechanism, the solid-state NMR spectrum of the LIFT-composite shows a distinct resonance at ~ 21 ppm, corresponding to the Zn–S–CH₂ moiety of MPA, which is absent in the untreated composite (**Fig. 2e**). In addition, XPS measurements

reveal a clear shift in the sulfur binding energy after LIFT treatment, consistent with the formation of Zn–SR bonding (**Extended Data Fig. 6a**). Together, these results provide direct spectroscopic evidence that MPA is incorporated and chemically bound to the QD surface.

Notably, the spatially confined ligand exchange is confirmed by contact angle measurements, which reveal a pronounced asymmetry between the ligand-engineered surface (91.3°) and the opposite side (107.7°) of the LIFT-composite film (**Supplementary Fig. 7**). Such asymmetric surface wettability is consistent with localized chemical modification at the exposed interface of the composite. To further quantify the extent of ligand displacement, CHNS-O elemental analysis was performed on untreated composites, LIFT-composites, and QD nanocomposite with a high fraction of MPA engineering (MPA composites) (**Supplementary Table 1**). The relative carbon-to-sulfur atomic ratios confirm partial ligand displacement in the LIFT-composite, consistent with surface-localised chemical modification enabled by the polymer matrix shielding effect. Additional structural evidence for ligand exchange is provided by thickness measurements. After LIFT treatment, the composite film thickness decreases from 37.8 nm to 35.9 nm (**Supplementary Fig. 4**). This reduction is consistent with the replacement of relatively long ligands (i.e., polymeric SEBS-g-MA and oleic acid (~2.0 nm)) with the much shorter MPA ligand (~0.4 nm) at the surface of composite.

Collectively, these complementary spectroscopic, structural, surface, and elemental analyses consistently demonstrate that the LIFT process induces partial and partially confined ligand displacement at the surface region, rather than indiscriminate ligand exchange throughout the bulk. This interfacial ligand engineering selectively reduces both the insulating barrier and the energy-level barrier at the hole transport layer (HTL)/emissive layer (EML) interface, leading to enhanced hole injection and improved charge balance in the device.

To address the reviewer’s concern, the additional experimental results and clarifications have been incorporated into **Fig. 2e**, **Extended Data Fig. 6a, b**, **Supplementary Fig. 4**, **Supplementary Fig. 7**, and **Supplementary Table 1**, and are discussed in the revised manuscript.

Our modification to the manuscript:

Page 8 in the revised manuscript:

“To enhance hole injection into QDs while enabling high-definition pixelation, the elastomeric polymer ligands surrounding the surface QDs at the HTL/EML interface are selectively displaced by MPA during the LIFT process (**Fig. 2d**, **Supplementary Fig. 4**). This selective ligand engineering preserves the bulk mechanical compliance of the composite while improving charge transport across the HTL/QD interface. During this process, the thiol group of MPA strongly coordinates with surface Zn atoms of the QDs, forming robust Zn–S–CH₂ bonds at the QD interface. Solid-state ¹³C nuclear magnetic resonance (NMR) spectroscopy of the LIFT-composite reveals a distinct resonance at ~21 ppm, corresponding to the Zn–S–CH₂ moiety of MPA, which is absent in the untreated composite (**Fig. 2e**). This assignment is further supported by X-ray photoelectron spectroscopy (XPS), which shows clear shifts in the sulfur binding energy after LIFT treatment (**Extended Data Fig. 6a**) consistent with thiol

coordination to the QD surface. Additional evidence of interfacial chemical modification is provided by FT-IR spectroscopy, where a newly emerged C=O stretching band at $\sim 1700\text{ cm}^{-1}$ is observed in the LIFT-composite (**Extended Data Fig. 6b**), which is attributed to protonated carboxylic acid groups of MPA introduced at the interface.”

Page 9 in the revised manuscript:

“Contact angle measurements and elemental analysis corroborate that ligand replacement occurs predominantly at QDs located near the surface of the composite (**Supplementary Fig. 7, Supplementary Methods, Supplementary Table 1**).”

Page 27 in the revised manuscript:

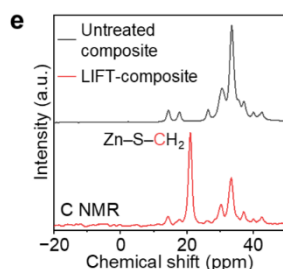
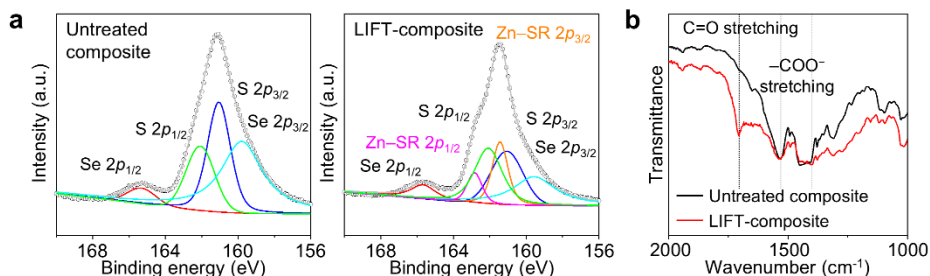


Fig. 2 | e–g, ^{13}C solid-state NMR spectra (e), AFM images (f), and TRPL decay profiles (g) of untreated- and LIFT-composites.

Page 35 in the revised manuscript:



Extended Data Fig. 6 | a, S 2p XPS spectra of untreated (left) and LIFT-composites (right). b, FT-IR spectra of untreated and LIFT-composites.

Page 2 in the revised Supplementary information:

“Elemental analysis

Elemental analysis was performed using powder samples obtained by scraping the composite films from the substrate. The QD:OA ratio in the SEBS-g-MA (10 wt%) QD nanocomposite was determined from thermogravimetric analysis data (**Supplementary Fig. 38**). For the untreated composite, the composition was assumed to be QD 75.3 wt%, OA 14.7 wt%, and SEBS-g-MA 10.0 wt%.

For simplicity, given the low MA content in SEBS-g-MA ($\sim 1\text{ wt}\%$), ligand exchange was assumed to occur predominantly through the displacement of OA by MPA, while the

contribution from SEBS-g-MA was considered negligible in the calculation. The OA:MPA ratio was estimated from the measured C/S atomic ratio using the following equation:

$$\frac{C}{S} = \frac{18(\frac{m_{OA,0}}{282.47} - n_{MPA}) + 0.86\frac{m_{SEBS-g-MA}}{12.011} + 3n_{MPA}}{n_{S,ZnS} + n_{MPA}}$$

In this calculation, 282.47 is the molecular weight of OA, 12.011 is the atomic weight of carbon, and 0.86 is the carbon mass fraction of SEBS-g-MA. $m_{OA,0}$ is the initial mass of OA, $m_{SEBS-g-MA}$ is the mass of SEBS-g-MA, $n_{S,ZnS}$ is the number of moles of sulfur originating from the QDs, and n_{MPA} is the number of moles of MPA.

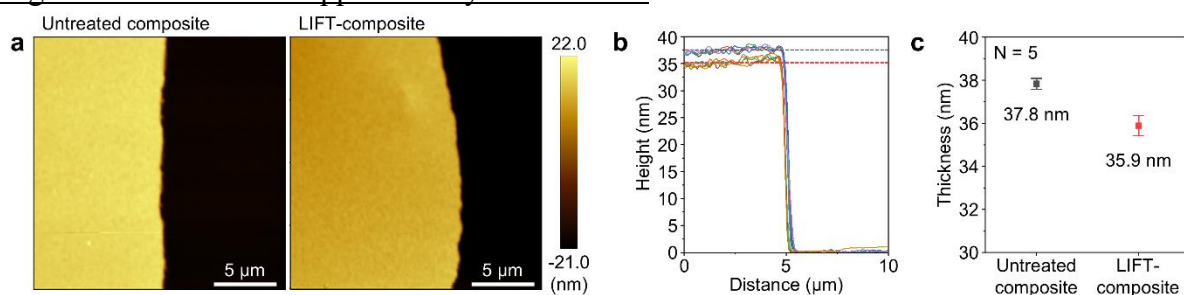
The value of $n_{S,ZnS}$ was first determined from the C/S ratio of the untreated composite. The amount of MPA in the LIFT-composite and the MPA composite was then calculated based on the elemental analysis results, and the OA:MPA ratio was subsequently determined from the calculated MPA content. The LIFT-composite shows a lower MPA ratio than the MPA composite, supporting the partial ligand displacement.”

Page 9 in the revised Supplementary information:

Supplementary Table 1 | Elemental analysis of untreated, LIFT-, and MPA QD nanocomposite films.

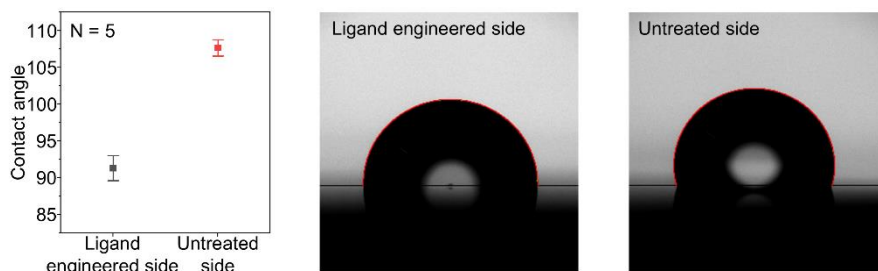
Sample	C wt%	S wt%	C/S ratio	OA:MPA
Untreated composite	16.81	12.83	3.49	1:0
LIFT-composite	16.80	14.89	3.01	1:0.32
MPA composite	14.25	18.66	2.03	1:3.55

Page 14 in the revised Supplementary information:



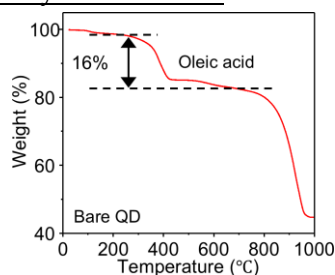
Supplementary Fig. 4 | a, AFM images of untreated (left) and LIFT- (right) composite films. b, Corresponding height profiles from (a). c, Comparison of the film thickness before and after ligand engineering.

Page 15 in the revised Supplementary information:



Supplementary Fig. 7 | Contact angles measured on the ligand engineered side and the opposite untreated side of the LIFT-composite.

Page 48 in the revised Supplementary information:



Supplementary Fig. 38 | Thermogravimetric analysis results of bare QD.

Comment #3: Furthermore, the preserved interparticle spacing observed via GISAXS before and after MPA treatment is used to argue that only surface QDs are modified. This interpretation is not valid. Given the minuscule size of MPA molecules compared to the QD nanoparticles and the polymer matrix, any ligand exchange would not be expected to cause a detectable shift in the average center-to-center distance between QDs measured by GISAXS. Consequently, this data cannot be used to support the claim that ligand displacement is confined exclusively to the interface and does not penetrate the bulk of the nanocomposite.

Our response to comment #3: We thank the reviewer for this insightful comment. We agree that preservation of the average interparticle spacing measured by GISAXS alone cannot conclusively demonstrate that ligand exchange is confined exclusively to the surface region. As also noted in our responses to **Comments #1 and #2**, we therefore rely on additional experimental observations and control experiments to support the interpretation of surface-localised ligand displacement.

In conventional film-state ligand exchange of QD solids with short-chain ligands such as MPA, replacement of long-chain native ligands typically results in a measurable reduction in interparticle spacing due to the shorter ligand length and denser packing of QDs (*Chem. Mater.* **27**, 474 (2015); *Nat. Nanotechnol.* **7**, 577 (2012)). If ligand exchange occurred extensively throughout the bulk of the nanocomposite, a similar contraction of the QD lattice would be expected. However, in our QD nanocomposite, the average interparticle spacing remains nearly unchanged after the standard LIFT treatment. While this observation alone does not prove

surface confinement, it is inconsistent with the interparticle contraction typically associated with extensive bulk ligand exchange in conventional QD films.

To further validate this point, we intentionally performed ligand exchange under significantly harsher conditions using a tenfold higher MPA concentration and a fivefold longer reaction time to promote ligand penetration into the bulk of the composite. Under these conditions, GISAXS reveals a clear reduction in interplanar spacing (**Supplementary Fig. 5**), consistent with extensive ligand exchange throughout the nanocomposite.

Furthermore, control experiments using physically mixed QD/SEBS composites (without chemical bonding between QDs and the polymer matrix) subjected to the same LIFT treatment exhibit a significant reduction in interparticle spacing (**Supplementary Fig. 6**). In contrast, the chemically bonded nanocomposite maintains its spacing under the standard treatment conditions. This comparative behaviour suggests that the covalently integrated elastomer matrix effectively restricts MPA penetration into the bulk of the nanocomposite, thereby suggesting that ligand displacement is predominantly confined to the near-surface region.

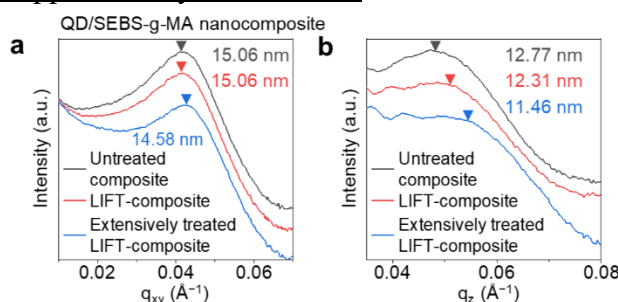
These additional control experiments and clarifications have been incorporated into **Supplementary Figs. 5 and 6** and described in the revised manuscript.

Our modification to the manuscript:

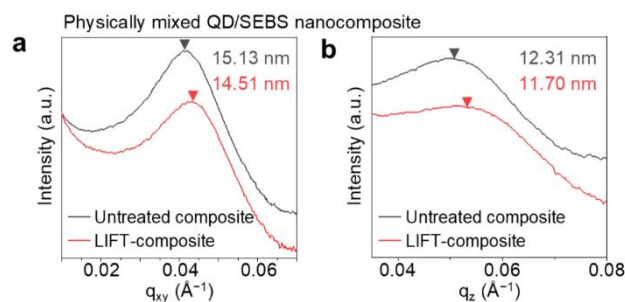
Page 9 in the revised manuscript:

“Consistently, GISAXS analysis reveals minimal change in interparticle spacing after LIFT treatment, whereas pronounced contraction occurs under harsher ligand-engineering conditions and in physically mixed QD/SEBS nanocomposites subjected to the same treatment (**Supplementary Fig. 5,6**).”

Page 16 in the revised Supplementary information:



Supplementary Fig. 5 | GISAXS line profiles along the in-plane (a) and out-of-plane (b) directions for untreated composite, LIFT-composite, and extensively treated LIFT composite under harsher ligand engineering conditions (10× MPA concentration and 5× reaction time). In conventional QD solids, film-state ligand exchange with short-chain ligands such as MPA typically induces contraction of the interparticle spacing through replacement of long-chain native ligands. By contrast, the QD/SEBS-g-MA nanocomposite shows largely unchanged interparticle spacing after the standard LIFT treatment, whereas the extensively treated LIFT-composite exhibits pronounced spacing contraction.



Supplementary Fig. 6 | GISAXS line profiles along in-plane (a) and out-of-plane (b) directions for untreated and LIFT-composite based on QD/SEBS nanocomposites. Unlike the chemically bonded QD/SEBS-g-MA nanocomposite, the physically mixed QD/SEBS nanocomposite exhibits a pronounced reduction in interparticle spacing after LIFT treatment.

Comment #4: *To convincingly establish the effect of MPA, critical control experiments are absent. The study lacks a control treatment using pure methanol to isolate the effects of the solvent itself on the nanocomposite's morphology and electronic properties. Additionally, to demonstrate the specificity of the thiol group interaction, a comparison with other short-chain molecules containing different terminal functional groups (e.g., amines, carboxylic acids without thiols) is necessary. Without these controls, the specific role of the thiol moiety in the purported interfacial engineering remains ambiguous and anecdotal.*

Our response to comment #4: We thank the reviewer for the valuable comment regarding the necessity of control experiments.

First, to isolate the effect of the solvent, we performed a control treatment using pure methanol. Because the QD nanocomposite film consists of QDs embedded within an elastomer matrix, no noticeable morphological change or increase in surface roughness is observed among untreated, methanol-treated, and LIFT-composites (**Supplementary Fig. 8**). Methanol treatment can partially remove loosely bound ligands from surface QDs; however, it also introduces defect sites, as evidenced by reduced exciton lifetimes in time-resolved photoluminescence (TRPL) measurements (**Supplementary Fig. 17**). As a result, QLEDs fabricated with methanol-treated composites show only marginal efficiency improvement and remain significantly inferior to devices fabricated using the MPA-based LIFT process (**Fig. R2**).

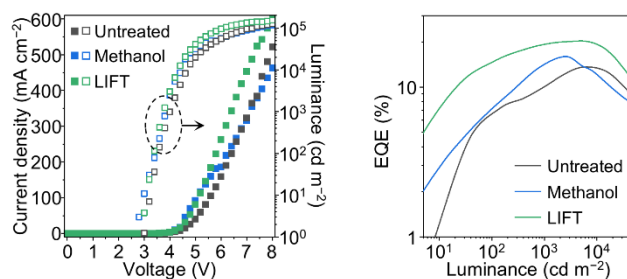


Fig. R2 | J - V - L characteristics (left) and EQE- L characteristics (right) of QLED fabricated with untreated, methanol-treated, and LIFT-composites.

Second, to examine the specificity of thiol-QD interactions, we compared the effects of short-chain ligands with different terminal functional groups (i.e., amine, carboxylic acid, and thiol). Ultraviolet photoelectron spectroscopy (UPS) measurements reveal that only thiol-containing ligands induce an upward shift of the energy levels, whereas amine and carboxylic acid treatments produce either a downshift or negligible change (**Supplementary Fig. 14**). These results indicate that the observed energy-level modulation is associated with thiol coordination to surface Zn atoms of the QDs rather than from nonspecific solvent or molecular effects.

These additional control experiments and discussions have been incorporated into **Supplementary Fig. 8, 14, 17, 18** as well as in the revised manuscript.

Our modification to the manuscript:

Page 9 in the revised manuscript:

“Atomic force microscopy (AFM) further shows that the LIFT-composite maintains a smooth surface morphology without aggregation (**Fig. 2f, Supplementary Fig. 8**), in contrast to the rough and aggregated morphology observed in bare QD films (**Supplementary Fig. 9**).”

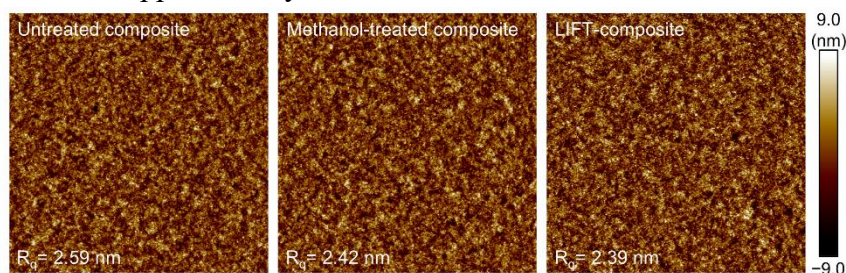
Page 10 in the revised manuscript:

“Notably, this modulation occurs exclusively with thiol-containing ligands, whereas amine ligands induce a slight downshift and carboxylic acid ligands produce negligible changes (**Supplementary Fig. 14**), highlighting the role of thiol coordination at the QD interface.”

Page 11 in the revised manuscript:

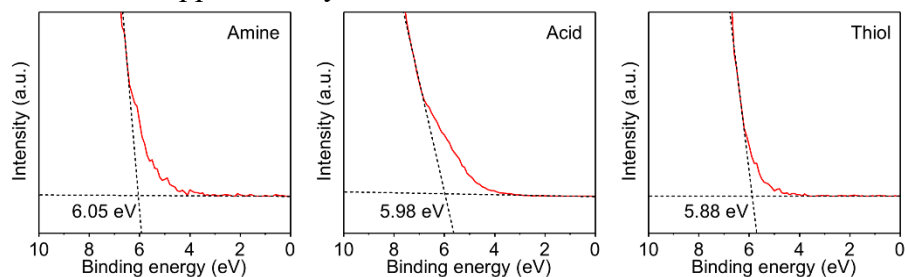
“In contrast, methanol treatment alone can partially remove loosely bound surface ligands with negligible energy level modulation, while simultaneously introducing defect sites, resulting in only marginal improvement in device performance (**Supplementary Figs. 17, 18**). This behaviour indicates that the performance enhancement is primarily associated with thiol-QD coordination rather than solvent-induced ligand removal.”

