

This Supplementary Information includes:

Supplementary Notes 1-11

Supplementary Figures 1-16

Supplementary Tables 1-6

Supplementary References 1-16

Supplementary Note 1. Room temperature deformation mechanism in perovskites

Perovskites are typically ionic crystals. Their plastic deformation at room temperature is universally dominated by dislocation slip on $\{011\}\langle 0\bar{1}1\rangle$ system. Although not the shortest lattice translation, this system is the only one that prevents close approach of like-charged cations during glide, thus introducing minimum Columb repulsion¹. This has been verified by diffraction contrast by weak-beam technique² and atomic-resolution STEM-HAADF observations³⁻⁷. The slip system is not only the most energetically favorable among common low-index slip systems, as evidenced by the generalized stacking fault energy profiles in our previous work⁷, but also exhibits the highest Schmid factor under the loading geometry along $[00\bar{1}]$ orientation. There are four equivalent variants of the $\{011\}\langle 0\bar{1}1\rangle$ system, as shown in Table S1, all of the four variants achieve the maximum Schmid factor of 0.5.

Table S1 The Schmid factors of slip systems when loaded along $[00\bar{1}]$ direction.

Loading direction	Slip system	Schmid factor
$[00\bar{1}]$	$(0\bar{1}\bar{1})[01\bar{1}]$ (denoted as “S ₁ ”) $(01\bar{1})[0\bar{1}\bar{1}]$ (denoted as “S ₂ ”) $(10\bar{1})[\bar{1}0\bar{1}]$ (denoted as “S ₃ ”) $(\bar{1}0\bar{1})[10\bar{1}]$ (denoted as “S ₄ ”)	0.5
	$\{01\bar{1}\}\langle 1\bar{1}\bar{1}\rangle$	0.41
	$\{1\bar{1}\bar{1}\}\langle 011\rangle$	0.41
	$\{001\}\langle 010\rangle$	0
	$\{001\}\langle 110\rangle$	0

Fig. S1 presents the stress-strain curve obtained from the in situ uniaxial compression of the CsPbBr₃ micropillar revealed in Figs. 1a-f. CsPbBr₃ exhibits exceptional metal-like plasticity through continuous slip activities on multiple equivalent $\{011\}\langle 0\bar{1}1\rangle$ systems, characterized by serrated flow in the stress-strain curve. The massive, pronounced stress drops indicate most of dislocations can glide out of the micron-scale sample, forming dislocation-starved state as proposed by Julia Greer et al.⁸, which is characteristic in micropillars. Such state help to minimize non-coplanar dislocation interactions, preventing strain hardening and crack formation, thereby contributing to the material’s exceptional plasticity.

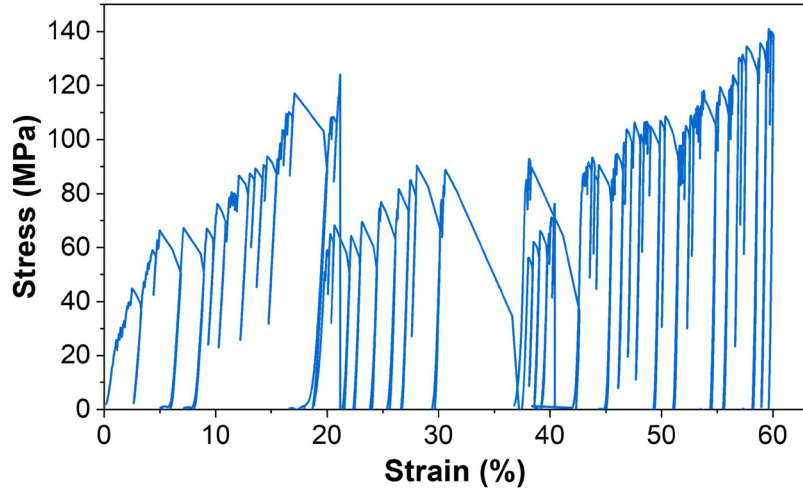


Fig. S1 Stress-strain curve of the CsPbBr₃ sample in Figs. 1a-e under uniaxial compression. Source data are provided as a Source Data file.

Supplementary Note 2. Glide-dissociated dislocations in CsPbBr₃ by simulation

A global charge-neutral model of CsPbBr₃ was constructed (Fig. S2a), incorporating an anion-rich Br-Br-cored dislocation and a cation-rich Cs-Pb-cored dislocation, with the enlarged views are shown in Figs. S2b and S2d. After energy minimization, both dislocations spontaneously dissociated into two collinear partial dislocations, adopting glide-dissociated configurations (Figs. S2c and 2e), confirming intrinsic glide-dissociation in CsPbBr₃.

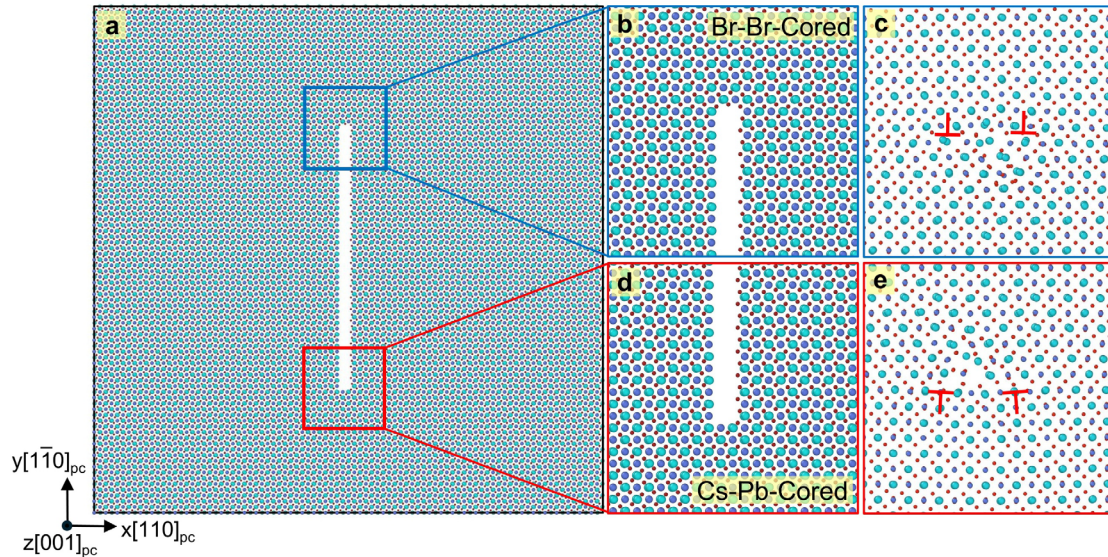


Fig. S2 Theoretical glide-dissociated dislocations in CsPbBr₃. **a**, Initial charge-neutral model. **b-c**, Enlarged view of the Br-Br-cored dislocation and its relaxed structure. **d-e**, Enlarged view of the Cs-Pb-cored dislocation and its relaxed structure. The blue, purple and red spheres are Cs, Pb and Br atoms, respectively.

Supplementary Note 3. Geometric phase analysis to identify dislocation cores

Geometric phase analysis (GPA) was employed to identify dislocation cores by mapping lattice strain fields with high precision. To visualize a $\{0\bar{1}1\}\langle 011\rangle$ dislocation, the HAADF image of CsPbBr₃ (Fig. S3a) with $[001]$ direction horizontal was rotated by 45° to align the $[011]$ axis horizontally. The (011) and $(0\bar{1}1)$ diffraction spots in the corresponding fast Fourier transform (FFT, Fig. S3b) were selected as geometric references to generate displacement fields of both planes. The resulting ϵ_{xx} strain map (Fig. S3c) displays dipole-like strain distributions, clearly locating the dislocation cores. The dashed white square in these figures highlights the region shown in Fig. 1h.