Page 18 in the revised Supplementary information:



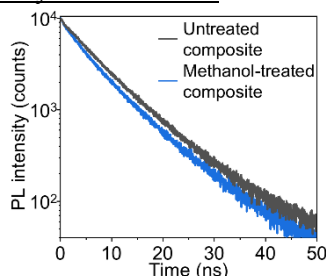
Supplementary Fig. 8 | AFM images of untreated (left), methanol-treated (middle), and LIFT-composites (right).

Page 24 in the revised Supplementary information:



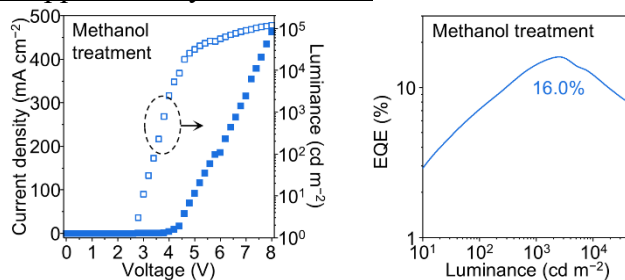
Supplementary Fig. 14 | UPS results of LIFT-composites treated with amine (propanamine), carboxylic acid (butyric acid), and thiol (propanethiol)-containing ligands.

Page 27 in the revised Supplementary information:



Supplementary Fig. 17 | TRPL decay profiles of untreated and methanol-treated composites.

Page 28 in the revised Supplementary information:



Supplementary Fig. 18 | J - V - L (left) and EQE- L (right) characteristics of QLED fabricated with methanol-treated composite.

Comment #5: *On the point of the proposed mechanism for enhanced hole injection, the authors' explanation requires further scrutiny and stronger evidence. The argument rests on two pillars: the reduced physical barrier due to the short-chain MPA ligand and the elevated HOMO level of the quantum dots (QDs) facilitated by the electron-donating thiol group. However, several critical issues undermine this proposed mechanism. Firstly, the use of Cyclic Voltammetry (CV) to determine the HOMO energy levels of solid-state nanocomposite films is questionable. For such materials, Ultraviolet Photoelectron Spectroscopy (UPS) is the established and more*

direct method for measuring work functions and ionization energies (HOMO levels). CV measurements in solution may not accurately reflect the electronic structure at the critical solid-solid interface within the device. Secondly, according to the energy level diagram presented in Figure 2h, the hole injection barrier from the HTL (TFB, HOMO ~ -5.3 eV) to the QDs remains substantial. Even after MPA treatment, the HOMO level of the QD composite is reported at -5.89 eV (LIFT) or -5.80 eV (full MPA), implying a persistent injection barrier of approximately 0.5 eV to 0.6 eV. A shift of merely 0.07 eV to 0.16 eV seems inconsequential for effectively overcoming such a high barrier. The authors do not provide a convincing physical model for how such a minor shift could lead to the observed dramatic improvements in device performance. Thirdly, the data from hole-only (HoDs) and electron-only devices (EoDs) further complicate the narrative. The manuscript states that the enhanced hole current in the LIFT-processed HoDs remains “markedly” lower than the electron current in the EoDs, which is largely unchanged. If the hole injection is still the severely limited factor, it is difficult to reconcile this with the achievement of a high EQE of 20.1% in the balanced full QLED structure. Such high efficiency typically requires nearly balanced charge injection, which the presented HoD/EoD data do not convincingly demonstrate. In light of these points, the claimed mechanism of improved charge balance via enhanced hole injection appears insufficiently supported. Therefore, the authors must provide further and more rigorous experimental evidence to clarify the origin of the efficiency enhancement.

Our response to comment #5: We thank the reviewer for the thoughtful comments regarding the proposed mechanism of enhanced hole injection and charge balance. In response, we have performed additional measurements and clarified the physical interpretation of the interfacial ligand engineering mechanism.

First, we clarify that the cyclic voltammetry (CV) measurements were conducted on solid-state nanocomposite films rather than in solution. The highest occupied molecular orbital (HOMO) levels were derived by correlating the film-state CV results with the corresponding photoluminescence (PL) emission peaks. Nevertheless, following the reviewer’s suggestion, we performed UPS measurements to directly determine the HOMO levels in the solid state (**Supplementary Fig. 12**). The UPS results reveal HOMO levels of -5.97 eV, -5.87 eV, and -5.50 eV for the untreated composite, LIFT-composite, and QD nanocomposite containing a high fraction of MPA-engineered QDs (MPA composite), respectively. These results confirm the same energetic trend obtained from CV and have now been incorporated into **Fig. 2h** and **Supplementary Fig. 12**.

Second, we agree that the HOMO shift observed in the LIFT-composite (~0.47 eV relative to the untreated film) alone cannot fully account for the large improvement in device performance. Similar energy-level modulation has been reported in previous studies employing thiol ligands in QLEDs (*ACS Photon.* **12**, 6578 (2025); *Nano Lett.* **24**, 15781 (2024); *Small* **20**, 2307298 (2024); *Adv. Optic. Mater.* **8**, 1901314 (2020)). Importantly, the LIFT process simultaneously reduces the effective tunnelling barrier at the HTL/EML interface by displacing insulating elastomeric components surrounding the surface QDs (**Supplementary Fig. 4**).

Consequently, hole injection is improved through a combination of moderate energy-level upshifting and reduced interfacial barrier thickness. This interpretation is further supported by the lower device impedance observed in the Nyquist plots of LIFT-fabricated QLEDs (**Fig. R3**). Collectively, these results support that the device improvement arises from the synergistic effects of interfacial barrier thinning and moderate HOMO upshifting rather than from energy-level shifting alone.

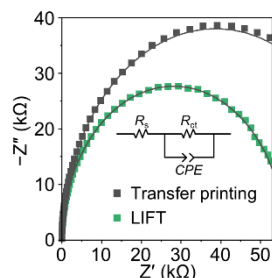


Fig. R3 | Nyquist plots of QLEDs fabricated via conventional transfer printing and LIFT.

Third, we appreciate the reviewer’s important point regarding the lower hole current observed in LIFT-fabricated hole-only devices (HoDs) compared with the electron current in electron-only devices (EoDs). However, the absolute current levels measured in single-carrier devices do not directly represent charge balance in operating QLEDs, but rather reflect relative carrier injection trends under simplified conditions. In practical QLED operation, charge balance is governed by additional processes including QD charging, field-dependent carrier accumulation, recombination dynamics, and leakage pathways that are absent in HoD and EoD configurations (*Nat. Photon.* **16**, 505 (2022); *Nat. Commun.* **11**, 2309 (2020)). Consequently, direct comparison of HoD and EoD current densities cannot quantitatively represent charge balance in full devices. Indeed, highly efficient QLEDs have frequently been reported to exhibit asymmetric HoD/EoD currents depending on device architecture and charge-blocking layers (*Adv. Funct. Mater.* **35**, 2418868 (2025); *Nat. Commun.* **16**, 10143 (2025); *Nature* **586**, 358 (2020); *Nat. Photon.* **13**, 192 (2019)). In this context, the single-carrier devices primarily demonstrate the relative enhancement of hole injection induced by the LIFT process, while the high EQE observed in the full QLED structure reflects the resulting improvement in charge recombination efficiency and effective charge balance.

Taken together, the additional UPS measurements, impedance analysis, and device studies collectively support that MPA-based interfacial engineering simultaneously induces moderate HOMO upshifting and reduces the effective tunnelling barrier at the HTL/QD interface. Consequently, the improved device performance arises from the combined effects of interfacial barrier thinning and energy-level modulation rather than from HOMO shifting alone.

Our modification to the manuscript:

Page 10 in the revised manuscript:

“The MPA ligands, containing electron-donating thiol groups, modulate the energy levels of QDs through selective interfacial coordination. **Ultraviolet photoelectron spectroscopy (UPS)**

reveals a progressive upshift of the HOMO level from -5.97 eV in untreated composites to -5.89 eV in LIFT-composites and -5.50 eV in QD nanocomposite containing a high fraction of MPA-engineered QDs (MPA composites), indicating localised surface potential modulation (Fig. 2h, Supplementary Fig. 12). Cyclic voltammetry (CV) measurements performed on solid-state films show consistent trends, further confirming these energy-level shifts (Supplementary Fig. 13).”

Page 10 in the revised manuscript:

“While electron-only devices (EoDs; indium tin oxide (ITO)/ZnO/QD nanocomposite/ZnO/Al) exhibit comparable current density regardless of the processing method (Supplementary Fig. 15), hole-only devices (HoDs; ITO/poly(3,4-ethylenedioxythiophene):polystyrene sulfonate (PEDOT:PSS)/poly(9,9-dioctylfluorene-alt-N-(4-sec-butylphenyl)-diphenylamine) (TFB)/QD nanocomposite/MoO₃/Au) fabricated via LIFT show markedly higher current densities (Fig. 2k). This behaviour is consistent with reduced interfacial tunnelling barriers at the HTL/EML interface together with moderate HOMO upshifting, which collectively promote hole injection.”

Page 27 in the revised manuscript:

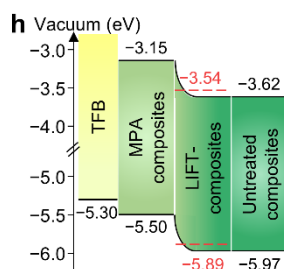
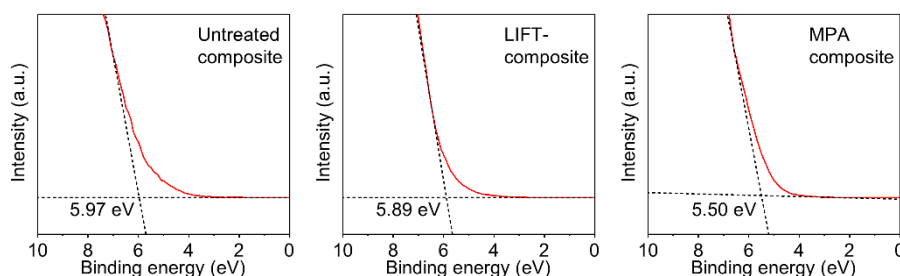


Fig. 2 | h, Energy level alignment of QD nanocomposite films (i.e., untreated, LIFT-, and MPA composites).

Page 22 in the revised Supplementary information:



Supplementary Fig. 12 | UPS results of untreated (left), LIFT- (middle), and MPA (right) composites.

Comment #6: *The authors propose a key mechanism for the LIFT process wherein mechanical strain is locally concentrated at the trench edges via thermal expansion of the PDMS stamp. They support this claim with finite element simulations, suggesting that this localized strain*

facilitates clean cleavage at the trench boundaries during patterning. This is presented as acting synergistically with MPA-induced interfacial adhesion enhancement to achieve high-definition patterning. However, I have significant concerns regarding the substantiation of this claim and its practical relevance to the final stretchable displays:

Firstly, While Figure 3e convincingly demonstrates the superior patterning yield of the LIFT technique on a rigid substrate, this advantage appears severely compromised in the actual stretchable displays shown in Figure 4e-4l. The operational devices exhibit noticeable edge blurring, non-uniform emission, and apparent pixel defects under strain. This observable degradation critically weakens the authors' central claim that the LIFT process enables high-resolution patterning for stretchable displays. The data suggest that the process, while effective for creating patterns on a static, rigid surface, may not fully maintain this precision when integrated into a multi-layer, mechanically dynamic device architecture. To objectively validate their comparison and exclude potential selection bias in the presented images, the authors should provide video evidence demonstrating the patterning process and the resulting pixel integrity across a large array, both before and after deformation.

Secondly, the authors rely solely on a finite element analysis of strain distribution to support the core premise that thermal expansion induces localized cleavage. This analysis lacks critical details and experimental validation. The model is incomplete without a corresponding simulation of the temperature field distribution. The thermal expansion is driven by heat applied via a hot plate; therefore, the thermal conduction through the substrate, the QD composite layer, and the PDMS stamp is fundamental to the proposed mechanism. The absence of this thermal data leaves the mechanical boundary conditions of the strain simulation unjustified. The claim that thermal expansion is the primary driver requires direct experimental evidence. All materials involved (substrate, QD composite, PDMS) will expand upon heating. The critical question is the differential thermal expansion and the resulting stress fields that lead to fracture. The argument is not fully convincing without quantitative, experimental measurements of the stress or strain evolution during the heating process, for instance using techniques like in-situ stress monitoring or digital image correlation. Furthermore, the influence of temperature should be systematically analyzed. A comprehensive analysis of how different temperatures affect the thermal expansion, resultant stress fields, and ultimately the patterning fidelity is absent. Such a study is crucial to establish a reliable process window and to fully validate the proposed thermally assisted cleavage mechanism.

Our response to comment #6: We thank the reviewer for the constructive comments regarding the patterning mechanism and its relevance to stretchable displays.

First, regarding the concern that the patterning fidelity observed on rigid substrates may not be maintained in the final stretchable devices, we would like to clarify that the high-definition pixel geometry established by the LIFT process is preserved throughout the device integration process and under mechanical deformation. Optical microscopy images acquired during the fabrication of intrinsically stretchable QLEDs show that the patterns initially formed on the PDMS stamp are successfully transferred onto the PUA/AgNWs/conductive

PEDOT:PSS/PEDOT:PSS/TFB stack without noticeable distortion or failure (**Supplementary Fig. 34**). After device fabrication, the electroluminescence (EL) emission spatially corresponds well to the transferred patterns (**Fig. 4g**). This correspondence is consistently maintained in large-array configurations with different pixel sizes and spacings (**Extended Data Fig. 10a–h**), and the pixel integrity is preserved under mechanical deformation (**Fig. R4**). To directly address the reviewer’s concern regarding possible selection bias and to provide objective evidence over large arrays, we have additionally included video evidence demonstrating both the pixelated pattern formation process and the preservation of pixel geometry during repeated stretching and relaxation, with sustained spatially resolved emission across the entire array and without noticeable pixel degradation (**Supplementary Video 1**).

These results confirm that the pattern fidelity established by the LIFT process is maintained not only on static substrates but also in the fully integrated stretchable device architecture.

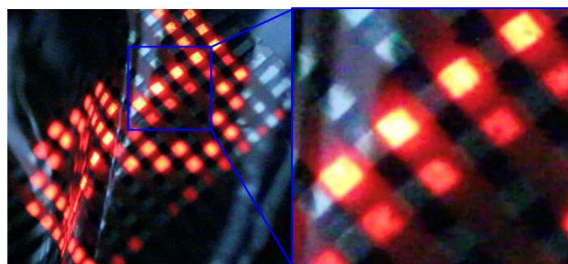


Fig. R4 | Images of stretchable QLED and corresponding magnified view.

Second, to further substantiate the thermally assisted cleavage mechanism, we performed additional temperature-dependent finite element analysis (FEA) together with systematic patterning experiments at different processing temperatures. The FEA was designed to capture strain localisation under steady-state heating on a hot plate. A steady-state heat transfer simulation was first performed to obtain the temperature field distribution during hot-plate heating (**Fig. R5**). Owing to the substantially higher thermal conductivity of the SiO₂ trench ($2.51 \text{ W m}^{-1} \cdot \text{K}^{-1}$) compared to the PDMS stamp ($0.15 \text{ W m}^{-1} \cdot \text{K}^{-1}$) and the QD nanocomposite layer ($0.1\text{--}0.3 \text{ W m}^{-1} \cdot \text{K}^{-1}$), heat is efficiently conducted through the trench, producing localised temperature field near the trench under steady-state conditions. This localised heating leads to spatially non-uniform thermal expansion, which contributes to strain concentration at the trench edges.

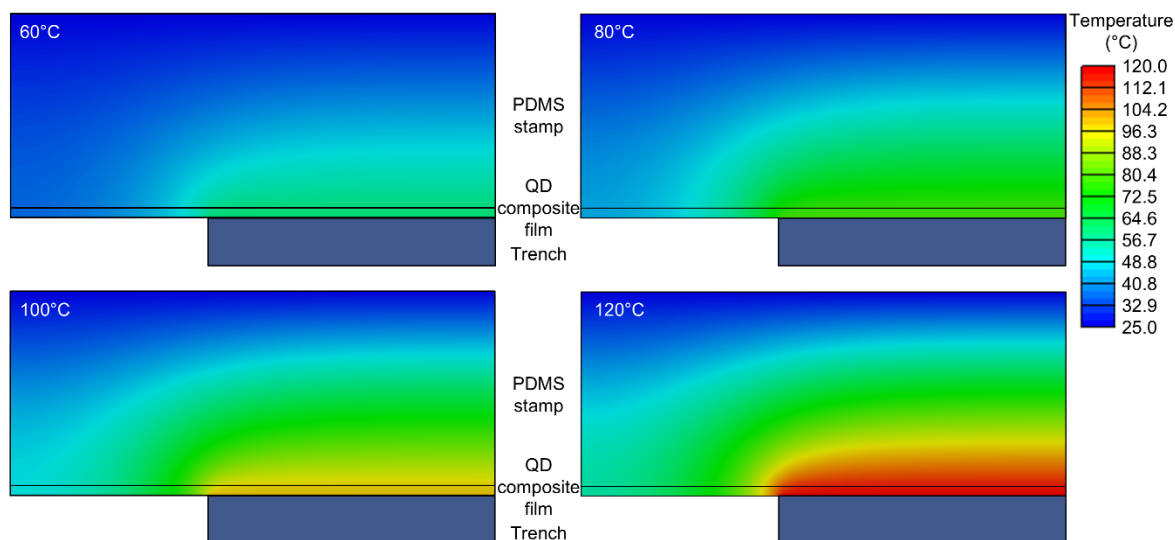


Fig. R5 | Simulated steady-state temperature distribution during hot-plate heating.

Importantly, because the PDMS stamp possesses a significantly higher thermal expansion coefficient ($3.1 \times 10^{-4} \text{ K}^{-1}$) than SiO_2 ($0.5 \times 10^{-6} \text{ K}^{-1}$) and is several orders of magnitude thicker (millimeter-scale) than the nanometer-scale QD nanocomposite film, the dominant thermomechanical response is governed by PDMS expansion, while the contribution from thermal expansion of the ultrathin QD layer is expected to be minor. Under these conditions, temperature-dependent FEAs show that increasing temperature amplifies differential thermal expansion and enhances strain localisation at the trench edges, which facilitates clean cleavage during patterning (**Supplementary Fig. 20**).

To experimentally validate this mechanism, we additionally performed systematic temperature-dependent patterning experiments, which show a clear improvement in pattern fidelity with increasing processing temperature (**Supplementary Fig. 21**). These experimental observations are in excellent agreement with the simulation results. Furthermore, we found that the LIFT-induced enhancement of interfacial adhesion between the QD nanocomposite and the trench broadens the process window, enabling stable and high-fidelity patterning over a wider temperature range without changing the underlying strain-localisation mechanism (**Supplementary Figs. 23–26**).

These results provide both experimental and simulation evidence supporting the proposed thermally assisted cleavage mechanism.

These clarifications and additional experimental results have been incorporated into **Fig. 4g**, **Extended Data Fig. 10a–h**, and **Supplementary Figs. 20, 21, 23–26, and 34** in the revised manuscript.

Our modification to the manuscript:

Page 12 in the revised manuscript:

“Finite element simulations reveal that this differential thermal expansion generates pronounced strain concentration within the QD nanocomposite film near the trench boundaries,

thereby facilitating clean cleavage during pattern formation (**Fig. 3b**, **Supplementary Fig. 20**). Consistent with this mechanism, temperature-dependent experiments show that pattern fidelity progressively improves with increasing processing temperature, in agreement with enhanced strain localisation predicted by the simulations (**Supplementary Fig. 21**). Concurrently, interfacial adhesion is reinforced via ligand engineering with MPA, whose polar carboxyl end groups form stronger interactions with the SiO₂ trench surface than untreated composites or composites treated with short-chain ligands lacking terminal functional groups (**Fig. 3c**, **Supplementary Fig. 22**, **Supplementary Method**). This enhanced interfacial adhesion broadens the processing window and enables robust pattern formation over a wider temperature range (**Supplementary Fig. 23–26**).”