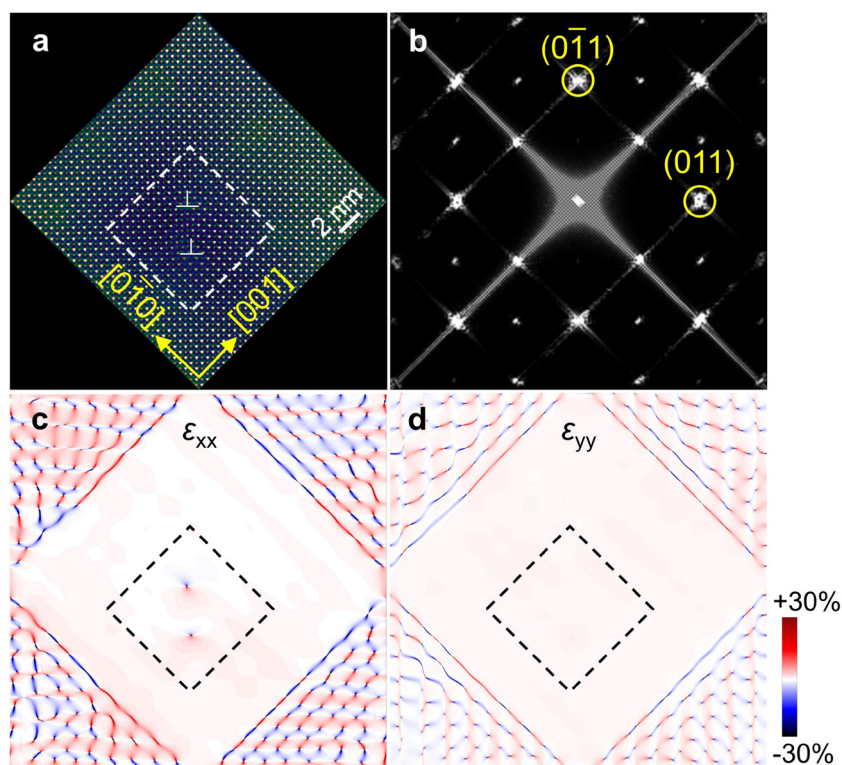


Fig. S3 GPA methodology for dislocation core identification. **a**, HAADF image of CsPbBr₃ rotated by 45°. **b**, Corresponding FFT with (011) and $(0\bar{1}1)$ diffraction spots selected as geometric references. **c-d**, ϵ_{xx} and ϵ_{yy} strain maps.

Supplementary Note 4. Atomic displacement cross section

Knock-on damage occurs when high-energy electron beam irradiation displaces atoms from their lattice sites to interstitial positions, creating vacancy and interstitial point defects⁹. This process requires the incident beam energy E_0 to exceed the threshold energy E_t , allowing the maximum transferable energy $E_{\max}$ from electron

beam to samples to exceed the displacement energy E_d of the target atom. The threshold energy E_t and maximum transferable energy $E_{\max}$ are calculated as follows¹⁰:

$$E_t = (511 \text{ KeV}) \{ [1 + AE_d / (561 \text{ eV})]^{1/2} - 1 \} \quad (1)$$

$$E_{\max} = 2E_0 \cdot (E_0 + 2m_0c^2) / (Mc^2) \quad (2)$$

where A is the atomic weight, m_0 is the electron mass, M is the mass of the target nucleus, and c is the speed of light.

The atomic displacement cross section σ quantifies the probability of such events¹⁰. For elements with atomic number $Z > 28$, the Rutherford formula is applied:

$$\sigma = (0.25 \text{ barn}) \cdot FZ^2(X-1) \quad (3)$$

For lighter elements ($Z < 28$), the Mackinley-Feshbach-Mott formula is used:

$$\sigma = (0.25 \text{ barn}) \cdot F \{ X + 2\pi\alpha\beta X^{1/2} - [1 + 2\pi\alpha\beta + (\beta^2 + \pi\alpha\beta) \cdot \ln(X)] \} \quad (4)$$

where $1 \text{ barn} = 10^{-28} \text{ m}^2$, $\alpha = Z/137$, $\beta = v/c$, v is the velocity of the incident electron, $F = (1 - \beta^2) / \beta^4$ is a relativistic factor, and $X = E_{\max} / E_d$.

Table S2 Atomic displacement cross sections for CsPbBr₃ and SrTiO₃ under a 300 KeV electron beam.

	E_d (eV)	$E_{\max}$ (eV)	E_t (KeV)	σ (barn)
Cs	25	6.4	834	-613
Pb	25	4.1	1124	-1530
Br	8.6	10.7	251	79
Sr	25	9.7	621	-240
Ti	15	17.8	261	11
O	13.6	53.2	91	38

As presented in Table S2, for a 300 keV incident electron beam in CsPbBr₃, the cross sections for Cs and Pb are negative, indicating negligible atomic displacement. Whereas the σ for Br is 79 barns, suggesting significant atomic displacement and the formation of Br vacancies and interstitials. In SrTiO₃, both Ti and O atoms exhibit positive cross sections, with Ti showing a smaller value. Consequently, O is identified as the primary source of point defects¹¹.

Supplementary Note 5. Dynamics of point defect concentration

This note reveals the calculated analytical solution of dynamics of point defect concentrations in CsPbBr₃ and SrTiO₃ perovskites under electron beam irradiation based on the rate equations¹², focusing on defect generation and recombination processes under both pulsed and continuous electron beam conditions. The results are summarized in Tables S3-5 and correlated with dislocation behaviours observed in Figs. 2 and 3, providing insights into the materials' responses to irradiation-induced damage.

i. Point defect generation rate

As described in Supplementary Note 4, a 300 KeV electron beam induces atomic displacements in CsPbBr₃ and SrTiO₃, primarily displacing Br and O atoms, respectively, to form vacancies and interstitials. The point defect generation rate K_0 (in unit of displacements per atom per second, dpa/s) quantifies the electron-induced atomic displacement damage and is calculated follows¹²:

$$K_0 = \sigma(\gamma E_0/4E_d)\Phi \quad (5)$$

where σ is the atomic displacement cross section, $\gamma = 4m_0M/(m_0+M)^2$ is the mass factor, Φ is the electron dose rate.

ii. Defect concentration dynamics

The concentrations of vacancies (C_v) and interstitials (C_i) in CsPbBr₃ and SrTiO₃ arise from a dynamic competition between radiation-induced defect generation and vacancy-interstitial recombination. The evolution of defect concentration can be mainly described in three distinct stages¹²:

(1) Build-up stage: At the onset of electron beam irradiation, defects generation dominates, leading to a linear increase in defect concentration with time (t):

$$dC/dt = K_0 \quad (6)$$

$$C = C_i = C_v = K_0 t \quad (7)$$

(2) Quasi-steady-state stage: As irradiation continues and defect density increases, vacancy-interstitial recombination becomes significant and counter compensates defect generation, leading to a saturation concentration C_0 at a critical time t_0 :

$$dC/dt = K_0 - K_{iv}C^2 = 0 \quad (8)$$

$$C_0 = (K_0/K_{iv})^{1/2}, \text{ and } t_0 = (K_0K_{iv})^{-1/2} \quad (9)$$

here, K_{iv} is the interstitial-vacancy recombination rate coefficient, defined as:

$$K_{iv} = 4\pi r_{iv} D_i / \Omega \quad (10)$$

where, r_{iv} is the recombination radius, approximated as the lattice constant a of the perovskite, Ω is atomic volume (unit cell volume divided by the number of atoms), D_i is the interstitial diffusion coefficient, given by:

$$D_i = \alpha a^2 \nu \exp(-E_{mi}/kT) \quad (11)$$

with $\alpha = 0.5$ (geometric factor for interstitials in perovskite crystals), $\nu = 10^{13} \text{ s}^{-1}$ is the Debye frequency, E_{mi} is the migration energy of interstitials (0.21 eV for Br atom¹³ and 0.3 eV for O atom¹⁴), $k = 8.6173 \cdot 10^{-5} \text{ eV/K}$ is the Boltzmann constant, T is the temperature. The parameters related to the diffusion process are listed in Table S3.