Page 15 in the revised manuscript:

“Beyond the device-level performance enhancement, our strategy also enables high-definition stretchable QLEDs with well-defined pixelated emission while preserving the integrity of the patterned QD arrays (**Fig. 4g**, **Supplementary Fig. 34**).”

Page 16 in the revised manuscript:

“The well-defined pixel architecture achieved at the single-device level is preserved in the array configuration across various pixel sizes and interpixel spacings (**Extended Data Fig. 10a–h**).”

Page 28 in the revised manuscript:

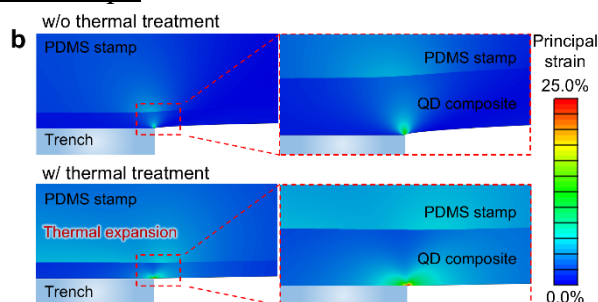


Fig. 3 | b, Finite element analysis of localised strain distribution at trench edges during pattern formation without and with thermal treatment.

Page 29 in the revised manuscript:

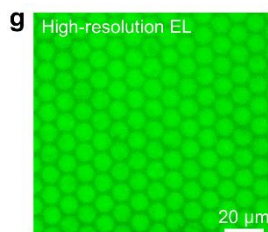
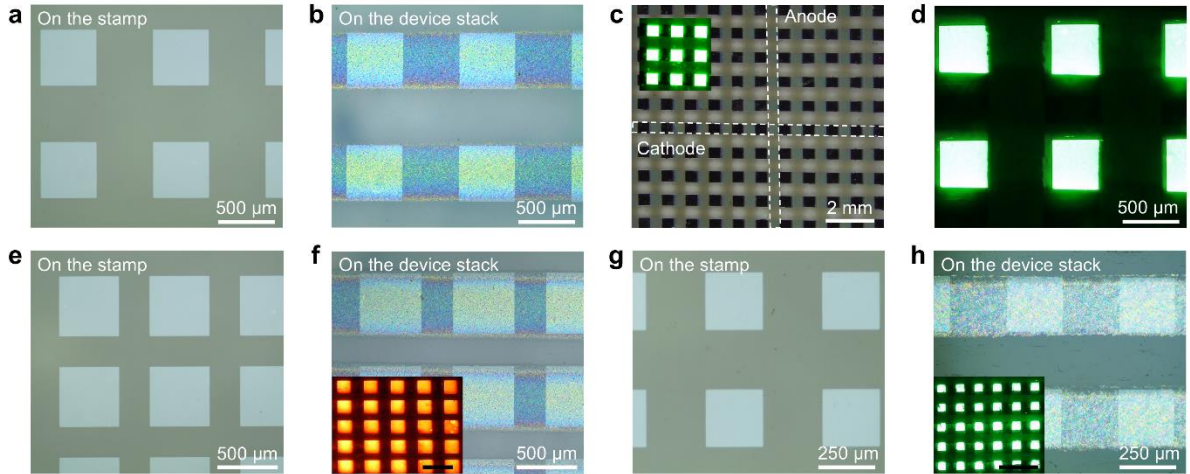


Fig. 4 | g, Optical microscope image of high-definition EL emission.



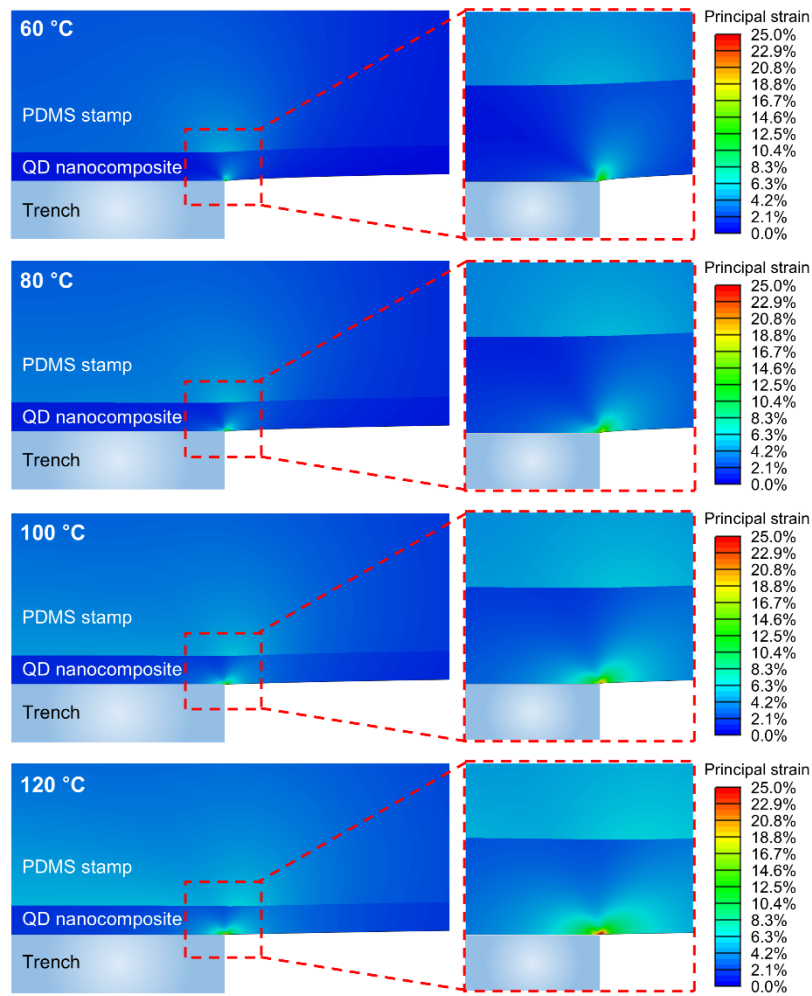
Extended Data Fig. 10 | **a–d**, Optical microscope images of pixel arrays on the stamp (a), transferred onto the device stack (b), and after full device fabrication (c), and the corresponding EL image during operation (d), for pixels with a size of 500 μm and an interpixel spacing of 500 μm. **e–h**, Optical microscope images of pixel arrays on the stamp (e,g) and after transfer onto the device stack (f,h), for arrays with 500 μm pixels and 250 μm spacing (e,f), and 250 μm pixels and 250 μm spacing (g,h). Insets show the corresponding EL images of the arrays during device operation.

“Finite element analysis

Thermo-mechanical strain and stress distributions during trench-assisted transfer of QD/elastomer composite films were simulated using commercial finite element software (Abaqus 2022, Dassault Systèmes, France) through a sequentially coupled thermal–mechanical approach. A two-dimensional plane strain model was constructed to represent the cross-sectional geometry of the PDMS stamp, QD composite layer, and underlying trench. The geometry was discretized using 4-node bilinear plane strain quadrilateral elements with reduced integration (CPE4R) for the mechanical analysis and corresponding 4-node heat transfer elements (DC2D4) for the thermal analysis. The final mesh consisted of 4,446 elements and 4,566 nodes, with locally refined meshing near the trench corner to accurately capture steep strain gradients. To realistically account for localised annealing, a steady-state heat transfer analysis was first performed, where a constant temperature corresponding to the annealing condition (60–120 °C) was prescribed at the trench region. Because the actual PDMS stamp thickness (~1 cm) is substantially larger than the modeled thickness (1 μm) used for computational efficiency, the truncated top surface of the PDMS domain was treated using an equivalent surface film boundary condition defined as $h_{eq} = k_{PDMS}/L_{omitted}$, where k_{PDMS} is the thermal conductivity of PDMS and $L_{omitted}$ represents the unmodeled bulk thickness. This approach approximates conductive heat dissipation into the bulk PDMS and enables formation of a physically realistic temperature gradient. The resulting nodal temperature field was

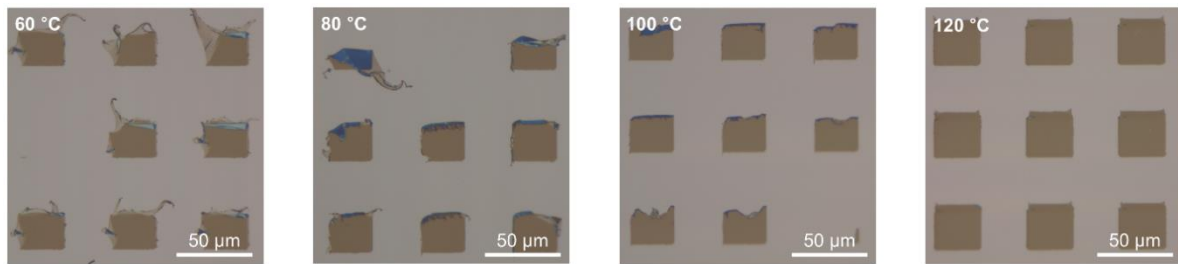
extracted from the thermal analysis and imported into a subsequent static mechanical step as a predefined temperature field. An initial uniform temperature of 25°C was assigned as the stress-free reference state, and thermal strain was incorporated through isotropic thermal expansion, $\epsilon_{th} = \alpha(T - 25^\circ\text{C})$. Mechanical boundary conditions were applied to prevent rigid body motion while allowing thermally induced deformation. Strain localisation near the trench edge was evaluated by extracting the maximum in-plane principal strain and comparing it with the far-field strain within the PDMS bulk to assess the influence of annealing temperature on deformation contrast governing pattern transfer.”

Page 30 in the revised Supplementary information:



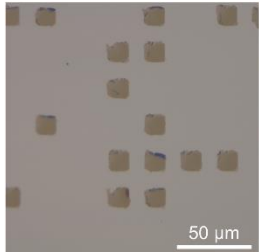
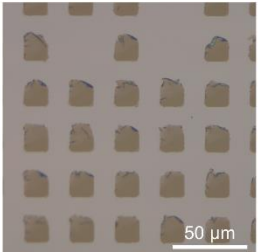
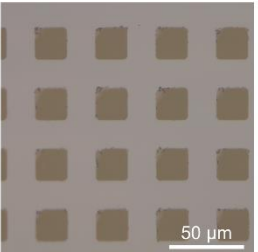
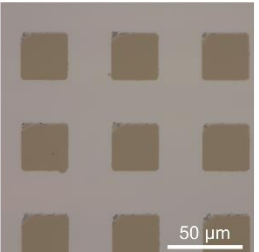
Supplementary Fig. 20 | Finite element analysis of localised strain distribution at trench edges during pattern formation under different processing temperatures.

Page 31 in the revised Supplementary information:



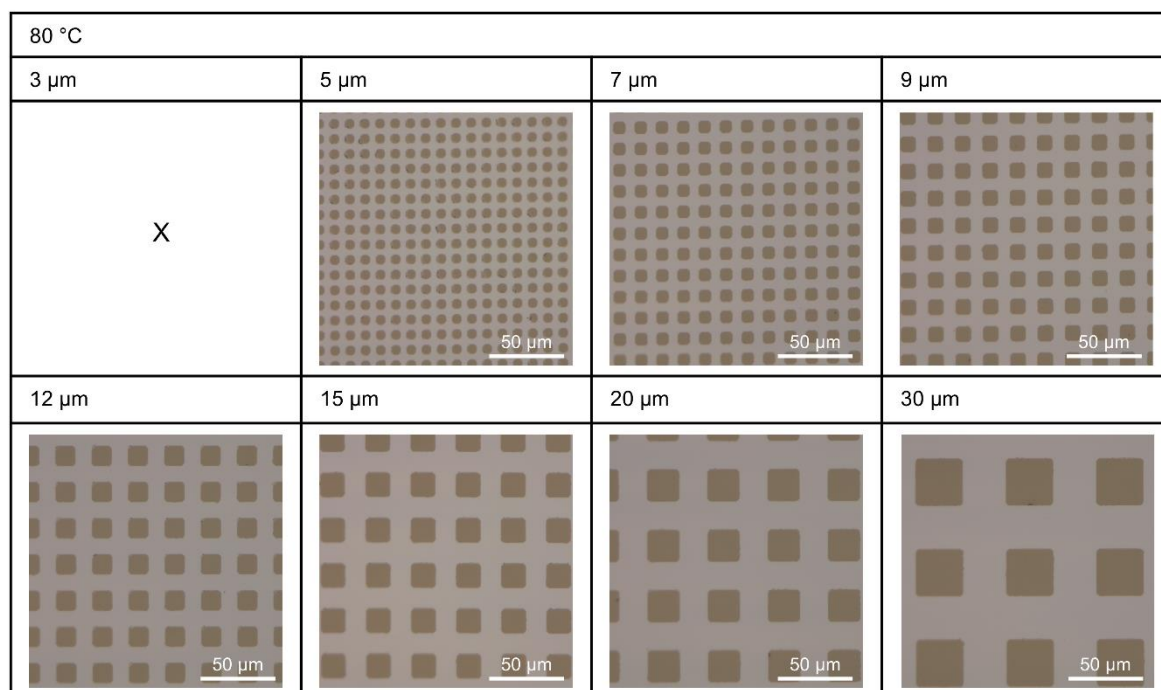
Supplementary Fig. 21 | Optical microscope images showing the temperature-dependent evolution of pattern fidelity in QD nanocomposite films processed at 60, 80, 100, and 120 °C.

Page 33 in the revised Supplementary information:

60 °C			
3 μm	5 μm	7 μm	9 μm
X	X	X	X
12 μm	15 μm	20 μm	30 μm
			

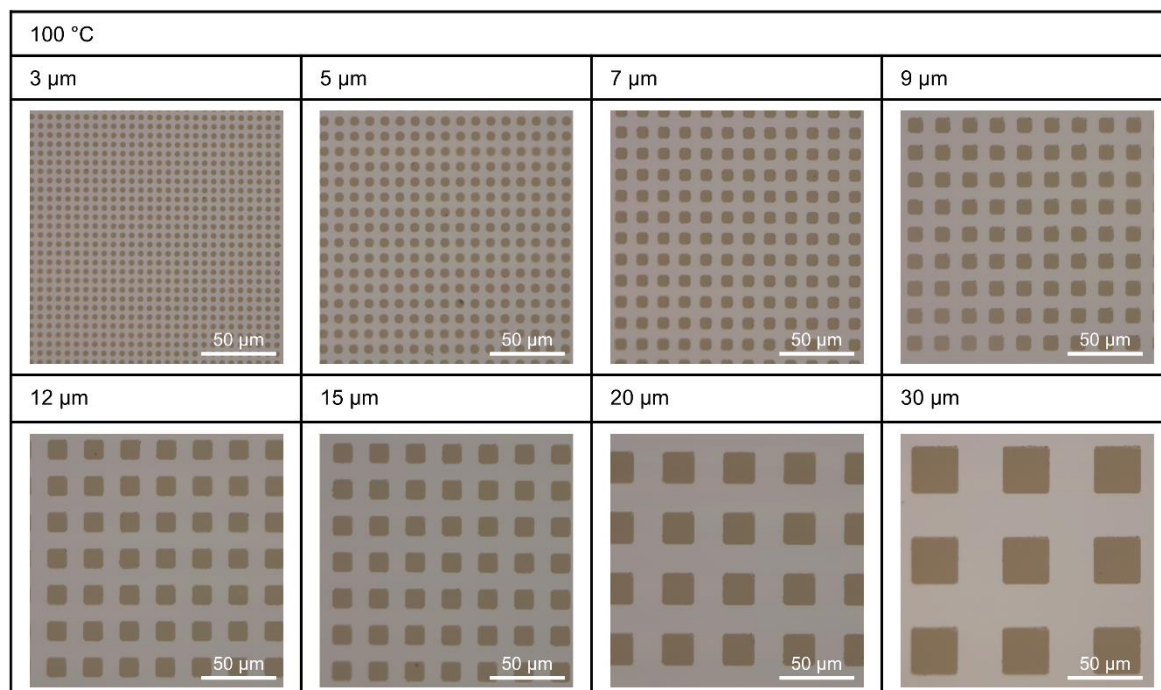
Supplementary Fig. 23 | Optical microscope images showing the effect of feature size on pattern fidelity in QD nanocomposite films processed at 60 °C.

Page 34 in the revised Supplementary information:

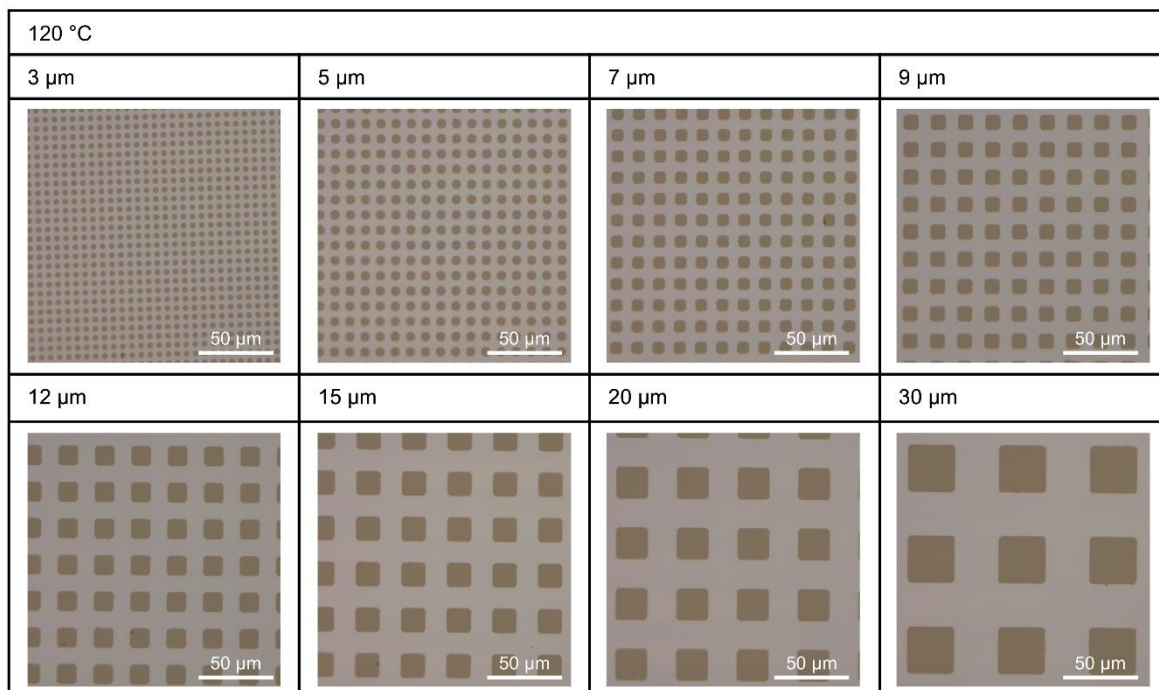


Supplementary Fig. 24 | Optical microscope images showing the effect of feature size on pattern fidelity in QD nanocomposite films processed at 80 °C.

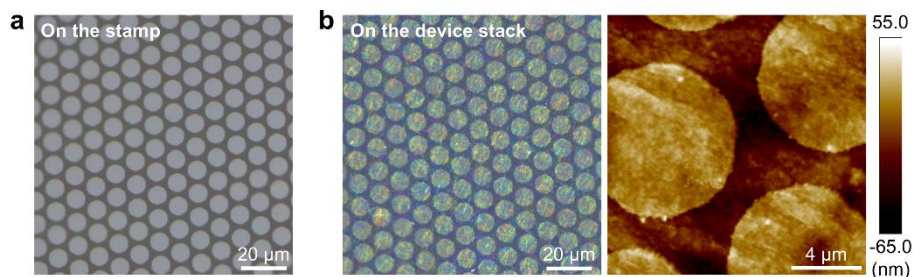
Page 35 in the revised Supplementary information:



Supplementary Fig. 25 | Optical microscope images showing the effect of feature size on pattern fidelity in QD nanocomposite films processed at 100 °C.



Supplementary Fig. 26 | Optical microscope images showing the effect of feature size on pattern fidelity in QD nanocomposite films processed at 120 °C.



Supplementary Fig. 34 | **a,b**, Optical microscope images of high-resolution pattern arrays on the stamp (a) and after transfer onto the device stack (b), demonstrating preservation of pattern fidelity during device integration. The corresponding AFM image of the transferred pattern is shown to the right of panel (b).

Minor comment #1: In Figure 1a, a schematic illustration outlines the LIFT strategy, detailing its four key fabrication steps. To allow readers to fully comprehend the process, this schematic must be substantiated with corresponding photographic or microscopic evidence. The following visual data would be essential for clarity: Optical or PL images, alongside SEM micrographs, of the PDMS stamp and the ODTS-treated donor substrate during and after the QD composite pickup step; PL images of the PDMS/QD composite sample following the

interfacial ligand engineering with MPA, demonstrating any changes in surface properties; PL and SEM images of the residual QD composite remaining on the PDMS stamp after the thermally assisted pattern formation on the intaglio trench; PL and SEM images of the PDMS stamp after the final pattern release, confirming the successful and clean transfer of the QD patterns.

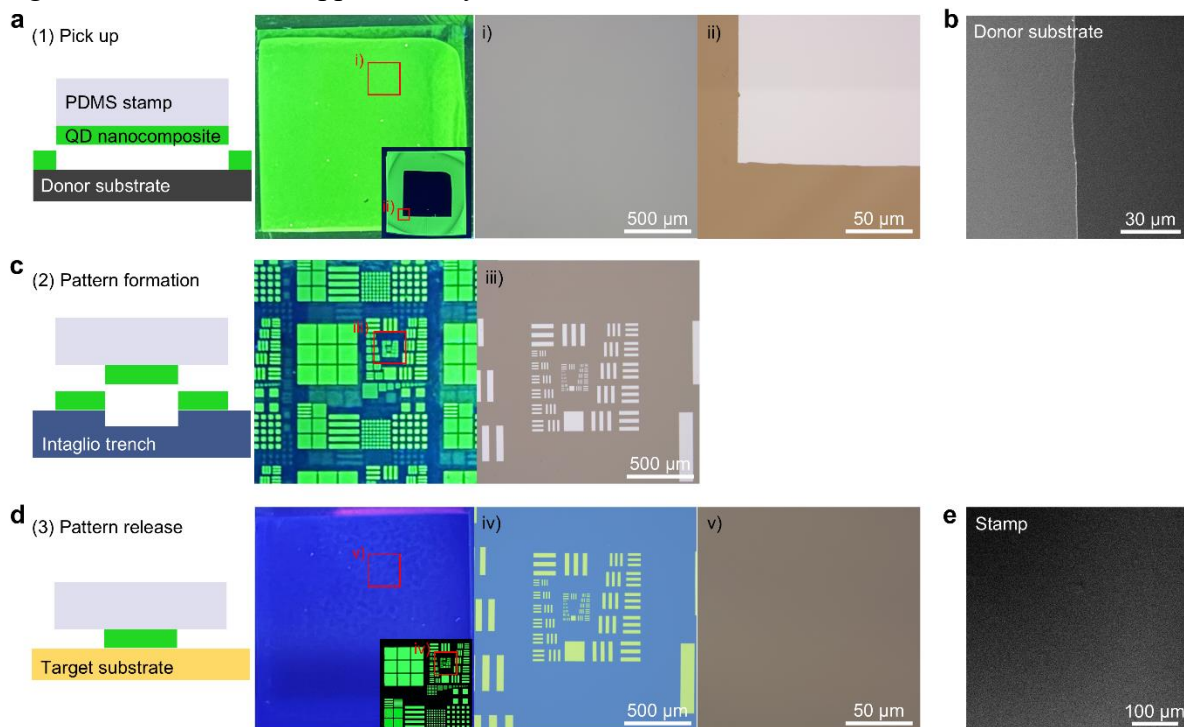
Our response to minor comment #1: We thank the reviewer for the helpful comment. To provide direct visual evidence supporting the schematic illustration in Fig. 1a and to facilitate a clearer understanding of each fabrication step, we have added comprehensive PL, OM, and scanning electron microscopy (SEM) images corresponding to the key stages of the LIFT process (**Supplementary Fig. 1**).

Specifically, after the pickup step, PL, OM, and SEM images confirm the clean transfer of the QD nanocomposite from the ODTS-treated donor substrate to the PDMS stamp, with no detectable residual material on the donor substrate (**Supplementary Fig. 1a,b**). Following the interfacial ligand engineering and thermally assisted pattern formation steps, PL and OM images of the PDMS stamp show well-defined patterns across multiple feature sizes (**Supplementary Fig. 1c**). After the final pattern release step, PL, OM, and SEM images of both the PDMS stamp and the receiver substrate confirm successful transfer of the patterned QD nanocomposite with no noticeable residual material remaining on the stamp (**Supplementary Fig. 1d,e**). Regarding the ligand engineering step, additional spectroscopic evidence has been provided in our responses to **Comments #1 and #2**.

As suggested by the reviewer, we have included comprehensive visual validation for each key step of the LIFT process in **Supplementary Fig. 1**.

Our modification to the manuscript:

Page 11 in the revised Supplementary information:



Supplementary Fig. 1 | **a**, Schematic illustration of the QD composite pick-up process (left), PL images of the PDMS stamp (middle) and the donor substrate (middle, inset) after pick up, and the corresponding optical microscope images (i, ii). **b**, SEM image of the donor substrate after the pick-up step, confirming the absence of residual material. **c**, Schematic illustration of the pattern formation process (left), PL image of the PDMS stamp after pattern formation (middle), and the corresponding optical microscope images (iii). **d**, Schematic illustration of the pattern release process (left), PL images of the PDMS stamp (middle) and the receiver substrate (middle, inset) after release, together with the corresponding optical microscope images (iv, v). **e**, SEM image of the PDMS stamp after pattern release, confirming clean transfer without residual QD nanocomposite.

Minor comment #2: Figure 1i shows PL images of high-resolution full-colour QD pixel arrays under planar and convex deformations. The image under convex deformation reveals localized pixel failure, notably in the vicinity of the scale bar, which compromises the authors claim about the LIFT strategy.

Minor comment #3: The description of the samples used for the full-colour QD pixel arrays presented in Figure 1j is critically insufficient for a proper evaluation of the results. Specifically, the manuscript fails to provide essential fabrication and structural details, such as: the precise thickness of the red, green, and blue QD nanocomposite films that constitute the subpixels; the nature and material properties of the substrate onto which these patterned arrays are transferred and imaged.

Our response to minor comment #2 and #3: We thank the reviewer for helpful comments.

To clarify the mechanical robustness of the patterned full-colour QD arrays under deformation, we have added additional PL images acquired under convex deformation from different viewing angles (**Supplementary Fig. 2a**). These images confirm that the overall pixel geometry remains well-defined and spatially preserved under deformation. Importantly, the apparent pixel irregularities observed near the scale bar in **Fig. 1j** originate from viewing-angle-dependent emission contrast and optical reflection effects during imaging, rather than from actual structural failure of the patterned QD arrays.

We agree that additional structural and fabrication details are important for proper evaluation of the full-colour QD pixel arrays. The intrinsically stretchable full-colour patterns were fabricated by vertically stacking red, green, and blue dot arrays with different feature sizes to achieve the desired color balance and image contrast (**Supplementary Fig. 2b**). The thickness of each QD nanocomposite film was individually optimized to account for differences in photoluminescence quantum yield (PLQY) and UV colour-conversion efficiency among the red, green, and blue emitters.

Following the reviewer's suggestion, the additional structural and fabrication details of the full-colour QD pixel arrays have been included in **Supplementary Fig. 2** and the revised manuscript.

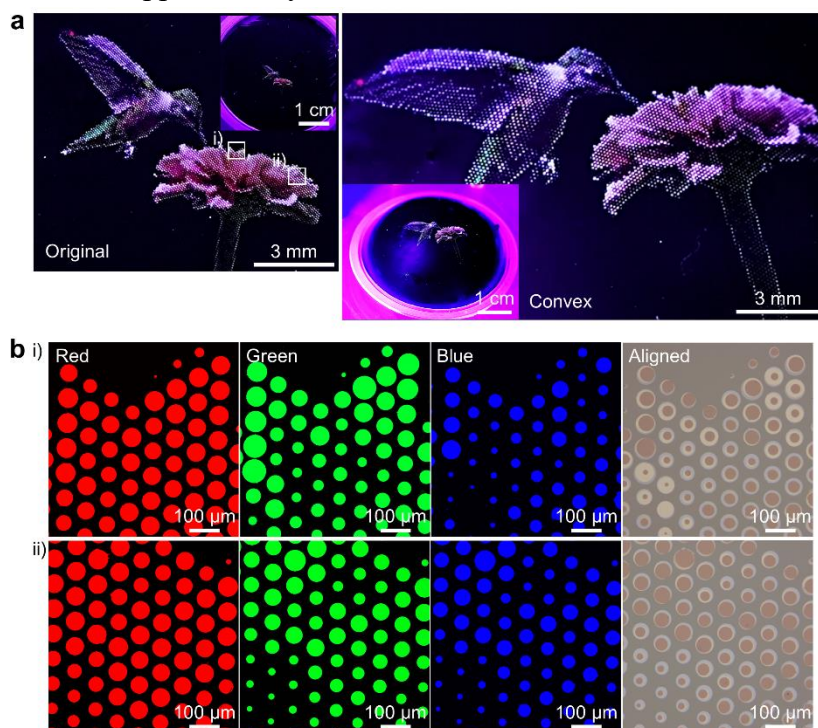
Our modification to the manuscript:

Page 26 in the revised manuscript:

“j, PL images of high-resolution full-colour QD pixel arrays (318 PPI) under planar and convex deformations.”

Page 13 in the revised manuscript:

“Beyond lateral patterning, the LIFT process also enables vertical stacking of QD nanocomposite films, allowing multilayered RGB architectures for full-colour emission (**Fig. 1j, Supplementary Fig. 2**).”



Supplementary Fig. 2 | a, PL images of high-definition full-colour QD pixel arrays under planar (left) and convex (right) deformation. A stretchable SEBS/carbon black composite is used as the substrate. **b**, Magnified fluorescence images of the RGB subpixels from the boxed region in (a) and the corresponding optical microscope image of the vertically stacked RGB subpixels (right).

Minor comment #4: Based on the data presented in Figures 2a-2c, Extended Data Figure 3, and Extended Data Figure 4, the authors claim that the chemical bonding between SEBS-g-MA and the QDs enhances mechanical stretchability. While this conclusion may be valid, the current evidence raises significant concerns regarding its reliability. In Extended Data Figure 3b, the FT-IR spectra comparison is made only between pristine QDs and the QD/SEBS-g-MA composite. The critical control—a spectrum of the QD/SEBS composite—is absent. This omission makes it impossible to definitively attribute the observed spectral changes to the specific chemical bonding between QDs and SEBS-g-MA, as opposed to general physical interactions present in both composites. The AFM images in Extended Data Figure 4b and 4c reveal that the QD/SEBS and QD/SEBS-g-MA composites exhibit fundamentally different surface morphologies even at 0% strain. This suggests a significant disparity in the initial dispersion and distribution of QDs within the two polymer matrices. Consequently, the superior mechanical performance of the QD/SEBS-g-MA composite cannot be ascribed solely to chemical bonding; it may be significantly influenced by this pre-existing difference in

nanocomposite homogeneity. To substantiate the authors' claim, Extended Data Figure 5 should include a direct comparison between the SEBS/QD and SEBS-g-MA/QD composites.

Our response to minor comment #4: We thank the reviewer for the insightful comments regarding the chemical interaction between SEBS-g-MA and the QDs, as well as the important point concerning the origin of the enhanced mechanical stretchability. We fully appreciate that appropriate control experiments are essential to validate the proposed interfacial bonding mechanism.

In response, we have expanded the FT-IR analysis to include pristine QDs, pristine SEBS-g-MA, the QD/SEBS nanocomposite, and the QD/SEBS-g-MA nanocomposite before and after the composite formation (**Extended Data Fig. 3b,c**). This expanded dataset now enables a direct comparison between systems with and without MA-derived functional groups, as well as before and after the composite formation process. Specifically, the characteristic anhydride C=O stretching bands of the succinic anhydride graft (~ 1780 and ~ 1870 cm^{-1}) are clearly observed in pristine SEBS-g-MA and in the QD/SEBS-g-MA nanocomposite prior to reaction, while these peaks are absent in both pristine QDs and the QD/SEBS nanocomposite (**Extended Data Fig. 3b**). After the reaction, these anhydride peaks are substantially diminished in the QD/SEBS-g-MA nanocomposite (**Extended Data Fig. 3c**), indicating consumption of the anhydride groups during composite formation. We attribute this change to ring-opening of the succinic anhydride groups during composite formation under sonication at 60 °C, followed by coordination to the QD surface (**Extended Data Fig. 3a**). These results provide direct spectroscopic evidence supporting covalent bonding between the QD surface ligands and the succinic acid moieties in SEBS-g-MA.

Consistent with this interpretation, the QD/SEBS-g-MA nanocomposite exhibits homogeneous dispersion with systematically increasing interparticle spacing as the elastomer content increases (**Extended Data Fig. 3d**), as well as enhanced solvent resistance compared with the QD/SEBS nanocomposite (**Extended Data Fig. 3e**).

We appreciate the reviewer's important point that the enhanced mechanical stretchability should not be interpreted as arising solely from chemical bonding. As noted, the AFM images in **Extended Data Fig. 4** reveal distinct differences in surface morphology between the QD/SEBS and QD/SEBS-g-MA composites even at 0% strain. In our system, however, these two effects are intrinsically coupled. Specifically, the succinic acid functional groups chemically bind to the QD surface and act as ligand-like compatibilizers already in the solution phase, which improves dispersion stability and suppresses aggregation prior to film formation. As a result, the QD/SEBS-g-MA nanocomposite forms a more homogeneous film after spin coating.

We have included additional FT-IR data in **Extended Data Fig. 3b and 3c**. In addition, we have revised the manuscript to clarify that the enhanced mechanical performance arises from combined effects of interfacial chemical bonding and improved nanocomposite homogeneity.

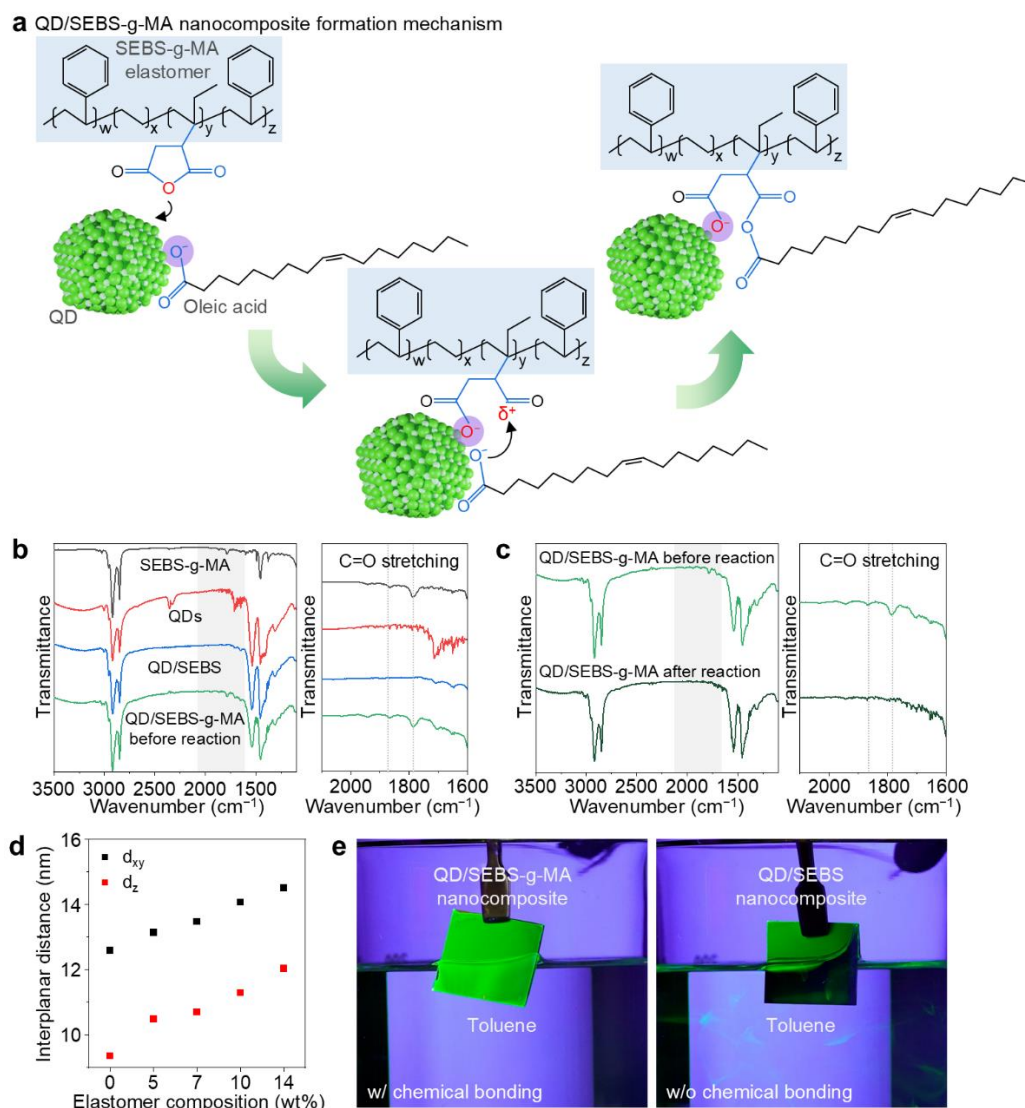
Our modification to the manuscript:

Page 6 in the revised manuscript:

“Colloidal QDs (**Extended Data Fig. 2**) are incorporated into polystyrene-block-poly(ethylene-ran-butylene)-block-polystyrene-graft-maleic anhydride (SEBS-g-MA) elastomer and spin-coated onto an octadecyltrichlorosilane (ODTS)-treated donor substrate to form composite films. In this structure, the succinic anhydride groups in the SEBS-g-MA undergo ring-opening during composite formation and become anchored to the QD surface (**Fig. 1f, Extended Data Fig. 3a, Methods**)^[51]. Fourier-transform infrared (FT-IR) spectroscopy shows that the characteristic anhydride C=O stretching bands (~ 1780 and ~ 1870 cm^{-1}) of the succinic anhydride graft, which are present in pristine SEBS-g-MA but absent in pristine QDs and QD/SEBS nanocomposite, are significantly diminished in QD/SEBS-g-MA after the composite formation (**Extended Data Fig. 3b,c**). Such chemical anchoring enhances structural integrity and solvent resistance, whereas physically blended QD/SEBS nanocomposites without MA grafting exhibit poor cohesion (**Extended Data Fig. 3d,e**).”

Page 8 in the revised manuscript:

“Chemical bonding within the QD/SEBS-g-MA nanocomposite enables homogeneous dispersion of QDs and outstanding mechanical stretchability exceeding 100%, while preserving a smooth surface morphology and uniform strain distribution under deformation (**Fig. 2a–c, Extended Data Fig. 4a**). In contrast, physically blended QD/SEBS nanocomposites exhibit pronounced QD aggregation and mechanical failure under strain (**Extended Data Fig. 4b**).”



Extended Data Fig. 3 | Chemical bonding between QDs and SEBS-g-MA elastomer. a, Schematic illustration of the ring-opening of SEBS-g-MA during composite formation, indicating chemical bonding between QDs and the elastomer matrix. **b,c**, FT-IR spectra of SEBS-g-MA, QDs, QD/SEBS, and QD/SEBS-g-MA before reaction (**b**), and of the QD/SEBS-g-MA before and after reaction (**c**). The right panels in (**b**) and (**c**) show magnified views of the anhydride C=O stretching band (~ 1780 and ~ 1870 cm^{-1}). **d**, Interplanar distances of QDs in the horizontal (d_{xy}) and vertical (d_z) directions in QD/SEBS-g-MA nanocomposites, measured by GISAXS as a function of SEBS-g-MA composition. **e**, Solvent resistance test of QD nanocomposite films with (left) and without (right) chemical bonding after immersion in toluene.