Table S3 Defect diffusion parameters.

	E_{mi} (eV)	D_i (nm ² /s)	K_{iv} (s ⁻¹)
Br	0.21	$4.9 \cdot 10^8$	$8.9 \cdot 10^{10}$
O	0.3	$1.5 \cdot 10^7$	$6 \cdot 10^9$

(3) Recombination stage: when electron beam is turned off or moved away from the pixel at time of t_1 , defect generation ceases, interstitial-vacancy recombination dominates. The defect concentration decays as:

$$dC/dt = -K_{iv} C_i C_v \quad (12)$$

$$C = C_i = C_v = C_0 / [1 + C_0 \cdot K_{iv} \cdot (t - t_1)] \quad (13)$$

iii. Point defect concentration dynamics of CsPbBr₃: pulsed vs. continuous beam

The beam parameters and point defect generation parameters for the datasets presented in Fig. 2 are summarized in Table S4.

Table S4 Point defect concentration dynamics in CsPbBr₃ in Fig. 2.

Parameter	Pulsed beam	Continuous beam
Φ_{on} (e ⁻ ·nm ⁻² ·s ⁻¹)	$1.14 \cdot 10^{11}$	$1.14 \cdot 10^{11}$
K_0 (s ⁻¹)	217	217
C_0	$4.95 \cdot 10^{-5}$	$4.95 \cdot 10^{-5}$
t_0 (μs)	0.23	0.23
Δt_s (μs)	0.77	3.77

iv. Point defect concentration dynamics in SrTiO₃: pulsed and continuous beam

Table S5 presents the key parameters of point defect concentration dynamics in SrTiO₃ using pulsed beam condition described in Fig. 3f, and continuous beam condition in Fig. 3i.

Table S5 Point defect concentrations dynamics in SrTiO₃.

Parameter	Pulsed beam	Continuous beam
Φ_{on} ($\text{e}^- \cdot \text{nm}^{-2} \cdot \text{s}^{-1}$)	$6.31 \cdot 10^{10}$	$1.33 \cdot 10^{11}$
K_0 (s^{-1})	181	382
t_0 (μs)	0.95	0.66
C_0	$1.45 \cdot 10^{-4}$ (at $t = 0.8 \mu\text{s}$)	$2.52 \cdot 10^{-4}$

Supplementary Note 6. Decoupling dose, dose rate, and pulsing effects on climb

To deconvolve the effects of total electron dose, dose rate, and beam pulsing on dislocation motion, we conducted comparable experiments under varying beam conditions while imaging different dislocations. Key beam parameters and the corresponding defect dynamics are summarized in Table S6, including beam type, beam-on dose rate Φ_{on} , beam-on duration per pixel dwell, and total accumulated dose per pixel over all acquired frames. Here, k_0 is the defect generation rate, t_0 is the time to reach quasi-steady state (peak defect concentration C_0 ; see Supplementary Note 5), and Δt_s is the sustained duration of peak concentration C_0 within each pixel dwell time.

Table S6 Calculated point defect generation in CsPbBr₃ and SrTiO₃ under various beam conditions.

Dataset	Beam	Φ_{on} ($\text{e}^- \cdot \text{nm}^{-2} \cdot \text{s}^{-1}$)	On time (μs)	Total dose ($\text{e}^- \cdot \text{nm}^{-2}$)	K_0 (s^{-1})	t_0 (μs)	Δt_s (μs)	C_0 (10^{-5})	Type
Fig. S4a-b	Pulsed	$4.43 \cdot 10^{10}$	1	$9.32 \cdot 10^5$	84	0.37	0.63	3.08	glide
Fig. 2c-d	Pulsed	$1.14 \cdot 10^{11}$	1	$2.98 \cdot 10^6$	216	0.23	0.77	4.95	glide
Fig. 3f (SrTiO ₃)	Pulsed	$6.31 \cdot 10^{10}$	0.8	$5.05 \cdot 10^4$	181	0.95	N/A	14.52 (at 0.8 μs)	climb
Fig. S5a	Continuous	$2.86 \cdot 10^{10}$	4	$1.14 \cdot 10^5$	54	0.46	3.54	2.47	climb
Fig. S5b	Continuous	$4.43 \cdot 10^{10}$	4	$1.77 \cdot 10^5$	84	0.37	3.63	3.08	climb
Fig. 2e-f	Continuous	$1.14 \cdot 10^{11}$	4	$4.56 \cdot 10^5$	216	0.23	3.77	4.95	climb
Fig. S4c	Continuous	$4.53 \cdot 10^{11}$	4	$1.81 \cdot 10^6$	860	0.11	3.89	9.85	climb

Fig. S4a reveals a glide-dissociated dislocation under a pulsed beam (4 μs period, 25% duty cycle, $\Phi_{\text{on}} = 4.43 \times 10^{10} \text{ e}^- \cdot \text{nm}^{-2} \cdot \text{s}^{-1}$). The dislocation remained stable after 21 frames (Fig. S4b), accumulating a total dose of $\sim 9.32 \times 10^5 \text{ e}^- \cdot \text{nm}^{-2}$ per pixel. In contrast, under a continuous beam (Fig. S4c) with a $\sim 10\times$ higher dose rate ($\sim 4.53 \times 10^{11} \text{ e}^- \cdot \text{nm}^{-2} \cdot \text{s}^{-1}$) than Φ_{on} , the dislocation climbed immediately, exhibiting much faster climb velocity than observed in Figs. 2e-f. The corresponding calculated point-defect concentration dynamics under pulsed and continuous beams are compared in Figs. S4d-e. Besides, imaging dislocations with continuous beam directly cause climb even when the total accumulated dose per pixel was far lower than under pulsed conditions (Fig. S5). These observations rule out total dose as the primary driver of climb.

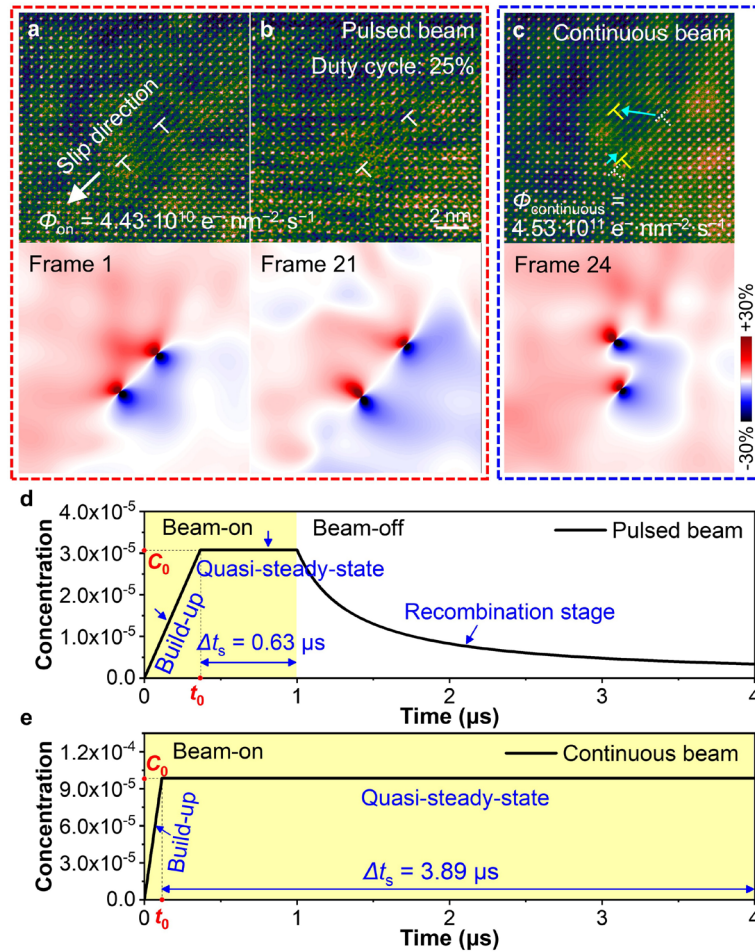


Fig. S4. Dislocation structures in CsPbBr_3 under pulsed beam and continuous beam. a,

HAADF-STEM image of an intrinsic, glide-dissociated dislocation using an EDM-based pulsed electron beam with 25% duty cycle. **b,** The 21st frame of the same dislocation under the pulsed beam. **c,** Glide-to-climb transition of the same dislocation induced by a continuous electron beam. **d-e,** Calculated dynamics of Br point defect concentration induced by electron beam irradiation under pulsed and continuous electron beam conditions within a 4 μs dwell time, respectively. Source data are provided as a Source Data file.