Minor comment #5: *A significant point of concern is the handling of the “positive aging effect” commonly observed in QLEDs utilizing ZnO electron transport layers. The device efficiency and stability are highly dependent on whether measurements are taken immediately after fabrication or after a stabilization period. The authors must explicitly state the timing of their J-V-L and EQE measurements relative to the device fabrication. Furthermore, to substantiate the performance claims, it is imperative that they provide data characterizing this aging process, such as the evolution of efficiency and luminance over time under constant operating conditions.*

Additionally, the manuscript fails to specify the substrate used for the core EL performance evaluation (e.g., Figure 2l-o, Extended Data Fig. 6). The properties of the substrate (e.g., rigid glass vs. flexible/stretchable polymer) fundamentally influence the device structure, processing, and ultimately, the extracted performance metrics. The absence of this fundamental information is a serious omission. This lack of transparency could be perceived as obscuring critical experimental conditions, thereby misleading readers in their interpretation of the device's capabilities and making it difficult to reproduce the work.

Our response to minor comment #5: We thank the reviewer for raising this important point regarding the positive aging effect commonly observed in ZnO-based QLEDs. We acknowledge that the efficiency and operational stability of such devices can be influenced by thermal or electrical stabilization processes after fabrication. In this work, all $J-V-L$ and EQE measurements of the intrinsically stretchable QLED devices were performed immediately after device fabrication without any intentional aging treatment or electrical preconditioning. This measurement protocol was adopted to evaluate the intrinsic as-fabricated device performance and to ensure consistent comparison across all devices, independent of post-fabrication aging effects. This clarification has been added to the revised manuscript.

We also clarify that the electroluminescence performance data presented for the conventional device architecture (**Fig. 2l-o and Extended Data Fig. 6**) were obtained from rigid devices fabricated on indium tin oxide (ITO)-coated glass substrates. These rigid devices were intentionally used to accurately assess the intrinsic optoelectronic properties of the QD EML and interfacial engineering effect without additional variables introduced by mechanical deformation or stretchable substrate-induced optical and electrical variations. In contrast, intrinsically stretchable devices were primarily employed to demonstrate mechanical deformability and system-level integration. This distinction has been explicitly clarified in the revised manuscript to ensure the experimental conditions.

Our modification to the manuscript:

Page 10 in the revised manuscript:

“To evaluate the effectiveness of our LIFT strategy for QLED fabrication, stretchable EMLs were integrated into conventional QLED architecture **fabricated on ITO-coated glass substrate** (ITO/PEDOT:PSS/TFB/QD nanocomposite/ZnO/Al; **Fig. 2l-o**).”

Page 19 in the revised manuscript:

“All EL measurements were performed immediately after device fabrication without any intentional aging treatment.”

Page 5 in the revised Supplementary information:

“The ITO-coated glass substrate was sequentially cleaned by sonication in detergent solution, deionized water, and isopropyl alcohol, followed by UVO treatment for 10 min.”

Minor comment #6: *The authors attribute the superior performance of the LIFT-processed devices, in part, to reduced current leakage compared to spin-coated films (Extended Data Fig. 6d-f). However, this conclusion may not be warranted. Spin-coating is a well-established method for fabricating high-performance QLEDs, with literature reports regularly demonstrating EQEs exceeding 30% and minimal leakage. The significantly poorer performance of the spin-coated control device in this work suggests it may not have been optimally fabricated. A more rigorous comparison would require the authors to demonstrate that their spin-coated device represents a state-of-the-art control, rather than a sub-optimal one. Without this, the claim that spin-coating inherently causes serious leakage in this context is not fully supported and weakens the credibility of the comparative analysis.*

Our response to minor comment #6: We thank the reviewer for this important comment and fully acknowledge that spin-coating is a well-established technique capable of producing state-of-the-art QLEDs with excellent efficiency and minimal leakage when conventional, additive-free QD inks are employed.

In the present work, however, mechanical stretchability is achieved by intentionally incorporating insulating elastomeric components (i.e., SEBS-g-MA) into the QD EML. These elastomeric additives inevitably introduce additional insulating barriers and reduce carrier transport efficiency, making it intrinsically more challenging to achieve high EQEs compared with conventional bare-QD emitters, irrespective of the deposition method. This behaviour is also consistent with previously reported LEDs employing stretchable EMLs, which generally exhibit low efficiencies (**Fig. R6**).

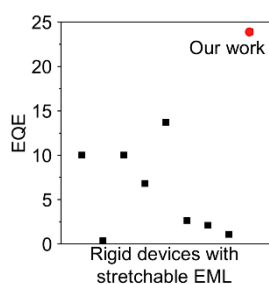


Fig. R6 | EQEs of LEDs employing stretchable EMLs.

In addition, unlike bare QD inks, the solvents required for processing stretchable QD nanocomposites present severe solvent-orthogonality challenges with the underlying HTL. In our system, the toluene-based QD nanocomposite solution partially dissolves the TFB layer even under minimized exposure conditions (e.g., dynamic casting), resulting in interfacial damage and increased leakage current in spin-coated control devices (**Fig. R7**).

By contrast, the LIFT process enables solvent-free transfer of the stretchable QD nanocomposite, thereby preserving the HTL interface and suppressing leakage pathways.

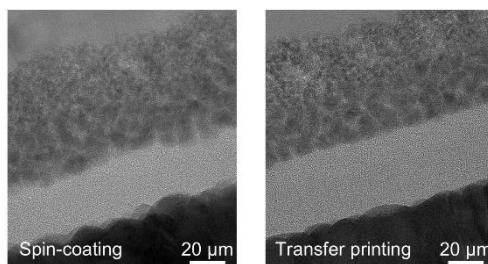


Fig. R7 | Cross-sectional TEM images of QLEDs fabricated by spin-coating (left) and transfer printing (right).

Therefore, the purpose of this comparison is to demonstrate the specific advantage of the LIFT process for elastomer-containing stretchable QD nanocomposites, where solvent-induced interfacial degradation becomes a critical limitation for direct solution deposition. These results indicate that the improved device performance originates from the ability of the LIFT process to preserve interfacial integrity during processing of stretchable nanocomposite EMLs, rather than from any inherent limitation of spin-coating itself.

***Minor comment #7:** Extended Data Fig. 8's analysis of the composite's mechanical stretchability, which employs a rigid ITO electrode on a non-stretchable substrate, appears misaligned with the study's objectives. To properly evaluate stretchability, the test structure should mirror the proposed flexible device architecture from Figure 4. Additionally, the experimental details regarding the integration of the pre-stretched EML into the rigid device are unclear. If the EML is pre-strained on PDMS (e.g., at 0%, 50%, and 100%) before device fabrication, this approach does not accurately represent the operational stability of a stretchable device and could undermine the reported advantages.*

Our response to minor comment #7: We thank the reviewer for this important comment. We would like to clarify that **Extended Data Fig. 8** is intended to evaluate the intrinsic strain tolerance of the LIFT-treated QD nanocomposite itself, rather than the full stretchable device architecture shown in **Fig. 4**. The evaluation of pre-strained EMLs is a well-established and widely adopted approach in studies of intrinsically stretchable light-emitting diodes (*Nature* **649**, 604 (2026); *Nat. Commun.* **15**, 7990 (2024); *Nat. Mater.* **22**, 737 (2023); *Sci. Adv.* **9**,

eadh1504 (2023); *Nature* **603**, 624 (2022); *Adv. Mater.* **34**, 2201844 (2022)). Such measurements allow direct evaluation of how mechanical deformation influences the intrinsic optical and electronic properties of the EML while decoupling the mechanically limiting effects of electrodes, substrates, interconnection, and encapsulation layers.

In this context, the use of a rigid ITO-coated glass substrate in **Extended Data Fig. 8e–h** was intentional, as it enables characterization of the effects of mechanical deformation of the EML under well-controlled electrical conditions. The purpose of this experiment is therefore not to represent the operational stability of the complete stretchable device, but rather to isolate the deformation-induced changes in the emissive layer itself.

The operational stability of the full intrinsically stretchable device architecture is separately demonstrated in **Fig. 4d–f** and **Extended Data Fig. 9g**, where the complete stretchable QLEDs are evaluated under tensile deformation and cyclic stretching conditions. Together, these two sets of measurements provide complementary information: **Extended Data Fig. 8** evaluates the intrinsic deformation tolerance of the EML material, whereas **Fig. 4** validates the performance of the fully integrated stretchable device.

The experimental details regarding the integration of the pre-stretched EML into the rigid device have been stated in the revised manuscript to clarify the experimental conditions.

Our modification to the manuscript:

Page 5 in the revised supplementary information:

“... The LIFT-composite film was transfer-printed onto the TFB layer. For the integration of pre-stretched EMLs, the QD/SEBS-g-MA nanocomposite film was first picked up using a low-modulus PDMS stamp. After LIFT process, the EML was mechanically pre-stretched on the PDMS stamp to the desired strain level and subsequently transfer-printed onto the TFB layer of the device architecture. The applied strain levels (from 0 to 100%) were controlled by mechanical elongation of the PDMS stamp prior to transfer, enabling systematic evaluation of the deformation-dependent optoelectronic properties of the EML.”

Minor comment #8: The authors’ successful optimization of the EML’s stretchability stands in stark contrast to the neglect of other critical device components, most notably the electron transport layer. This represents a fundamental limitation of the presented strategy. The reported high luminance and EQE are achieved in a device architecture that remains critically dependent on the ZnO nanoparticle layer, which is intrinsically brittle and lacks any meaningful stretchability. Consequently, the superior performance metrics are primarily demonstrated on rigid glass substrates, which severely undermines the practical relevance and claimed applicability of this research for genuinely stretchable LEDs.

Our response to minor comment #8: We thank the reviewer for the important comment regarding the mechanical compliance of the electron transport layer (ETL) and its relevance to the practical realization of fully stretchable QLEDs.

We would first like to clarify that pristine ZnO nanoparticle films were not used as the ETL in the intrinsically stretchable device architecture. Instead, the stretchable QLEDs employ a ZnO/PFN-Br composite ETL (20 wt%), which was specifically designed to enhance mechanical compliance while maintaining efficient electron transport.

As correctly pointed out by the reviewer, bare ZnO nanoparticle films are intrinsically brittle and show pronounced crack formation at relatively low strains (~20%; **Supplementary Fig. 31a**). In contrast, the ZnO/PFN-Br composite ETL exhibits substantially improved mechanical deformability, with no noticeable large-scale crack formation under comparable strain conditions (**Supplementary Fig. 31b**). AFM images and the corresponding height profiles further confirm the absence of significant fracture features in the composite ETL under deformation (**Supplementary Fig. 31c,d**). Importantly, even under 60% tensile strain, no crack propagation from the ETL into the QD nanocomposite emissive layer (EML) is observed (**Fig. R8**), thereby enabling stable device operation up to ~65% strain.

Therefore, the high luminance and EQE reported in the stretchable device are not based on a brittle ZnO-only ETL, but are achieved within a fully stretchable device architecture incorporating the mechanically compliant composite ETL. These additional experimental results have now been incorporated into **Supplementary Fig. 31**.

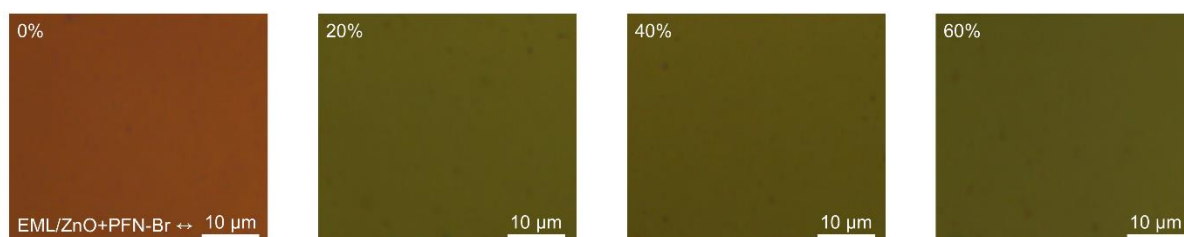
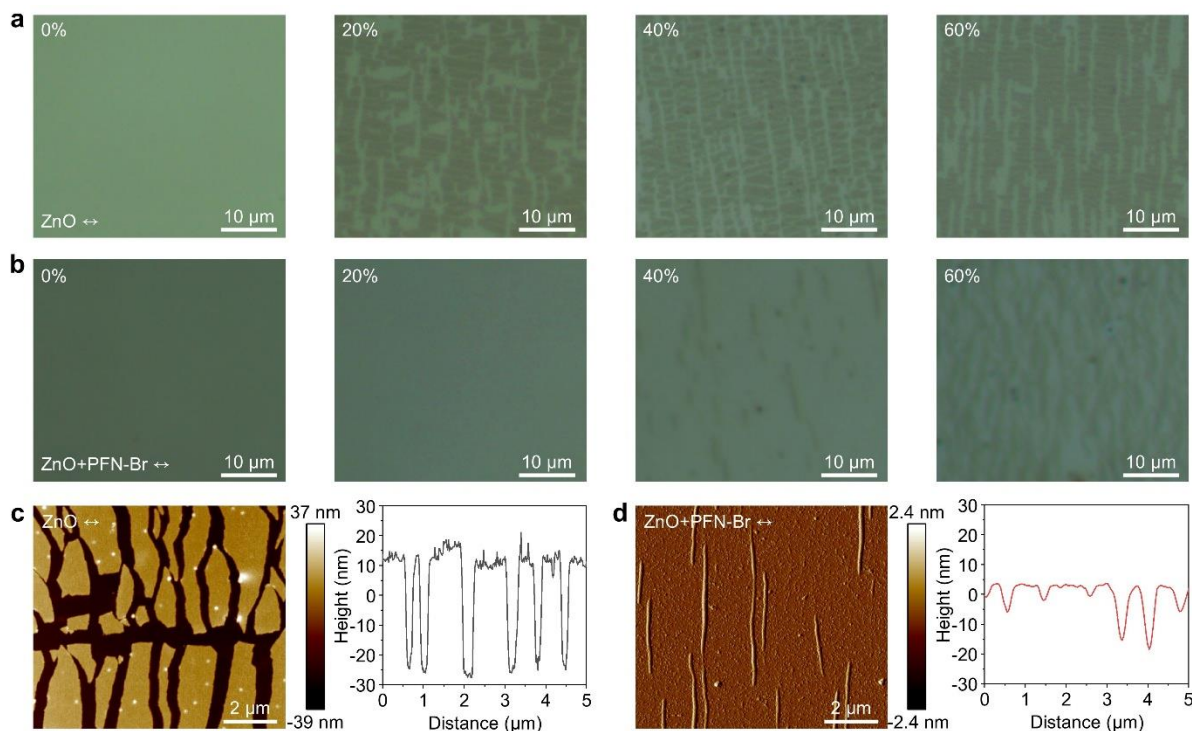


Fig. R8 | Optical microscope images of EML with ZnO/PFN-Br composite under stretching.

We further clarify that the performance comparison presented in **Fig. 1i** is based on fully stretchable QLED devices using the composite ETL in the stretchable architecture described in **Fig. 4**, rather than rigid glass-based devices. The reported luminance and efficiency values are benchmarked against previously reported fully stretchable light-emitting devices to enable a fair comparison within the same device class. Although the performance of fully stretchable devices does not yet reach that of state-of-the-art rigid QLEDs, this limitation reflects the broader materials constraints currently inherent to intrinsically stretchable optoelectronic systems. The present work demonstrates that high efficiency and high-resolution patterning can be simultaneously achieved within a fully stretchable QLED architecture.

Page 14 in the revised Supplementary information:

“The stretchability of the ETL was enhanced by blending ZnO nanoparticles with PFN-Br to form a hybrid inorganic–organic structure. While bare ZnO nanoparticle films undergo severe microscale cracking and tearing at tensile strains below 20%, the ZnO/PFN-Br hybrid ETL suppresses crack propagation, exhibiting only mild wrinkling and localised, non-propagating cracks under comparable strain conditions (**Supplementary Fig. 31**).”



Supplementary Fig. 31 | Mechanical stretchability of ETL layers. a, Optical microscope images of bare ZnO ETL layers under different tensile strains. **b,** Optical microscope images of ZnO/PFN-Br hybrid ETL layers under different tensile strains. **c,d,** AFM images of ZnO (c) and ZnO/PFN-Br hybrid ETL layers (d) under 40% tensile strain, with corresponding height profiles (right).

Thank you again for your insightful comments. We feel that these comments have helped to improve the quality of the manuscript significantly.

Reviewer #2:

***Summary Comments:** The proposed LIFT strategy is innovative, combining ligand engineering with thermal transfer printing. This approach presents a well-conceived solution to address current limitations in pixelation and performance of stretchable quantum dot displays, likely to have a positive impact on the development of next-generation display technologies. However, several critical points need to be addressed before the manuscript can be considered for publication. I recommend the following revisions:*

Our response to summary comments: We sincerely thank the reviewer for the thoughtful and constructive evaluation of our work and for recognizing the potential impact of the proposed LIFT strategy on stretchable display technologies.

In response to the reviewer's comments, we have conducted additional experiments and revised the manuscript to address the key points raised, including strain-dependent behaviour and its reversibility, mechanical reliability under cyclic deformation of the intrinsically stretchable quantum dot light-emitting diodes (QLEDs), and clarification of the patterning mechanism in comparison with previous methods.

The manuscript has been revised accordingly to strengthen the experimental evidence and improve the clarity of the discussions.

***Comment #1:** It is suggested that the authors include cross-sectional SEM images of the intrinsically stretchable QLEDs employing AgNWs as the anode. These types of images will help readers better understand device architecture.*

Our response to comment #1: We thank the reviewer for this helpful suggestion. In response, we have acquired cross-sectional SEM images of the intrinsically stretchable QLEDs employing silver nanowires (AgNW)-based anodes (**Supplementary Fig. 30**). These images clearly reveal the multilayer device architecture, including the PUA encapsulation layer, the AgNW anode embedded within the elastomer matrix, the active layers, and the top electrode.

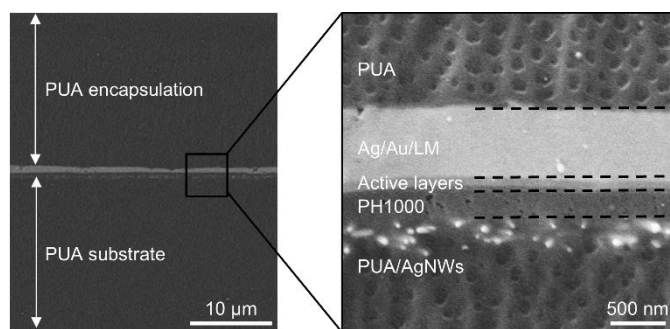
The additional data improve the clarity and structural understanding of the intrinsically stretchable QLED architecture. We have included the obtained images in the **Supplementary Fig. 30**.

Our modification to the manuscript:

Page 14 in the revised manuscript:

“The device structure consisted of silver nanowires (AgNWs)/conductive PEDOT:PSS (PH1000), PEDOT:PSS, TFB, QD nanocomposite, ZnO/poly(9,9-bis(3'-(N,N-dimethyl)-N-ethylammonium-propyl-2,7-fluorene)-alt-2,7-(9,9-dioctylfluorene))dibromide (PFN-Br), and Ag/Au/liquid metal (LM), serving as the stretchable anode, HIL, HTL, EML, ETL, and stretchable cathode, respectively (**Supplementary Fig. 30**).”

Page 40 in the revised Supplementary information:



Supplementary Fig. 30 | Cross-sectional SEM images of the intrinsically stretchable QLED in different magnifications.

Comment #3: *In Figure 1i and Extended Data Table 1, the authors compare the performance of their device with other intrinsically stretchable light-emitting devices. However, PeLEDs are not depicted in Figure 1i but instead illustrated in Extended Data Table 1, which creates confusion. Additionally, a more thorough review of relevant literature should be provided to contextualize the significance of this work.*

Our response to comment #3: We thank the reviewer for this helpful comment. As suggested by reviewer, we have included intrinsically stretchable perovskite light-emitting diodes in **Fig. 1i** to ensure a consistent and intuitive comparison across different classes of stretchable light-emitting devices.

In addition, we have expanded the literature survey to incorporate recently reported intrinsically stretchable LEDs and have updated both **Fig. 1i** and **Extended Data Table 1** accordingly.

Our modification to the manuscript:

Page 3 in the revised manuscript:

“However, these modifications frequently hinder charge transport due to the insulating nature of the added components, disrupted molecular ordering, and suboptimal surface morphology, resulting in compromised device performance with typical luminance values below 34,000 cd m⁻²[35].”