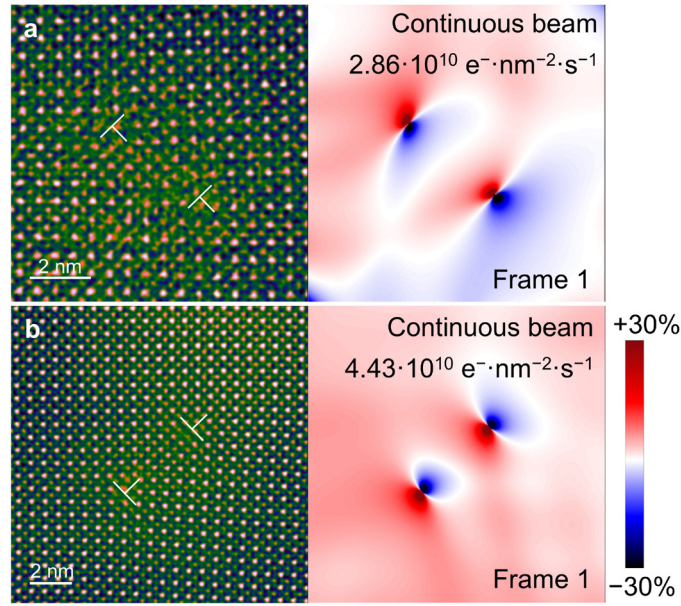


Fig. S5. Climb-dissociated dislocations in CsPbBr₃ under continuous beam irradiation.

a, Dislocation structure observed at a dose rate of $2.86 \cdot 10^{10} \text{ e}^- \cdot \text{nm}^{-2} \cdot \text{s}^{-1}$, showing immediate climb in the first frame. **b**, Dislocation structure observed at a dose rate of $4.43 \cdot 10^{10} \text{ e}^- \cdot \text{nm}^{-2} \cdot \text{s}^{-1}$, with climb occurring in the first frame.

Moreover, from the results of SrTiO₃ (Fig. 3f), the dislocation exhibits climb even under a pulsed beam with a 20% duty cycle and the lowest dose rate that still permits lattice resolution. This observation indicates that a material-specific critical dose rate threshold exists: below this threshold, dislocations can possibly be imaged in their glide state regardless of total accumulated dose; above it, climb is initiated even under pulsed illumination.

Finally, climb requires sufficient time for generated defects to diffuse to the dislocation cores, reach supersaturation, and drive non-conservative motion. Thus, the beam-on time per pixel, or the sustained duration of peak concentration (Δt_s), emerges as a critical parameter beyond dose rate alone. As shown in Table S6 and Fig. S6, though the pulsed beam condition in Fig. S6a (Fig. 2d) has a much higher Φ_{on} , K_0 and C_0 than the continuous beam Fig. S6b (Fig. S5a), it nevertheless preserves the intrinsic glide-dissociated structure, owing to its beam-off intervals that significantly shorten Δt_s to 0.77 μs and allow rapid defect recombination. Under the continuous beam, however, the peak concentration C_0 persists for $\Delta t_s = 3.54 \mu\text{s}$, providing sufficient time for defects to migrate to the dislocation cores and initiate climb.

Overall, these comparisons demonstrate that pulsing nature, which provides beam-off intervals for defect annihilation, is the primary factor suppressing climb, rather than

simply dose rate or total dose.

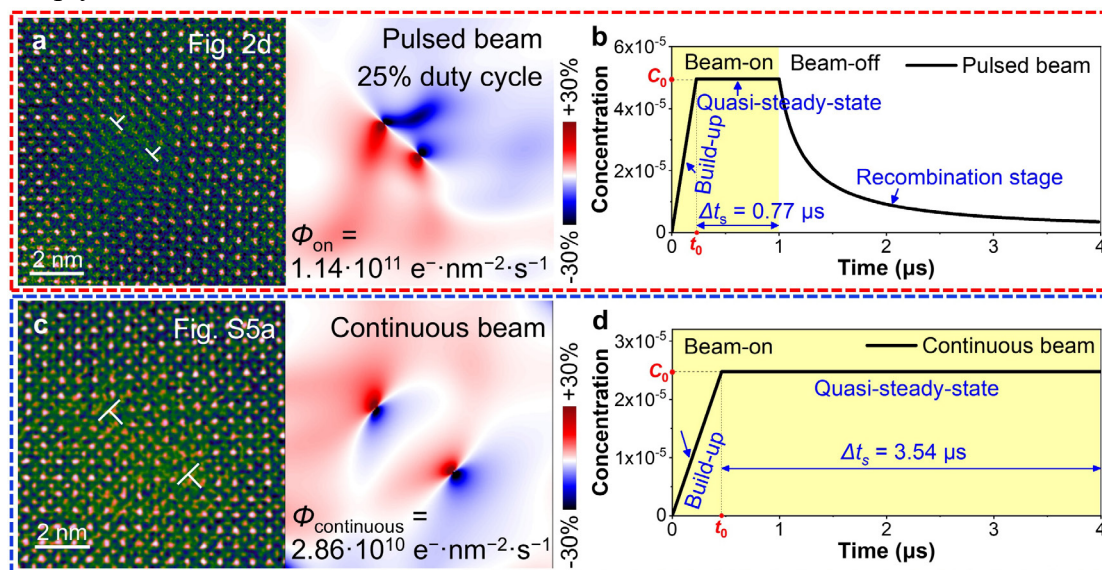


Fig. S6. Dislocation structures and point defect dynamics under pulsed vs. continuous beam. **a-b**, The dislocation structure (Fig. 2d) and calculated defect concentration evolution within one 4 μs dwell time under pulsed beam conditions, respectively. **c-d**, The dislocation structure (Fig. S5a) and calculated defect concentration evolution within one 4 μs dwell time under continuous beam conditions, respectively. Source data are provided as a Source Data file.

Supplementary Note 7. Atomic structure and stress-strain curve of SrTiO_3

HAADF image of SrTiO_3 (Fig. S7a) resolves cation columns (Sr and Ti) but not oxygen columns due to their weak scattering contrast. Complementary annular bright-field (ABF) STEM image (Fig. S7b) can visualize Sr, Ti and O columns, enabling precise characterization of the perovskite lattice structure. An inverted-contrast ABF image (Fig. S7c) is provided for enhanced visibility of the atomic arrangement.

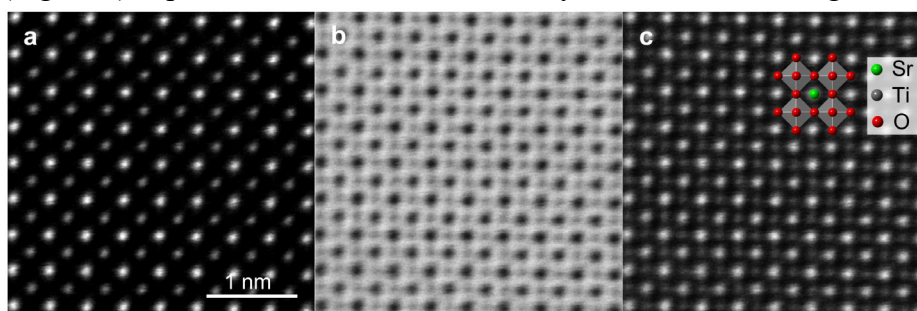


Fig. S7 Lattice structure of SrTiO_3 . **a**, STEM-HAADF image showing cation columns. **b**, STEM-ABF image revealing both cation and oxygen columns. **c**, Inverted-contrast STEM-ABF image for improved visualization.