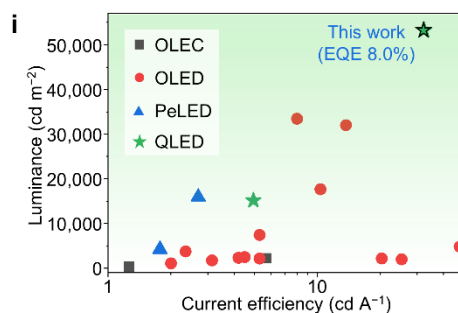


Fig. 1 | i, Performance comparison of the intrinsically stretchable QLEDs fabricated via LIFT with previously reported direct current-driven intrinsically stretchable light-emitting devices, including organic LECs (OLECs), organic LEDs (OLEDs), and **perovskite LEDs (PeLEDs)**, highlighting superior luminance and current efficiency.

Extended Data Table 1 | Performance of previously reported intrinsically stretchable light-emitting devices. Max., maximum; Ref., reference; PeLED, perovskite LED.

Device type	Journal [year]	Emitters	Additives	Max. luminance [cd m ⁻² , @V]	Turn-on voltage [V]	Max. EQE [%]	Max. current efficiency [cd A ⁻¹]	Max. stretchability [%]	Ref. #
OLEC	Adv. Mater. (2011)	PF-B	PEO-DMA/LiTf	300 (@12)	4.8	-	1.26	45	[55]
	Nat. Photon. (2013)	Super Yellow	ETPTA/PEO/LiTf	2,200 (@21)	6.8	-	5.7	120	[56]
	ACS Appl. Mater. Interfaces (2016)	SY-PPV	PEO/ETPTA/LiTf	~300 (@21)	13	-	-	50	[57]
	Nat. Commun. (2020)	Super Yellow	Ion-conducting polymer/ETT-15/LiTf	-	15	-	-	30	[38]
OLED	ACS Nano (2014)	White emitting polymer	OXD-7	1,100 (@21)	7	-	2.0	130	[58]
	Sci. Adv. (2021)	Super Yellow	Triton-X	2,500 (@11)	8.3	-	4.5	80	[25]
	Polym. Int. (2021)	PF-b-PI	-	70 (@10)	2	-	-	140	[20]
	Adv. Mater. (2022)	Super Yellow	Triton-X	2,185 (@15)	5.3	-	20.3	73	[27]

	Adv. Mater. (2022)	L-SY-PPV	PAN	3,780 (@13)	6.5	0.69	2.35	30 (serpentine)	[32]
	Nature (2022)	Super Yellow	PU	7,450 (@15)	5	-	5.3	100	[30]
	Adv. Funct. Mater. (2023)	Super Yellow	PEO/KCF3SO3	1,754 (@11)	3.7	-	3.14	30	[28]
	Nat. Mater. (2023)	PDKCD	-	2,175 (@10)	4.7	1.7	5.3	60	[22]
	Angew. Chem. Int. Ed. (2023)	SBSPy-n	-	-	-	-	-	54	[21]
	npj Flex. Electron. (2024)	Super Yellow	Triton-X	2,340 (@9.0)	6	-	4.2	70	[59]
	Adv. Mater. (2025)	Super Yellow	PU/PCL	32,013 (@9.0)	3.7	4.7	13.7	50	[60]
	Nat. Mater. (2025)	PDKCD	-	~2,000 (@10.0)	3.5	8.0	25.3	80	[24]
	Adv. Mater. (2025)	Super Yellow	Triton-X	17,670 (@~10.0)	3.0	-	10.35	60	[29]
	J. Mater. Chem. C (2026)	Super Yellow	SBS/Triton-X	1,703 (@~15.0)	9.2	-	-	100	[33]
	Nature (2026)	TCTA/TPBi/Ir(ppy) ₂ acac	PU	4,840 (@~15.0)	~5.0	17.0	47.9	60	[34]
	Light Sci. Appl. (2026)	Poly(9,9-dialkylfluorene) derivative	SBS	33,443 (@~11.0)	~4.0	2.3	~8.0	120	[35]
PeLED	Adv. Mater. (2017)	MAPbBr ₃	PEO	15,960 (@8.0)	2.4	-	2.7	40	[61]
	J. Alloys Compd. (2020)	MAPbBr ₃	PEO	4,257 (@7.0)	2.5	-	1.77	70	[62]
	ACS Appl. Mater. Interfaces (2020)	MAPbBr ₃	PEO/PVP	380.5 (@7.5)	2	0.27	-	100	[44]
QLED	Nat. Electron. (2024)	CdSe/ZnS QD	SEBS-g-MA/TFB	15,170 (@6.2)	3.2	1.2	4.95	50	[39]
	This work	CdSe/ZnS QD	SEBS-g-MA	53,300 (@7.0)	3.0	8.0	32.41	65	

Comment #4: In Figure 4d, the strain-luminance-current density curves demonstrate trends that differ significantly. The luminance improves by more than 1.5 times with increasing strain but decreases to 0.6 times at a strain of 70%. This behavior requires a detailed explanation in the manuscript. Furthermore, it would be valuable to report the changes in luminance after strain release to assess the reversibility of performance.

Our response to comment #4: We thank the reviewer for this important comment regarding the strain-dependent luminance behaviour in intrinsically stretchable QLEDs.

The strain-dependent luminance evolution can be understood in two distinct deformation regimes. At low to moderate tensile strain, the initial increase in luminance is primarily attributed to mechanically induced thinning of the device stack, which enhances interfacial contact between adjacent functional layers and increases the effective electric field across the emissive region, thereby facilitating carrier injection (*Nat. Electron.* **7**, 365 (2024)). In addition, the QDs embedded within the nanocomposite EML exhibit a modest increase in packing density under tensile deformation (**Fig. R1a**), which is consistent with the behaviour observed in rigid QLEDs incorporating pre-strained EMLs (**Fig. R1b–e**). These combined effects account for the enhanced luminance observed at moderate strain levels.

At higher tensile strains, however, the luminance gradually decreases, which we attribute primarily to the strain-induced increase in device resistance and the resulting limitation in efficient charge transport across the multilayer stack.

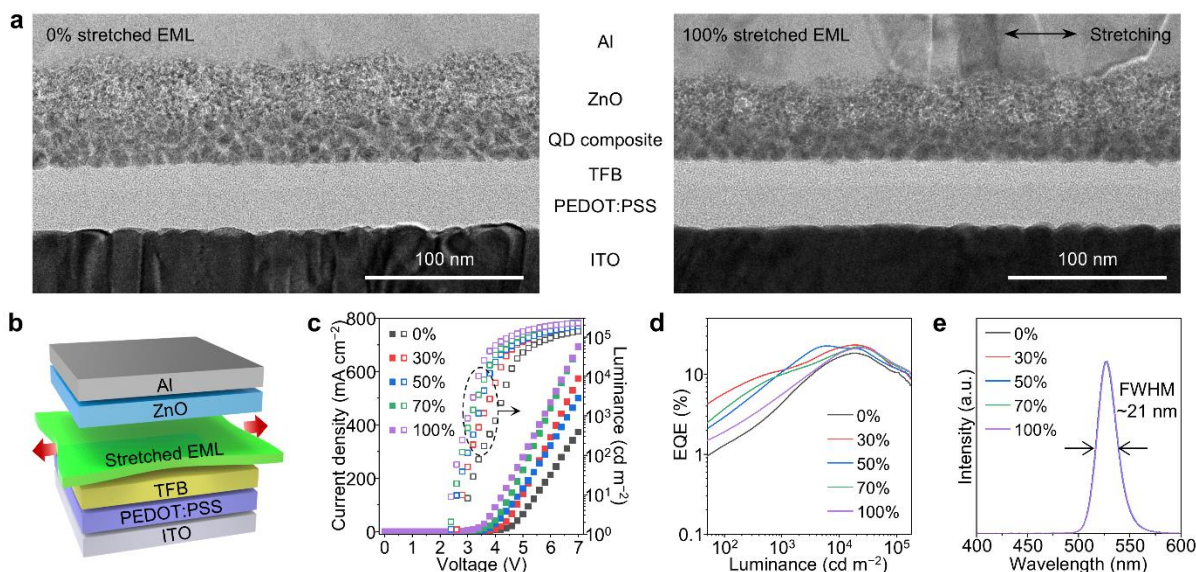


Fig. R1 | **a**, Cross-sectional TEM images of QLEDs incorporating EMLs pre-stretched to 0% (left) and 100% (right). **b–e**, QLEDs incorporating pre-stretched EML under different strain conditions (0%, 30%, 50%, 70%, and 100%): schematic illustration of device structure (**b**), J – V characteristics (**c**), EQE– L curves (**d**), and EL spectra (**e**).

To address the reviewer’s suggestion regarding reversibility, we further evaluated the device performance after strain release. The luminance largely recovers after release from 30% tensile

strain (**Fig. 4f**), indicating that the QD nanocomposite EML fabricated via the LIFT process maintains its structural and optoelectronic integrity under reversible mechanical deformation.

These explanations and additional experimental results have been incorporated into **Fig. 4f** and the revised manuscript.

Our modification to the manuscript:

Page 15 in the revised manuscript:

“The devices exhibit an initial increase in luminance upon stretching due to the enhanced electric field arising from thickness reduction during tensile deformation^[39], followed by a gradual decrease at higher tensile strains as increased device resistance progressively limits charge transport across the multilayer stack. Upon release, the voltage–luminance characteristics remain nearly unchanged, demonstrating the outstanding mechanical robustness and reversibility of the LIFT-composite EML (**Fig. 4f**).”

Page 29 in the revised manuscript:

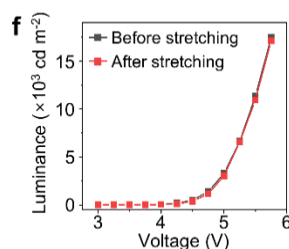


Fig. 4 | *I*, *L*–*V* characteristics of stretchable QLEDs before stretching and after release from 30% strain.

Comment #2: *The authors attribute the enhanced mechanical robustness and stable resistance during stretching to partial alloying at the Ag/Au/LM interface. However, the conclusions drawn from a single tensile test are insufficient. It is recommended to include cyclic tensile stability tests to evaluate the long-term mechanical reliability of the device under repeated deformation.*

Comment #5: *To confirm the mechanical stability of the device at each strain level, it is recommended to include data on mechanical reliability under repeated cycles of deformation.*

Our response to comment #2 and #5: We appreciate the reviewer for these valuable comments regarding the mechanical reliability of the device under repeated deformation. We agree that cyclic tensile testing is essential to evaluate the long-term mechanical stability of intrinsically stretchable devices. In response, we have performed cyclic tensile reliability tests at multiple structural levels, including the electrode, device, and material levels.

At the electrode level, the partially alloyed Ag/Au/LM electrode exhibits highly stable resistance during 400 stretching cycles at 40% tensile strain, with less than 10% variation in

resistance (**Extended Data Fig. 9f**), confirming the robustness of the alloyed interface under repeated deformation.

At the device level, we evaluated cyclic reliability by monitoring the luminance of intrinsically stretchable QLEDs under repeated tensile deformation. The devices retain their initial luminance after 100 stretching cycles at 30% tensile strain (**Extended Data Fig. 9g**, **Supplementary Video 1**), demonstrating stable optoelectronic performance under cyclic strain.

At the material level, AFM characterization shows no noticeable increase in surface roughness or crack formation in the QD nanocomposite EML even after 10,000 stretching cycles at 50% strain (**Fig. R2**), further confirming the intrinsic mechanical robustness of the system.

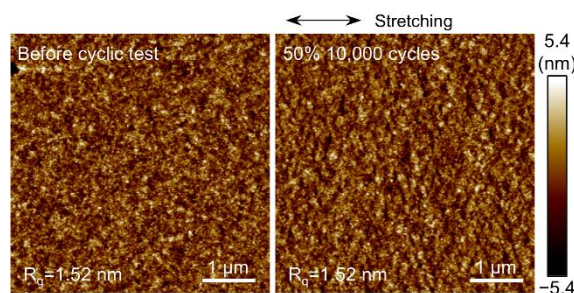


Fig. R2 | AFM images of QD nanocomposite films before (left) and after (right) 10,000 cycles of 50% uniaxial stretching.

Taken together, these results provide comprehensive evidence of the long-term mechanical reliability of the stretchable electrode, emissive layer, and fully integrated device under repeated deformation.

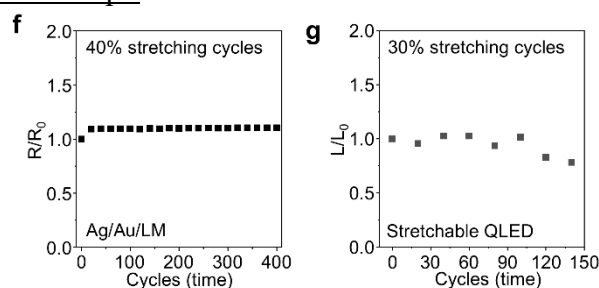
These results have been incorporated into **Extended Data Fig. 9f, g**, **Supplementary Video 1**, and the revised manuscript.

Our modification to the manuscript:

Page 15 in the revised manuscript:

“Furthermore, the devices retain their initial luminance over more than 100 cycles of 30% tensile strain (**Extended Data Fig. 9g**).”

Page 38 in the revised manuscript:



Extended Data Fig. 9 | f, Resistance change of stretchable cathode during cyclic stretching at 40% strain. g, Luminance change of the intrinsically stretchable QLED during cyclic stretching at 30% strain.

***Comment #6:** The authors report fabricating monochromatic pixel arrays with resolutions of up to 16,000 pixels per inch and RGB subpixel arrays with 2,565 pixels per inch. However, these values are similar to those reported in their previous work (Nature Photonics, 18(10), 1105-1112), where RGB pixelated patterns with 2,565 pixels per inch and monochromatic QD patterns with approximately 20,526 pixels per inch were achieved. Is it feasible to achieve the same or higher pixel density? If not, what are the associated challenges?*

Our response to comment #6: We thank the reviewer for the insightful comment regarding the achievable pixel density compared to our previous work.

In conventional intaglio transfer printing, the maximum pixel density is fundamentally governed by the fracture-assisted release mechanism of the QD film at trench edges (*Nat. Commun.* **6**, 7149 (2015)). In our previous study, the brittle nature of QD films enabled clean and well-defined cleavage at the trench boundaries, thereby facilitating ultrahigh-resolution patterning (**Supplementary Fig. 19, bottom left**).

In contrast, the intrinsically stretchable QD nanocomposite films employed in this work possess substantially higher mechanical compliance, which is essential for deformability but inherently suppresses crack initiation and promotes crack blunting at the trench edges (**Supplementary Fig. 19, bottom right**). As a result, there exists an intrinsic trade-off between maximum achievable patterning resolution and mechanical stretchability.

To address this limitation, the LIFT process introduces localized strain amplification at the trench edges via thermally induced expansion of the PDMS stamp, while simultaneously enhancing interfacial adhesion via MPA-mediated ligand engineering. This synergistic effect enables improved edge definition in stretchable systems, as supported by both finite element simulations and experimental observations (**Supplementary Fig. 20, 21, 23–26**). These clarifications and additional experimental results have been incorporated into **Supplementary Fig. 19, 20, 21, and 23–26** and revised manuscript.

Using this strategy, we achieved RGB subpixel arrays with a density of 2,565 PPI (3 μm subpixel size) and monochromatic pixel arrays with a density of 16,000 PPI (700 nm feature size). Although the absolute maximum monochromatic pixel density remains slightly lower than that of our previous rigid brittle-QD system, the achieved resolution is already sufficient for high-resolution near-eye display applications, where the practical requirement is typically $\sim 4,000$ PPI.

Therefore, the key advance of the present work lies not in surpassing the absolute maximum pixel density of brittle QD systems, but in extending ultrahigh-resolution patterning capability to intrinsically stretchable nanocomposite platforms while simultaneously achieving high luminance, high efficiency, and mechanical deformability.

Our modification to the manuscript:

Page 12 in the revised manuscript:

“In conventional intaglio transfer printing using bare QDs, the brittle nature of the QD films leads to clean fracture at the edge of trench, resulting in well-defined patterns (**Fig. 1d**, **Supplementary Fig. 19**)^[47,48]. In contrast, when using stretchable QD nanocomposites, the elastic polymer matrix suppresses localised fracture and instead promotes film delamination and elongation near the trench boundaries, thereby degrading pattern fidelity and limiting the maximum achievable pixel density (**Fig. 3a, top**).”

Page 12 in the revised manuscript:

“Finite element simulations reveal that this differential thermal expansion generates pronounced strain concentration within the QD nanocomposite film near the trench boundaries, thereby facilitating clean cleavage during pattern formation (**Fig. 3b**, **Supplementary Fig. 20**). Consistent with this mechanism, temperature-dependent experiments show that pattern fidelity progressively improves with increasing processing temperature, in agreement with the enhanced strain localisation predicted by the simulations (**Supplementary Fig. 21**). Concurrently, interfacial adhesion is reinforced via ligand engineering with MPA, whose polar carboxyl end groups form stronger interactions with the SiO₂ trench surface than untreated composites or composites treated with short-chain ligands lacking terminal functional groups (**Fig. 3c**, **Supplementary Fig. 22**, **Supplementary Method**). This enhanced interfacial adhesion broadens the processing window and enables robust pattern formation over a wider temperature range (**Supplementary Fig. 23–26**).”

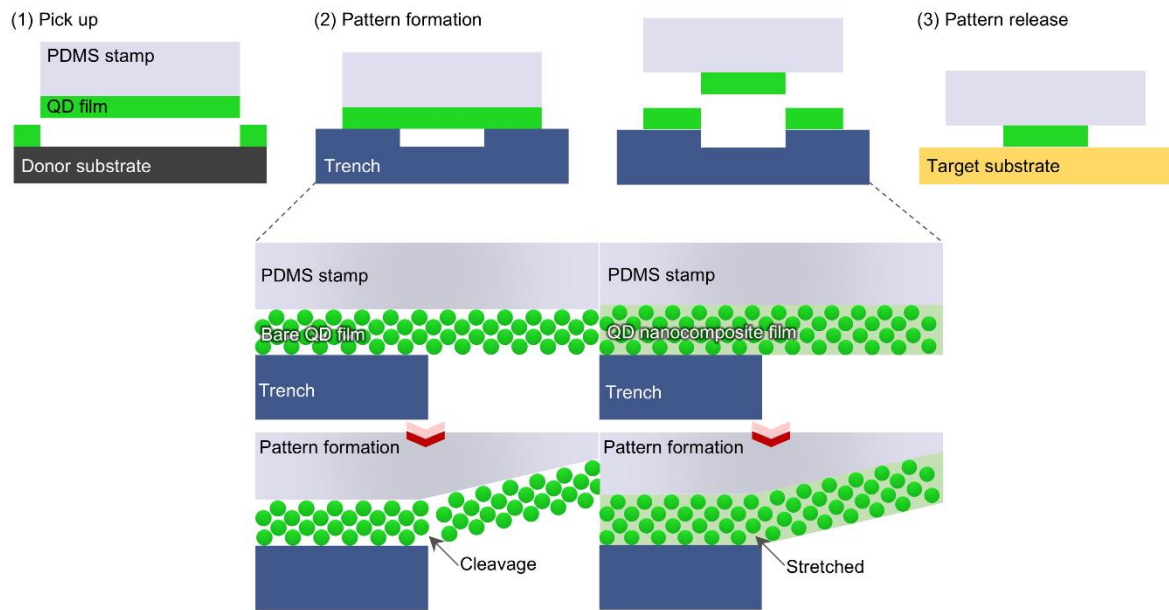
Page 6 in the revised Supplementary information:

“Finite element analysis

Thermo-mechanical strain and stress distributions during trench-assisted transfer of QD/elastomer composite films were simulated using commercial finite element software (Abaqus 2022, Dassault Systèmes, France) through a sequentially coupled thermal–mechanical approach. A two-dimensional plane strain model was constructed to represent the cross-sectional geometry of the PDMS stamp, QD composite layer, and underlying trench. The geometry was discretized using 4-node bilinear plane strain quadrilateral elements with reduced integration (CPE4R) for the mechanical analysis and corresponding 4-node heat transfer elements (DC2D4) for the thermal analysis. The final mesh consisted of 4,446 elements and 4,566 nodes, with locally refined meshing near the trench corner to accurately capture steep strain gradients. To realistically account for localised annealing, a steady-state heat transfer analysis was first performed, where a constant temperature corresponding to the annealing condition (60–120 °C) was prescribed at the trench region. Because the actual PDMS stamp thickness (~1 cm) is substantially larger than the modeled thickness (1 μm) used for computational efficiency, the truncated top surface of the PDMS domain was treated using an equivalent surface film boundary condition defined as $h_{eq} = k_{PDMS} / L_{omitted}$, where k_{PDMS} is the thermal conductivity of PDMS and $L_{omitted}$ represents the unmodeled bulk thickness. This

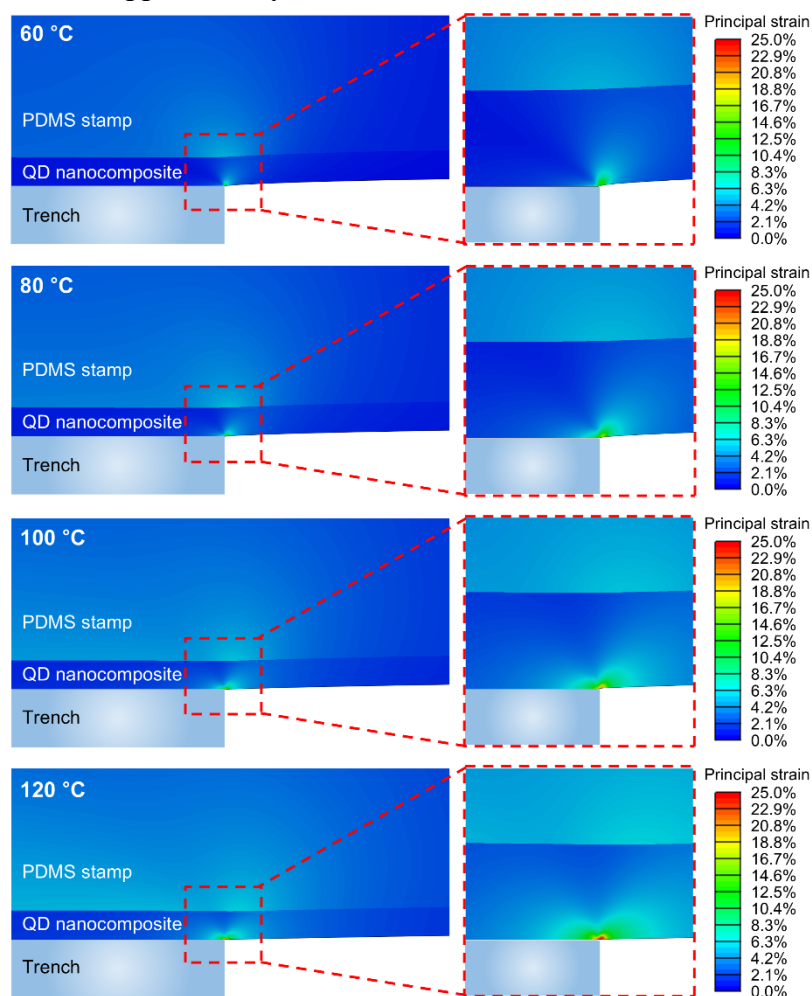
approach approximates conductive heat dissipation into the bulk PDMS and enables formation of a physically realistic temperature gradient. The resulting nodal temperature field was extracted from the thermal analysis and imported into a subsequent static mechanical step as a predefined temperature field. An initial uniform temperature of 25°C was assigned as the stress-free reference state, and thermal strain was incorporated through isotropic thermal expansion, $\epsilon_{th} = \alpha(T - 25^\circ\text{C})$. Mechanical boundary conditions were applied to prevent rigid body motion while allowing thermally induced deformation. Strain localisation near the trench edge was evaluated by extracting the maximum in-plane principal strain and comparing it with the far-field strain within the PDMS bulk to assess the influence of annealing temperature on deformation contrast governing pattern transfer.”

Page 29 in the revised Supplementary information:



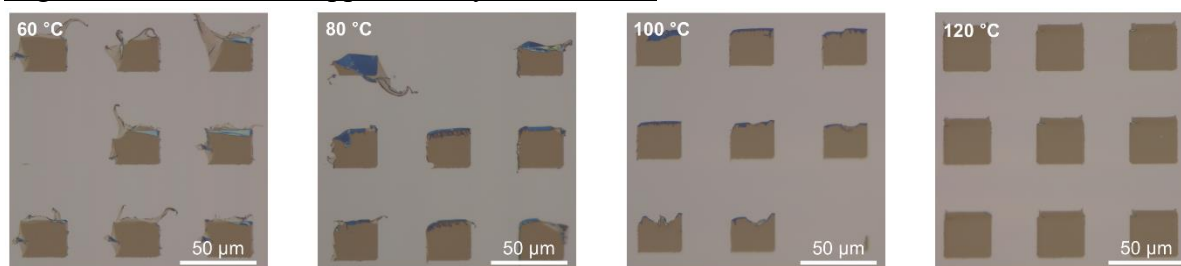
Supplementary Fig. 19 | Comparison of intaglio patterning for bare QD films (bottom left) and QD nanocomposite films (bottom right).

Page 30 in the revised Supplementary information:



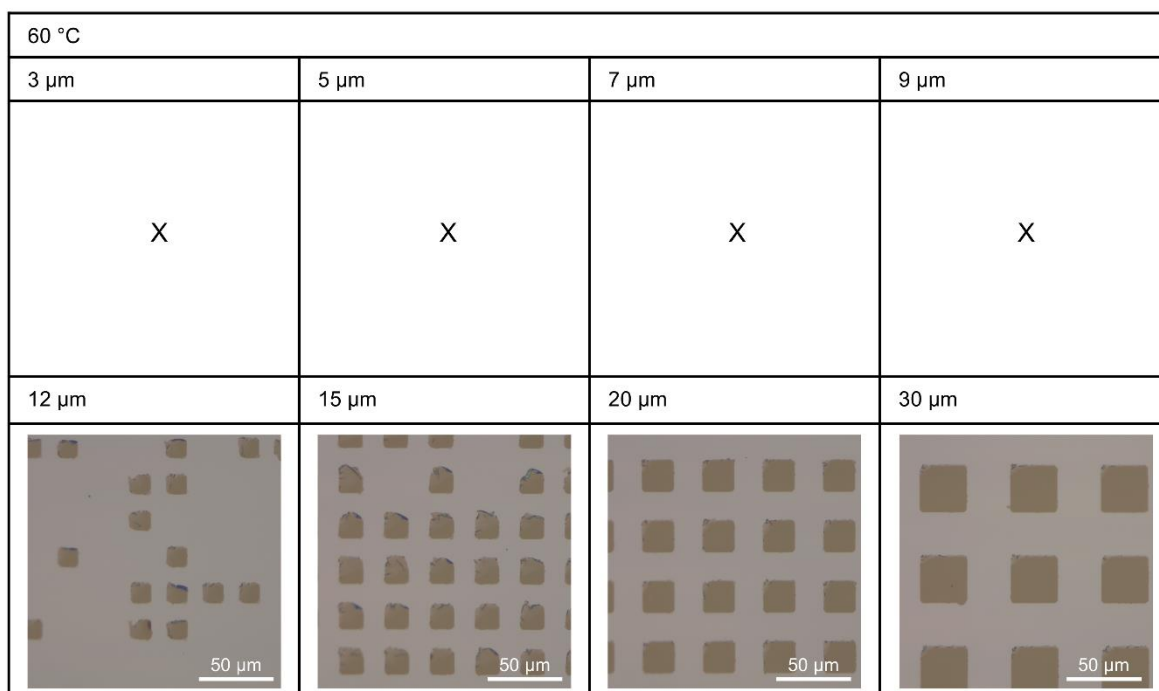
Supplementary Fig. 20 | Finite element analysis of localised strain distribution at trench edges during pattern formation under different processing temperatures.

Page 31 in the revised Supplementary information:



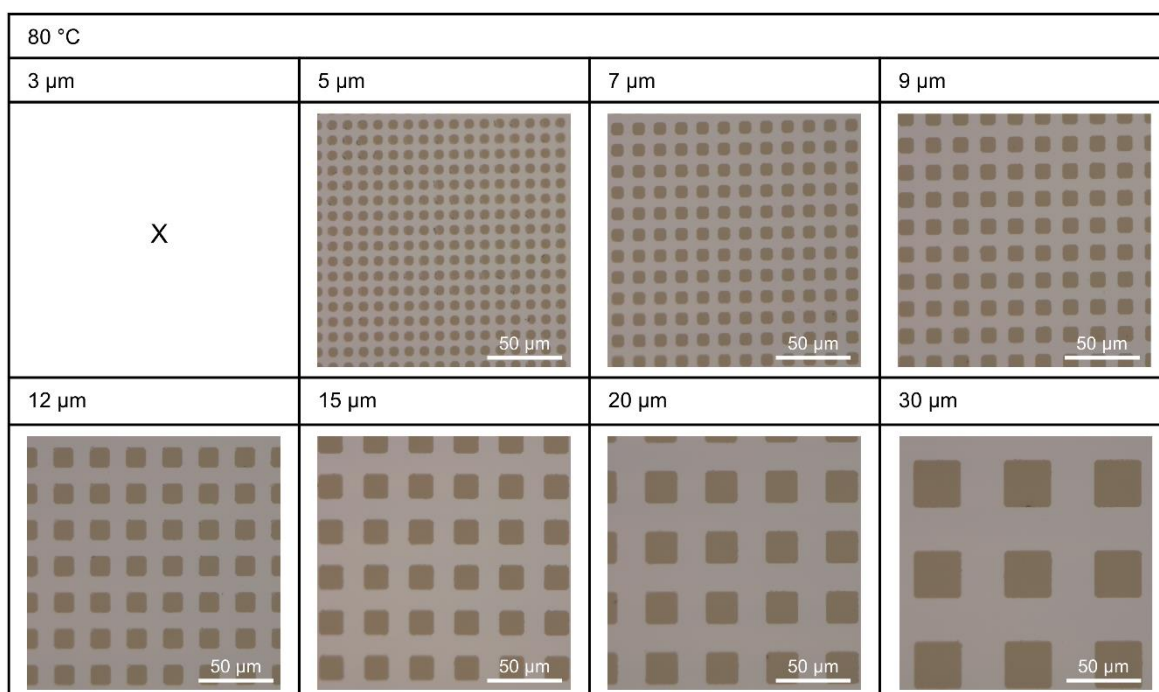
Supplementary Fig. 21 | Optical microscope images showing the temperature-dependent evolution of pattern fidelity in QD nanocomposite films processed at 60, 80, 100, and 120 °C.

Page 33 in the revised Supplementary information:



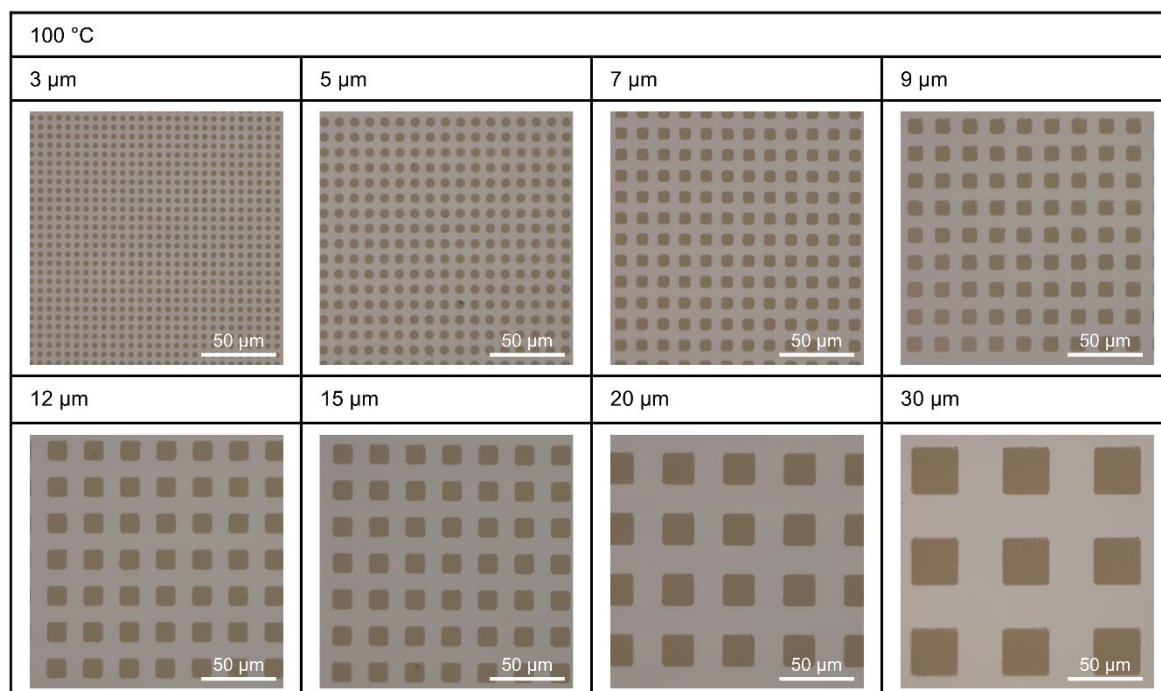
Supplementary Fig. 23 | Optical microscope images showing the effect of feature size on pattern fidelity in QD nanocomposite films processed at 60 °C.

Page 34 in the revised Supplementary information:



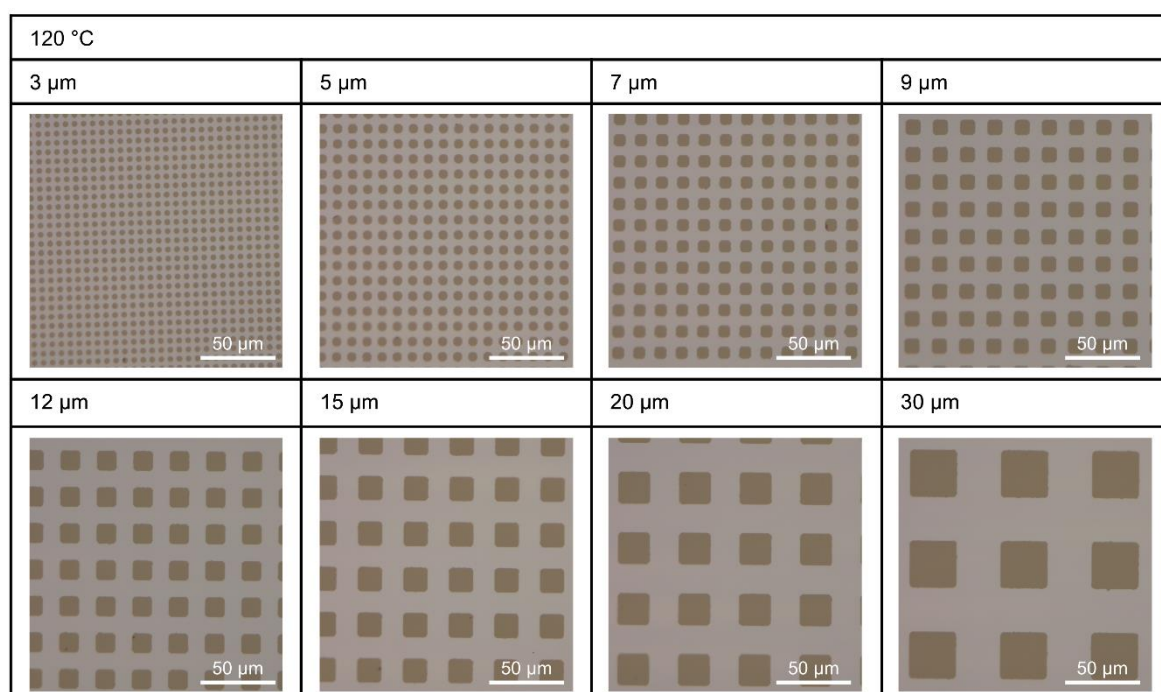
Supplementary Fig. 24 | Optical microscope images showing the effect of feature size on pattern fidelity in QD nanocomposite films processed at 80 °C.

Page 35 in the revised Supplementary information:



Supplementary Fig. 25 | Optical microscope images showing the effect of feature size on pattern fidelity in QD nanocomposite films processed at 100 °C.

Page 36 in the revised Supplementary information:



Supplementary Fig. 26 | Optical microscope images showing the effect of feature size on pattern fidelity in QD nanocomposite films processed at 120 °C.

Thank you again for your insightful comments. We feel that these comments have helped to improve the quality of the manuscript significantly.

Reviewer #3:

Summary Comments: *In this work, Yoo et al. report high-performance, high-resolution, intrinsically stretchable QLEDs using QD/elastomer composites as the emissive layer and intaglio-transfer printing for patterning. The patterning resolution exceeds 10,000 ppi for single-color QD stripes and >2,000 ppi for full-color QD patterns. Key device metrics—turn-on voltage, luminance, and EQE—surpass those reported for stretchable OLEDs, OLECs, and QLEDs. As the authors argue, their materials design and patterning strategy address a central challenge in intrinsically stretchable LEDs: the long-standing trade-off between high brightness and high patterning resolution.*

In 2024, some of the same authors published a related study in Nat. Electron. (2024, 7, 365) employing a similar materials combination (QDs and SEBS-g-MA). In my view, the major differences between the present submission and the Nature Electronics (2024) work are as follows. (1) Materials/composite design. Here, the authors perform a post-deposition ligand exchange, replacing SEBS-g-MA at the QD film surface with mercaptopropionic acid (MPA). Substituting the long SEBS-g-MA chains with short MPA ligands at the interface lowers the hole-injection barrier and supports the improved QLED performance. By contrast, the Nature Electronics (2024) study employed a ternary QDs/SEBS-g-MA/TFB composite to tackle a similar hole-injection issue. (2) Patterning method and resolution. MPA at the QD interface appears to enhance adhesion of intrinsically stretchable QD films to SiO₂ trenches—likely due to polarity matching—and, combined with localized strain during thermal-assisted transfer printing, enables high-definition patterning of stretchable QD layers. The earlier Nature Electronics (2024) work used conventional transfer printing, which may be less suited to pattern elastic QD/polymer composites at very high resolution. (3) Device performance. The present stretchable QLEDs achieve record-level luminance and current efficiency among stretchable electroluminescent devices based on both inorganic and organic emitters.

On this basis, I consider the manuscript a significant advance for intrinsically stretchable QLEDs with implications for next-generation displays conformable to non-developable surfaces. That said, several issues should be addressed to strengthen the claims:

Our response to summary comments: We sincerely appreciate the reviewer's thoughtful evaluation of our work and for clearly articulating the distinctions between the present study and our previous publication. We are grateful for the recognition that this work enables the concurrent realization of ultrahigh resolution patterning and record-level device performance in intrinsically stretchable QLEDs through LIFT process.

In response to the reviewer's comments, we have conducted additional analyses and revised the manuscript to address the key points raised, particularly regarding the proposed chemical interaction in the QD nanocomposite and the evidence for ligand engineering. The manuscript

has been updated accordingly to clarify these aspects and to better support the underlying mechanism.

Comment #1: *The manuscript posits chemical bonding between maleic anhydride/acid groups in SEBS-g-MA and free oleic acid via half-esterification. However, maleic anhydride is relatively stable. Its conversion to the diacid may not proceed readily in the solid state at room temperature. In Extended Data Fig. 3b, the QD/SEBS-g-MA film shows only a marginal increase near $\sim 1200\text{ cm}^{-1}$, assigned to additional C–O bonds.*

The authors need to be cautious about this. First, as depicted in Extended Data Fig. 3a, the total number of C–O bonds in the system would not change before or after reaction, if any. Second, the signal-to-noise ratio is low, and the apparent “bump” could reflect noise. Additionally, the authors used the same combination of QDs and SEBS-g-MA in their Nat. Electron. 2024 work without claiming covalent bonding at this interface.

This point does not detract from the demonstrated stretchability or device metrics, but it should be clarified. A simple control—mixing SEBS-g-MA with oleic acid in solution and monitoring by reliable spectroscopy—could test whether the reaction occurs under mild conditions. If it does not proceed in solution at room temperature, it is unlikely to occur efficiently in the solid state.