Mechanically, SrTiO₃ exhibits inferior plasticity compared to CsPbBr₃ (Fig. S8), characterized by extremely high strength and limited plasticity. Although the activation of a high density of dislocations under compression, their glide is hindered by the material's high lattice stiffness. This rigidity leads to dislocation accumulation within the sample, limiting the plastic deformation.

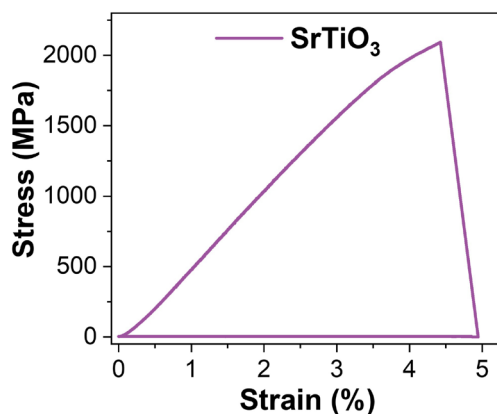


Fig. S8 The stress-strain curve of SrTiO₃ under uniaxial compression. Source data are provided as a Source Data file.

Supplementary Note 8. Molecular statics (MS) and molecular dynamics (MD) simulations for dislocation configurations in SrTiO₃

i. Generalized stacking fault energy (GSFE) profiles of SrTiO₃

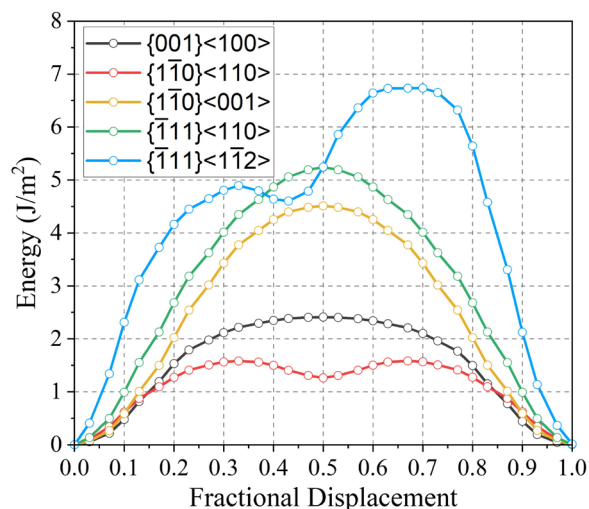


Fig. S9 GSFE profiles of low-index slip systems in SrTiO₃. Source data are provided as a Source Data file.

GSFE profiles are generally utilized to evaluate the slip energy barriers associated with various slip systems in crystalline materials. Our previous study demonstrated $\{1\bar{1}0\}\langle 110\rangle$ slip system is energetically favourable in CsPbBr₃ perovskite⁷ with the

lowest energy barrier. Here, the GSFE profiles of SrTiO₃ were calculated using MS to clarify the observed deformation mechanisms. Five slip systems were considered: $\{001\}\langle 100\rangle$, $\{1\bar{1}0\}\langle 110\rangle$, $\{1\bar{1}0\}\langle 001\rangle$, $\{\bar{1}11\}\langle 110\rangle$, and $\{\bar{1}11\}\langle 1\bar{1}2\rangle$, as shown in Fig. S9. Among these, the $\{1\bar{1}0\}\langle 110\rangle$ slip system exhibits the lowest energy barrier, confirming its energetic favourability. An energy minimum at a fractional displacement of $0.5b$ indicates that a full dislocation dissociates into two partial dislocations, aligning with experimental observations.

ii. Glide-dissociated dislocations in SrTiO₃ by simulation

A global charge-neutral model of SrTiO₃ was constructed, incorporating a negatively charged O-O-cored dislocation and a positively charged Sr-Ti-cored dislocation (Fig. S10a). Enlarged views of these two dislocation cores are shown in Figs. S10b and S10d. After energy minimization, both dislocations spontaneously dissociated into two collinear partial dislocations, adopting glide-dissociated configurations (Figs. S10c and 10e). These findings confirm theoretical glide-dissociation in SrTiO₃ perovskites.

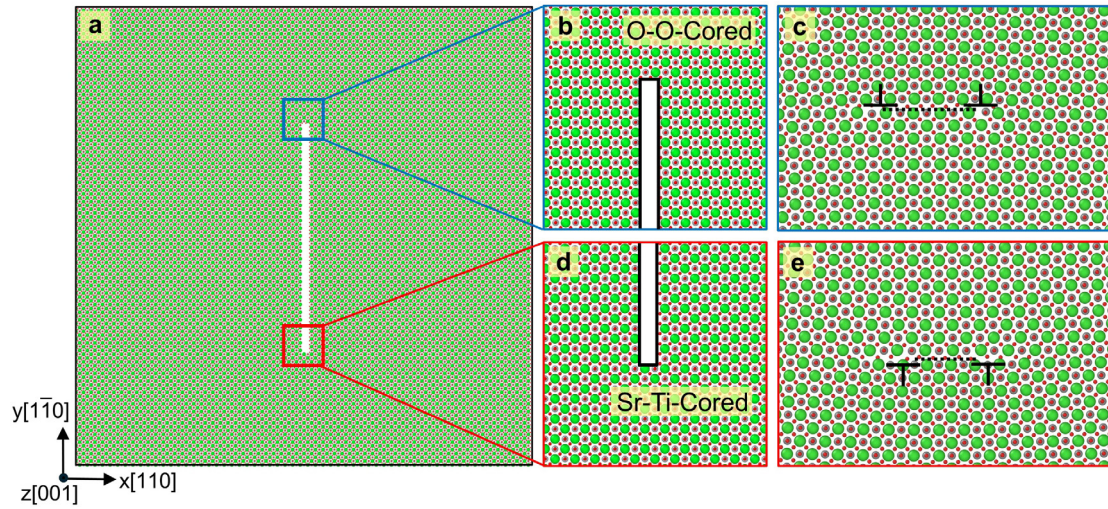


Fig. S10 Simulated glide-dissociated dislocations in SrTiO₃. **a**, Initial charge-neutral model. **b-c**, Magnified view of the O-O-cored dislocation and its relaxed structure, respectively. **d-e**, Magnified view of the Sr-Ti-cored dislocation and its relaxed structure, respectively. The green, grey and red spheres are Sr, Ti, O atoms, respectively.

iii. Convergence test for supercell size optimization

To minimize interactions between dislocations at the top and bottom surfaces of the simulation cell, a convergence test was conducted. As shown in Fig. S11a, the energy per atom for dislocation models decreases with increasing cell size and stabilizes around $N=40$. The average length of four incorporated $1/2a\langle 110\rangle$ partial dislocations

increase with supercell size until $N=50$ (Fig. S11b). Notably, the dislocation dissociation width w remains unaffected across all tested supercell sizes (Fig. S11c). Based on these results, a $60 \times 60 \times 12$ supercell was selected to eliminate boundary interactions while maintaining computational efficiency.

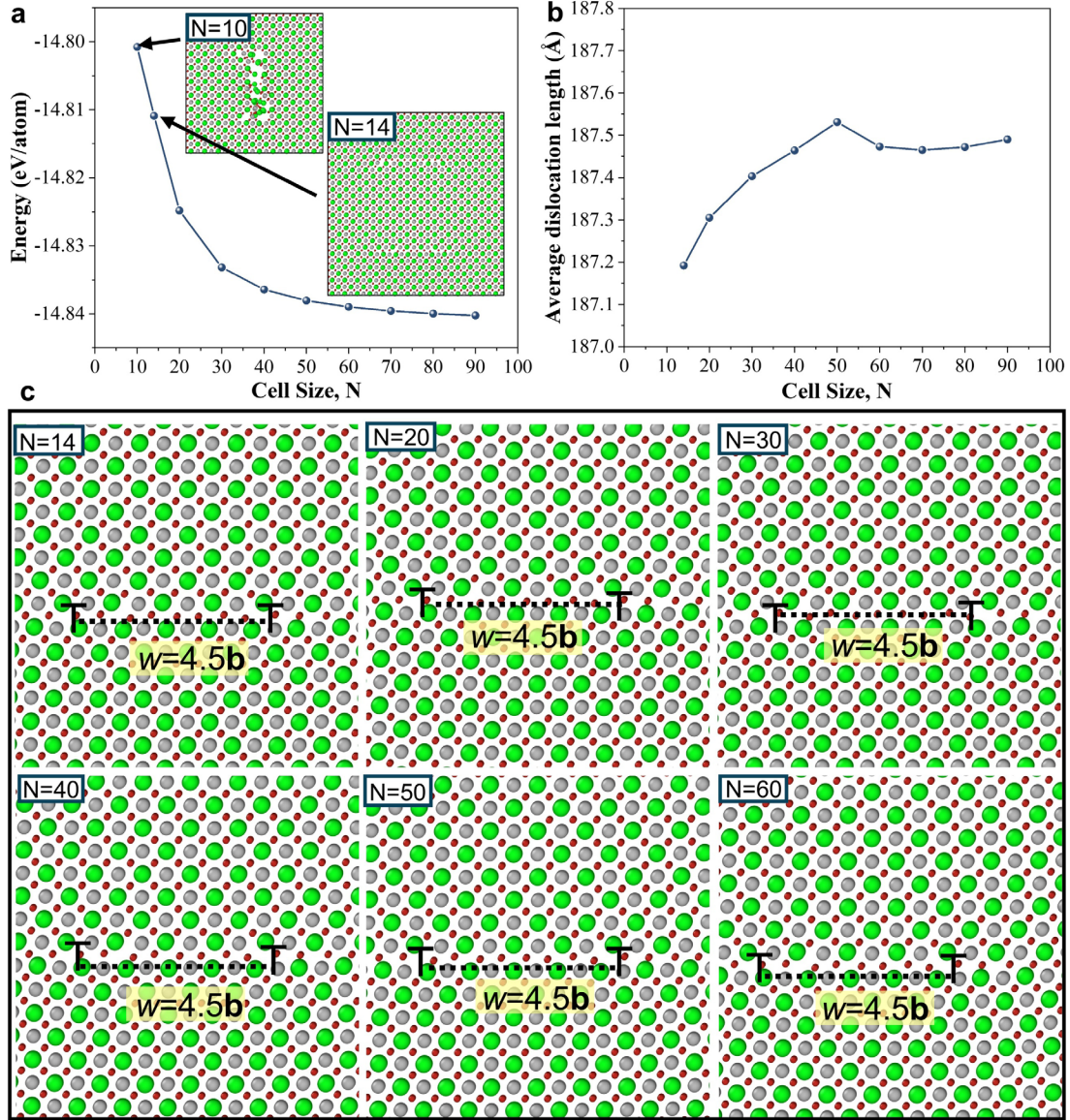


Fig. S11 Convergence test for supercell size optimization. **a**, Energy per atom as a function of supercell size. **b**, Average length of four $1/2a\langle 110 \rangle$ partial dislocations versus supercell size. **c**, Dislocation dissociation width w across tested supercell size. Source data are provided as a Source Data file.

iv. Shear deformation of glide-dissociated dislocations

To investigate the structural evolution of glide-dissociated dislocations under mechanical deformation, shear deformation simulations were performed. As illustrated in Fig. S12, both O-O-cored and Sr-Ti-cored dislocations exhibit stable glide-

dissociated configurations throughout the plastic deformation process. These results indicate that the glide-dissociated structures are robust under shear, with no significant reconfiguration or transition to other dissociation modes. Consequently, these findings rule out shear deformation as the cause of the observed climb-dissociated configuration, suggesting alternative mechanisms.

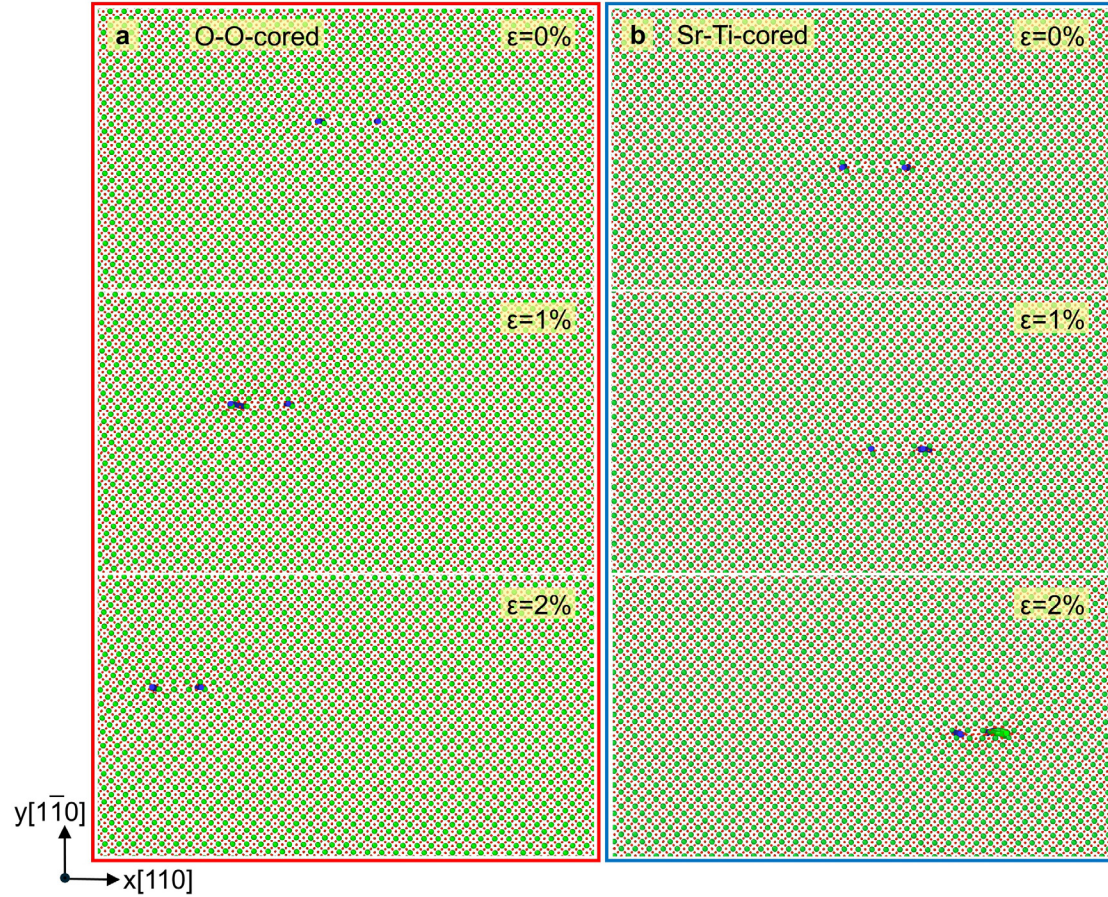


Fig. S12 Shear deformation of glide-dissociated dislocations. **a**, Structural evolution of O-O-cored dislocation under shear deformation. **b**, Structural evolution of Sr-Ti-cored dislocations under shear deformation. The blue/green regions highlight partial dislocation lines. The slight curvature of dislocation line produces a finite width rather than a single point when viewed edge-on. Green indicates regions with higher local atomic distortion.

v. Energy preference of glide- and climb-dissociated dislocations

To evaluate the energy preference of glide- and climb-dissociated configurations, a global charge-neutral model was constructed (Fig. S13a), incorporating both O-O-cored (Figs. S13b-c) and Sr-Ti-cored (Figs. S13e-f) dislocations with varying climb heights, h . The system energy of these models was compared to that of glide-dissociated configurations after energy minimization. The energy landscape, shown in Fig. S13d, reveals that the system energy decreases with increasing climb height, reaching a

minimum at $h=3b$. This result confirms that the climb-dissociated configuration is thermodynamically more stable than the glide-dissociated configuration. The transition from glide to climb dissociation is expected to occur once given sufficient vacancy flux to facilitate climb.