Our response to comment #1: We sincerely thank the reviewer for raising this important and constructive point regarding the possible chemical interaction between SEBS-g-MA and the QD surface ligands.

We agree that maleic anhydride is relatively stable under dry solid-state conditions. We would like to clarify, however, that the composite is formed under solution-phase processing conditions rather than in the solid state. Specifically, the QDs and SEBS-g-MA are mixed in toluene, stirred for 2 h, and subsequently subjected to sonication at 60 °C prior to film formation. Under these conditions, cyclic anhydrides undergo direct ring opening upon interaction with metal or oxide surfaces, forming surface-bound carboxylate species (*J. Phys. Chem. B*, **102**, 8807–8815 (1998)). This processing condition is distinct from that in our previous work (*Nat. Electron.* 2024), and the corresponding details have been clarified in the revised manuscript.

Considering the reviewer’s concern, we no longer rely on the weak feature near $\sim 1200\text{ cm}^{-1}$ as evidence of chemical bonding. Instead, we focus on the characteristic anhydride C=O stretching bands (~ 1780 and $\sim 1870\text{ cm}^{-1}$), which serve as a more reliable indicator of chemical transformation. These bands are clearly observed in pristine SEBS-g-MA and in the QD/SEBS-g-MA mixture prior to nanocomposite processing but are absent in pristine QDs and the QD/SEBS nanocomposite (**Extended Data Fig. 3b**). After thermal sonication treatment, these anhydride peaks are substantially diminished in the QD/SEBS-g-MA nanocomposite (**Extended Data Fig. 3c**), indicating consumption of the anhydride groups under the composite-forming conditions.

These results support the ring opening of the succinic anhydride groups during composite formation processing and their subsequent interaction with the QD surface ligands, which likely contributes to the enhanced structural integrity and solvent resistance of the nanocomposite. The manuscript has been revised accordingly, and the expanded FT-IR analysis has been included in **Extended Data Fig. 3b and 3c**.

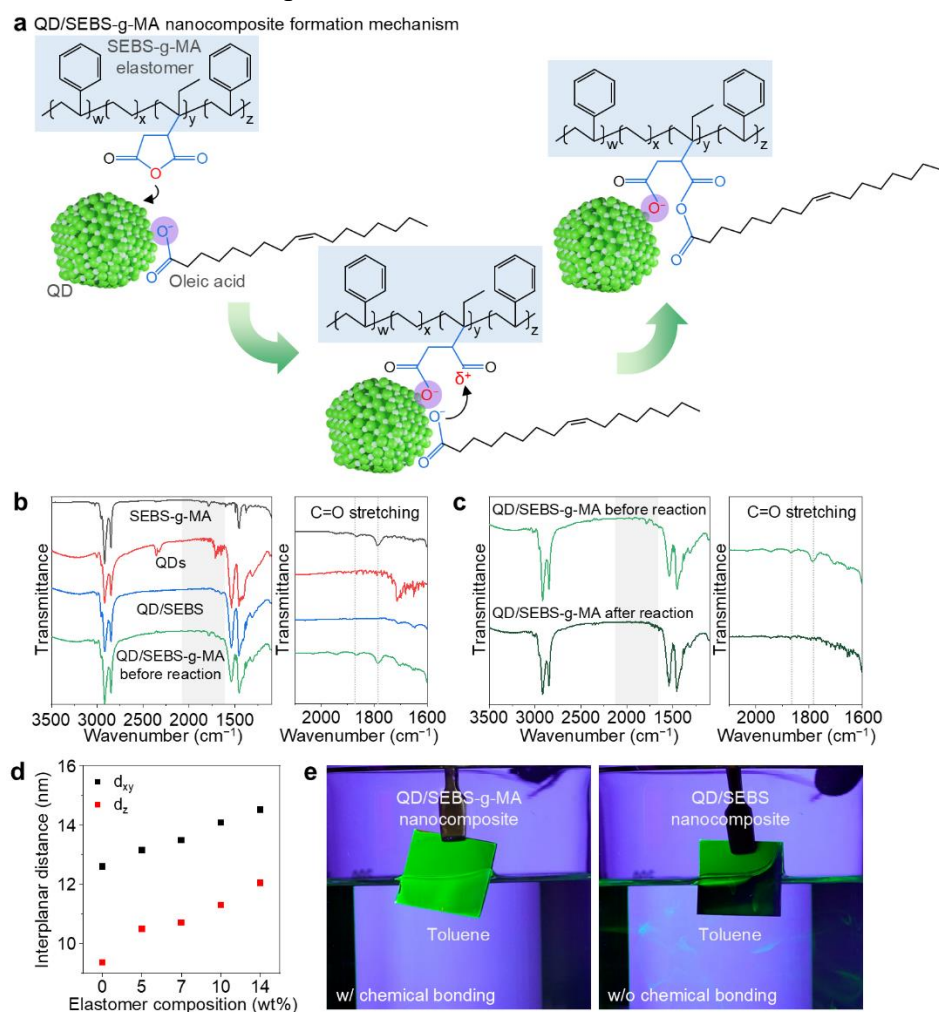
Our modification to the manuscript:

Page 6 in the revised manuscript:

“Colloidal QDs (**Extended Data Fig. 2**) are incorporated into polystyrene-block-poly(ethylene-ran-butylene)-block-polystyrene-graft-maleic anhydride (SEBS-g-MA) elastomer and spin-coated onto an octadecyltrichlorosilane (ODTS)-treated donor substrate to form composite films. In this structure, the succinic anhydride groups in the SEBS-g-MA undergo ring-opening during composite formation and become anchored to the QD surface (**Fig. 1f, Extended Data Fig. 3a, Methods**)^[51]. Fourier-transform infrared (FT-IR) spectroscopy shows that the characteristic anhydride C=O stretching bands (~ 1780 and ~ 1870 cm^{-1}) of the succinic anhydride graft, which are present in pristine SEBS-g-MA but absent in pristine QDs and QD/SEBS nanocomposite, are significantly diminished in QD/SEBS-g-MA after the composite formation (**Extended Data Fig. 3b,c**). Such chemical anchoring enhances structural integrity and solvent resistance, whereas physically blended QD/SEBS nanocomposites without MA grafting exhibit poor cohesion (**Extended Data Fig. 3d,e**)”

Page 18 in the revised manuscript:

“The mixture was stirred for 2 h, followed by sonication at 60 °C for 40 min condition.”



Extended Data Fig. 3 | Chemical bonding between QDs and SEBS-g-MA elastomer. a, Schematic illustration of the ring-opening of SEBS-g-MA during composite formation, indicating chemical bonding between QDs and the elastomer matrix. **b,c,** FT-IR spectra of SEBS-g-MA, QDs, QD/SEBS, and QD/SEBS-g-MA before reaction (**b**), and of the QD/SEBS-g-MA before and after reaction (**c**). The right panels in (**b**) and (**c**) show magnified views of the anhydride C=O stretching band (~ 1780 and ~ 1870 cm^{-1}). **d,** Interplanar distances of QDs in the horizontal (d_{xy}) and vertical (d_z) directions in QD/SEBS-g-MA nanocomposites, measured by GISAXS as a function of SEBS-g-MA composition. **e,** Solvent resistance test of QD nanocomposite films with (left) and without (right) chemical bonding after immersion in toluene.

Comment #2: Similarly, FTIR data in Fig. 2e provides limited support for ligand exchange, even though MPA exchange is plausible and well documented. Because SEBS-g-MA contains two C=O groups (or three if it reacts with oleic acid to form the proposed complex in Fig. 1f) whereas MPA has one, the net C=O trend depends on the relative coverage of each species.

The manuscript already includes QDs fully coated with MPA (Fig. S7). Using those and QDs coated solely with SEBS-g-MA as FTIR standards would strengthen the assignment and validate the exchange.

Our response to comment #2: We thank the reviewer for the insightful comments regarding the interpretation of the Fourier transform infrared spectroscopy (FT-IR) data for MPA-based ligand exchange.

In Fourier transform infrared spectroscopy (FT-IR), the newly emerged peak at $\sim 1700\text{ cm}^{-1}$ after MPA treatment is assigned to the C=O stretching vibration of protonated carboxylic acid groups in MPA, which is distinct from the carboxylate-related stretching bands (~ 1540 and $\sim 1400\text{ cm}^{-1}$) associated with native oleic acid ligands and the SEBS-g-MA matrix (**Extended Data Fig. 6b**). We have revised the manuscript to clarify that this FT-IR feature serves as supportive rather than standalone evidence for the ligand exchange process.

To provide more direct evidence for selective ligand engineering using MPA, we additionally performed solid-state ^{13}C nuclear magnetic resonance (NMR) and X-ray photoelectron spectroscopy (XPS) analyses. During the selective ligand engineering process, the thiol group of MPA strongly coordinates with the surface Zn atoms of the QDs, forming Zn–S–CH₂ bonds at the QD interface, which is a well-established binding motif for thiol-functional ligands on Zn-rich QD surfaces. Consistent with this mechanism, the solid-state NMR spectrum of the LIFT-composite shows a distinct resonance at ~ 21 ppm, corresponding to the Zn–S–CH₂ moiety of MPA, which is absent in the untreated composite (**Fig. 2e**). In addition, XPS measurements reveal a clear shift in the sulfur binding energy after LIFT treatment, consistent with the formation of Zn–SR bonding (**Extended Data Fig. 6a**). Together, these results provide direct evidence that MPA is selectively incorporated and chemically bound to the QD surface.

The corresponding explanation and data have been incorporated into **Fig. 2e and Extended Data Fig. 6a, b** as well as the revised manuscript.

Our modification to the manuscript:

Page 8 in the revised manuscript:

“To enhance hole injection into QDs while enabling high-definition pixelation, the elastomeric polymer ligands surrounding the surface QDs at the HTL/EML interface are selectively displaced by MPA during the LIFT process (**Fig. 2d, Supplementary Fig. 4**). This selective ligand engineering preserves the bulk mechanical compliance of the composite while improving charge transport across the HTL/QD interface. During this process, the thiol group of MPA strongly coordinates with surface Zn atoms of the QDs, forming robust Zn–S–CH₂ bonds at the QD interface. Solid-state ^{13}C nuclear magnetic resonance (NMR) spectroscopy of the LIFT-composite reveals a distinct resonance at ~ 21 ppm, corresponding to the Zn–S–CH₂ moiety of MPA, which is absent in the untreated composite (**Fig. 2e**). This assignment is further supported by X-ray photoelectron spectroscopy (XPS), which shows clear shifts in the sulfur binding energy after LIFT treatment (**Extended Data Fig. 6a**) consistent with thiol coordination to the QD surface. Additional evidence of interfacial chemical modification is

provided by FT-IR spectroscopy, where a newly emerged C=O stretching band at $\sim 1700\text{ cm}^{-1}$ is observed in the LIFT-composite (**Extended Data Fig. 6b**), which is attributed to protonated carboxylic acid groups of MPA introduced at the interface.”

Page 27 in the revised manuscript:

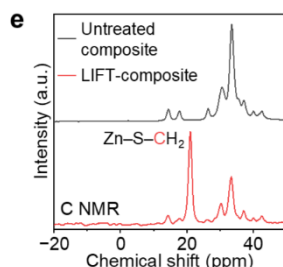
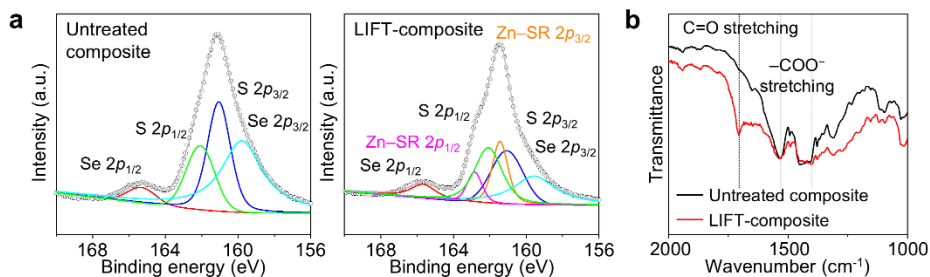


Fig. 2 | e–g, ^{13}C solid-state NMR spectra (e), AFM images (f), and TRPL decay profiles (g) of untreated- and LIFT-composites.

Page 35 in the revised manuscript:



Extended Data Fig. 6 | a, S 2p XPS spectra of untreated (left) and LIFT-composites (right). b, FT-IR spectra of untreated and LIFT-composites. b, FT-IR spectra of untreated and LIFT-composites.

Comment #3: Fig. 1j is compelling. It is helpful to provide the patterning resolution in ppi in the figure or caption.

Our response to comment #3: We thank the reviewer for this helpful comment. The full-colour stretchable pixel array shown in **Fig. 1j** consists of hexagonally aligned dot patterns with a dot diameter of $75\text{ }\mu\text{m}$ and an inter-dot spacing of $5\text{ }\mu\text{m}$, corresponding to a resolution of approximately 318 PPI.

The red, green, and blue subpixels are vertically stacked with controlled feature sizes to achieve the desired colour balance and image contrast (**Supplementary Fig. 2**). The thickness of each QD nanocomposite film was independently optimized to account for differences in photoluminescence quantum yield and UV colour-conversion efficiency among the red, green, and blue emitters. As suggested, the corresponding patterning resolution has now been explicitly added to **Fig. 1j** and its caption, together with the additional structural information in **Supplementary Fig. 2** and the revised manuscript.

Our modification to the manuscript:

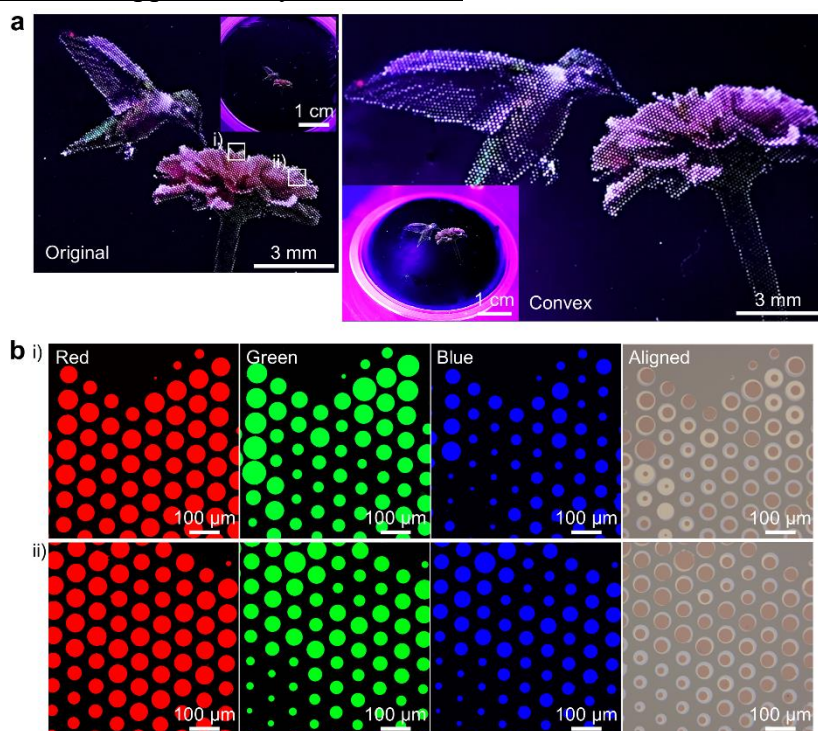
Page 26 in the revised manuscript:

“j, PL images of high-resolution full-colour QD pixel arrays (318 PPI) under planar and convex deformations.”

Page 13 in the revised manuscript:

“Beyond lateral patterning, the LIFT process also enables vertical stacking of QD nanocomposite films, allowing multilayered RGB architectures for full-colour emission (Fig. 1j, Supplementary Fig. 2).”

Page X in the revised Supplementary information:



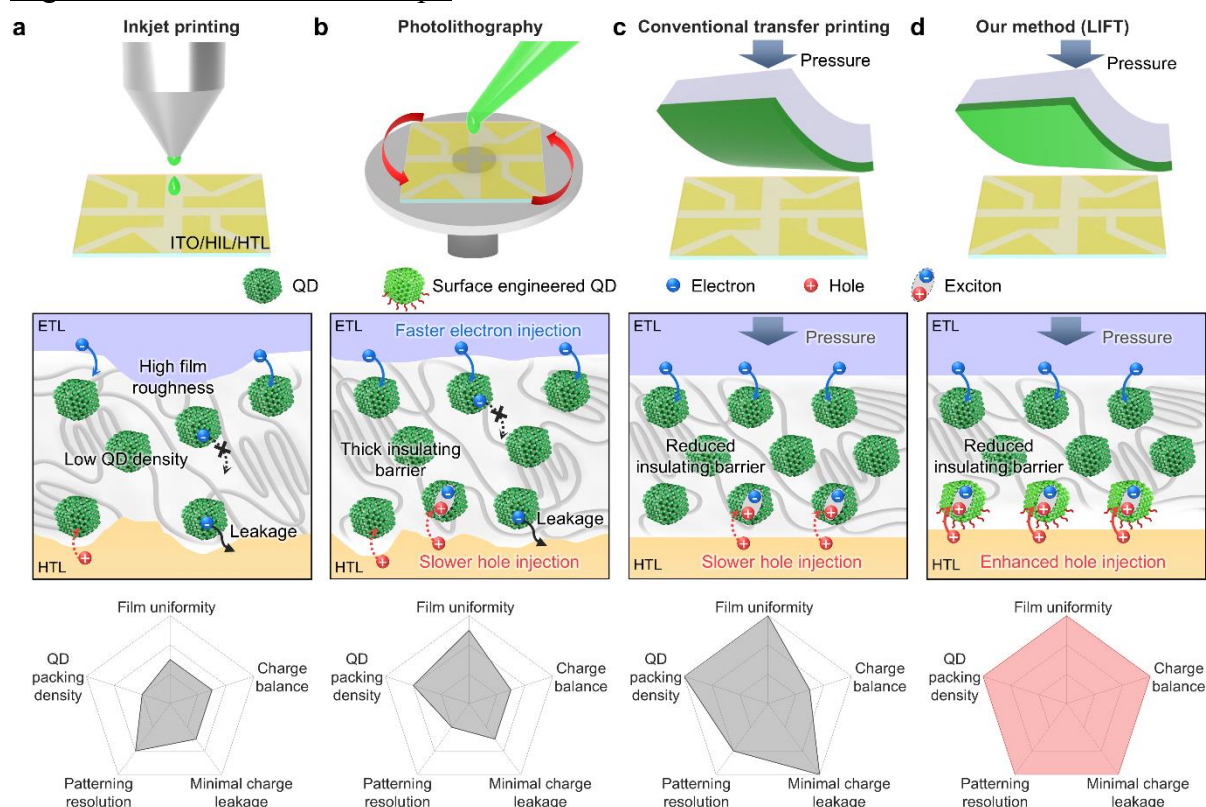
Supplementary Fig. 2 | a, PL images of high-definition full-colour QD pixel arrays (318 PPI) under planar (left) and convex (right) deformation. A stretchable SEBS/carbon black composite is used as the substrate. **b**, Magnified fluorescence images of the RGB subpixels from the boxed region in (a) and the corresponding optical microscope image of the vertically stacked RGB subpixels (right).

Comment #4: *Extended Data Fig. 1 compares the pros and cons of several manufacturing methods. I The pros/cons ranking for inkjet printing and photolithography merits reconsideration. For example, with bank structures or pixel-defining matrices, inkjet-printed devices can also mitigate charge leakage. Photolithography uses spin-coated QD films whose packing density should be comparable to that achieved in transfer-printed films.*

Our response to comment #4: We thank the reviewer for this constructive comment. We agree that the original comparison may have appeared overly generalized. Accordingly, we have revised the comparison of advantages and limitations in **Extended Data Fig. 1**, as suggested by the reviewer, to better reflect the current capabilities and design strategies of each manufacturing method.

Our modification to the manuscript:

Page 30 in the revised manuscript:



Extended Data Fig. 1 | Comparative evaluation of QD nanocomposite film fabrication strategies. a–d, Schematic illustrations of four representative QD pattern formation techniques (top), corresponding film cross-section (middle), and radar plots of key QLED performance metrics (bottom): inkjet printing (a), spin-coating during photolithography (b), conventional transfer printing (c), and LIFT (d).

Thank you very much again for your insightful comments. We feel that these comments have helped to improve the quality of the manuscript significantly.