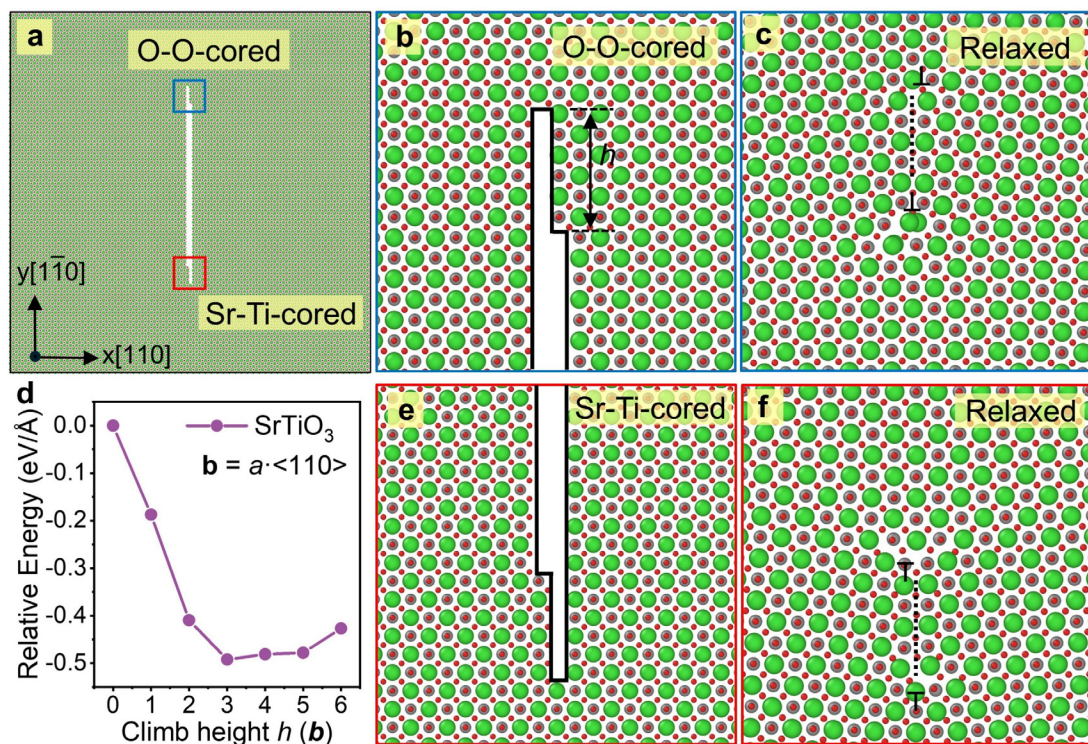


Fig. S13 Energy profiles of glide- and climb-dissociated dislocation. **a**, A global charge-neutral model incorporating O-O and Sr-Ti-cored dislocations. **b-c**, Enlarged view and relaxed configuration of the O-O-cored dislocation. **d**, Energy landscape as a function of climb height h . **e-f**, Enlarged view and relaxed configuration of the Sr-Ti-cored dislocation.

Source data are provided as a Source Data file.

Supplementary Note 9. Charge density analyses of climb-dissociated dislocations with distinct core-chemistry in SrTiO₃

To uncover the physical mechanism underlying the distinct climb-dissociated configurations observed for anion-terminated (compact) and cation-terminated (extended) dislocations, we performed charge density analyses on dislocation models in SrTiO₃ with method reported by Hirel et al¹⁵. We projected the nominal ionic charges of all atoms onto the (001) plane and convoluted them with 2-D Gaussian kernels (standard deviation $s = 3$ Å) to obtain a smooth charge-density map $\Delta\rho(x,y)$:

$$\Delta\rho(x,y) = \sum_{i=1}^N \frac{q_i}{s\sqrt{2\pi}} \exp\left(-\frac{(x_i - x)^2 + (y_i - y)^2}{2s}\right) \quad (14)$$

where q_i is the formal charge of atom i . The standard deviation $s = 3 \text{ \AA}$ is chosen so that $\Delta\rho$ vanishes in bulk regions.

Fig. S14a shows the charge density of a model containing two O-O-cored climb-dissociated dislocations before energy minimization, with the atomic arrangements around the two dislocations magnified in Figs. S14b-c. The partial dislocations carry localized negative charges, while the APB junction introduces a region of positive charge. The charge density of Sr-Ti-cored dislocations is shown in Fig. S14d (with magnified views in Figs. S14e-f). It exhibits three positive charge concentrations: two from the partial cores themselves and one from the APB junction.

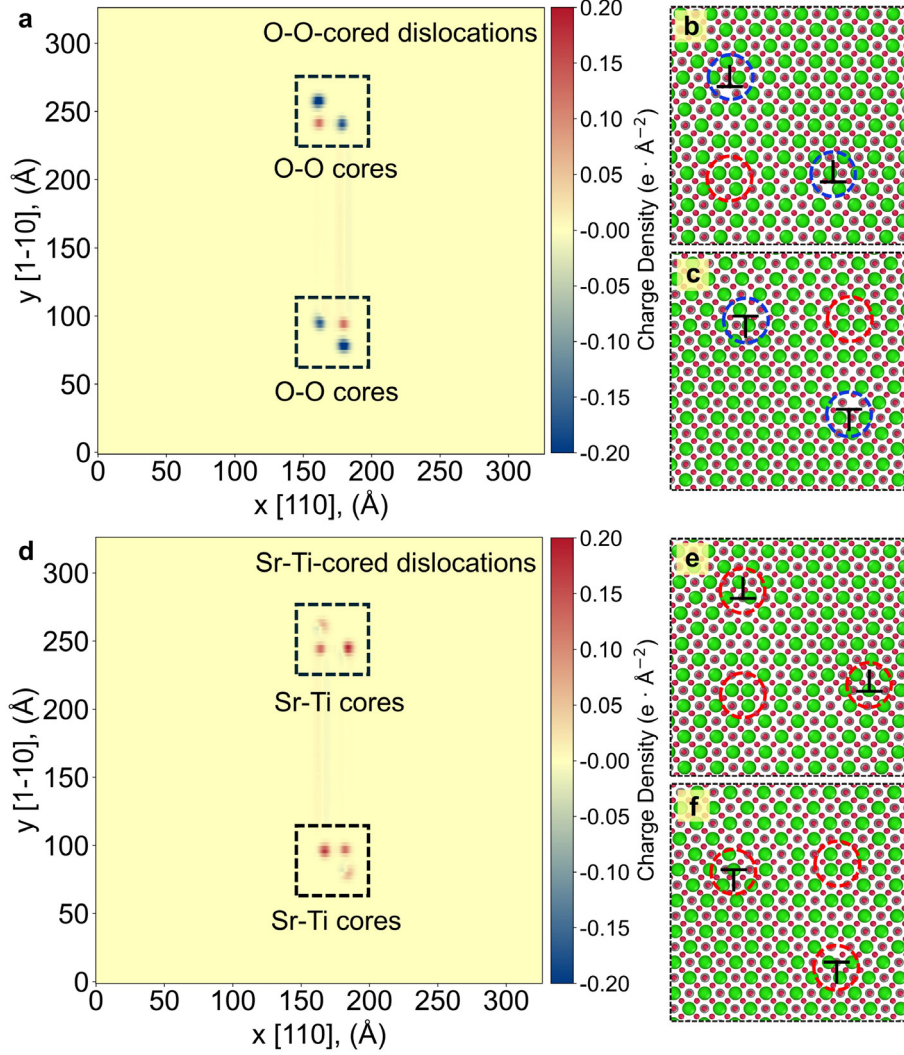


Fig. S14 Charge density analyses of distinct climb-dissociated dislocations in SrTiO₃. **a**, Charge density map of a model containing two climb-dissociated O-O-cored dislocations. **b-c**, Magnified atomic arrangements around the two dislocations shown in (a). **d**, Charge density

map of a model containing two climbed Sr-Ti-cored dislocations. **e-f**, Magnified atomic arrangement around the two dislocations shown in **(d)**. Source data are provided as a Source Data file.

Supplementary Note 10. The minimum energy configuration of climb-dissociated dislocations in SrTiO₃

The minimum energy structures of climb-dissociated dislocations in SrTiO₃ (Fig. S15) reveal distinct core preferences: the O-O-cored dislocation adopts a compact configuration ($w = 0.5b$, $h = 4b$), whereas the Sr-Ti-cored dislocation favours a wider, more extended dissociation ($w = 2.5b$, $h = 3b$).

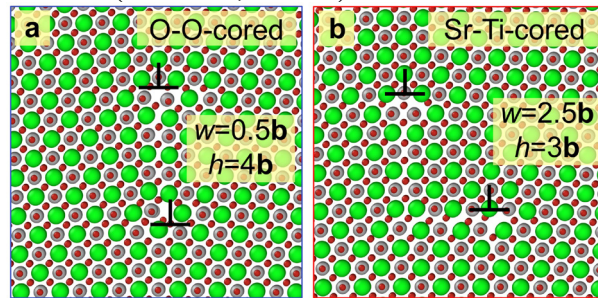


Fig. S15. Minimum-energy configurations of climb-dissociated dislocations in SrTiO₃. **a**, O-O-cored dislocation core structure. **b**, Sr-Ti-cored dislocation core structure.

Supplementary Note 11. The stable structures of climb-dissociated dislocations with Sr-Ti cores

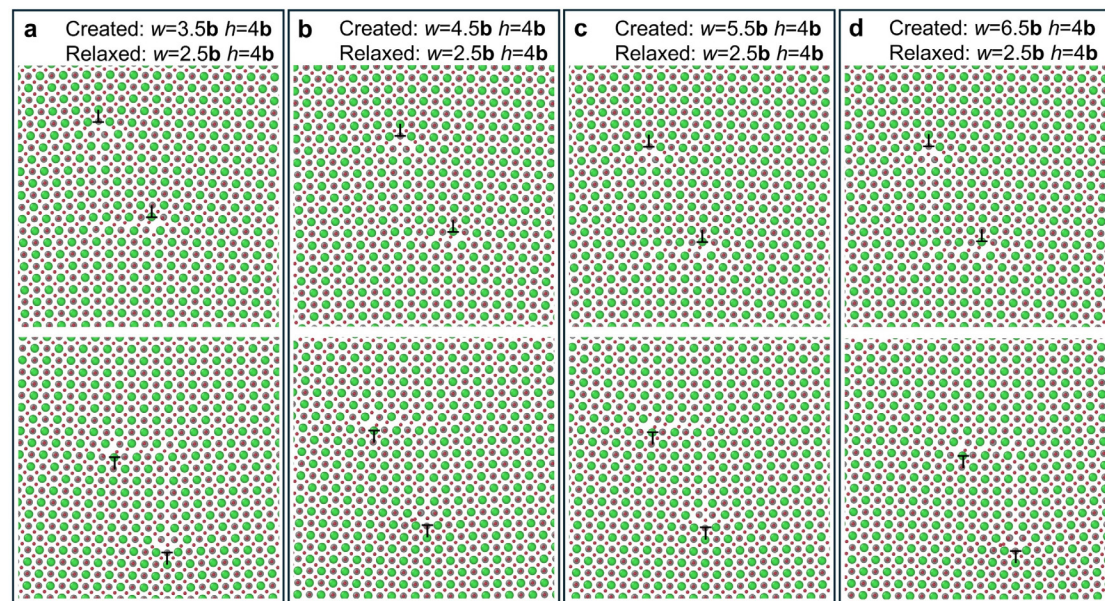


Fig. S16 Relaxed configurations of Sr-Ti-cored dislocations with w ranging from $3.5b$ to $6.5b$ at a climb height of $h = 4b$.

To investigate the stability of Sr-Ti-cored dislocations, models were constructed with varying dissociation widths w and climb heights h . Following full energy minimization, dislocations with $w \geq 3.5b$ spontaneously collapsed to $w = 2.5b$ when $h \geq 3b$. The minimum energy configurations were achieved at $h = 4b$, as shown in Fig. S16, indicating that stable climb-dissociated structures of cation-rich dislocations are constrained to a width of $2.5b$ for climb heights $\geq 3.5b$.

Supplementary References

- 1 Doukhan, N. & Doukhan, J. Dislocations in perovskites BaTiO₃ and CaTiO₃. *Phys. Chem. Miner.* **13**, 403-410 (1986).
- 2 Poirier, J., Beauchesne, S. & Guyot, S. Deformation mechanisms of crystal with perovskite structure. *Perovskite: A Structure of Great Interest to Geophysics and Material Science, Geophys. Monogr. Ser* **16** (1989).
- 3 Liu, Y. *et al.* Giant room temperature compression and bending in ferroelectric oxide pillars. *Nat. Commun.* **13**, 335 (2022).
- 4 Porz, L. *et al.* Dislocation-toughened ceramics. *Materials Horizons* **8**, 1528-1537 (2021).
- 5 Porz, L. *et al.* Conceptual Framework for Dislocation-Modified Conductivity in Oxide Ceramics Deconvoluting Mesoscopic Structure, Core, and Space Charge Exemplified for SrTiO₃. *ACS Nano* **15**, 9355-9367 (2021).
- 6 Fang, X. *et al.* Harvesting room-temperature plasticity in ceramics by mechanically seeded dislocations. *Mater. Today* **82**, 81-91 (2025).
- 7 Li, X. *et al.* Multislip-enabled morphing of all-inorganic perovskites. *Nat. Mater.* **22**, 1175-1181 (2023).
- 8 Greer, J. R., Oliver, W. C. & Nix, W. D. Size dependence of mechanical properties of gold at the micron scale in the absence of strain gradients. *Acta Mater.* **53**, 1821-1830 (2005).
- 9 Williams, D. B. & Carter, C. B. *Transmission Electron Microscopy: A Textbook for Materials Science.* (Springer, 1996).
- 10 Egerton, R. F. *Electron energy-loss spectroscopy in the electron microscope.* 3rd ed. edn, (Springer, 2011).
- 11 Yoo, S. J. *et al.* Comparing the Impacts of Strain Types on Oxygen-Vacancy Formation in a Perovskite Oxide via Nanometer-Scale Strain Fields. *ACS Nano* **18**, 18465-18476 (2024).
- 12 Was, G. S. *Fundamentals of radiation materials science: metals and alloys.* Vol. 73 (Springer, 2007).
- 13 Pols, M., Brouwers, V., Calero, S. & Tao, S. How fast do defects migrate in halide perovskites: insights from on-the-fly machine-learned force fields. *Chem. Commun.* **59**, 4660-4663 (2023).
- 14 Thomas, B. S., Marks, N. A. & Begg, B. D. Defects and threshold displacement energies in SrTiO₃ perovskite using atomistic computer simulations. *Nuclear Instruments and Methods in Physics Research Section B: Beam Interactions with Materials and Atoms* **254**, 211-218 (2007).
- 15 Hirel, P., Cordier, P. & Carrez, P. $\langle 110 \rangle$ - $\{ -110 \}$ edge dislocations in strontium titanate: Charged vs neutral, glide vs climb. *Acta Mater.* **285**, 120636 (2025).