

Encapsulation epitaxy of air-stable 2D superconductors for quantum circuits

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Two-dimensional (2D) superconductors are emerging platforms supporting both strongly correlated physics and quantum information science^{1,2}. Their reduced dimensionality, atomically flat interfaces and high crystallinity are particularly attractive for realizing compact lumped-element devices in superconducting circuits^{3–5}. However, large-scale synthesis of monolayer 2D superconductors remains challenging as they are easily oxidized in air⁶. Here we report an ‘encapsulation epitaxy’ mechanism that enables the growth of large-area (more than 1 inch), air-stable, monolayer niobium diselenide (NbSe₂) films (1L-NbSe₂) and explore their potential for superconducting quantum circuits. This work represents a distinct growth phenomenon in which a 2D encapsulation layer, such as graphene or hexagonal boron nitride, pre-deposited on a 3D substrate (for example, SiO₂ or Si₃N₄) simultaneously serves as a template for the epitaxial growth of 1L-NbSe₂ underneath it at the encapsulation–substrate interface and as a protective capping layer against ambient degradation. The as-grown 1L-graphene/NbSe₂ heterostructures exhibit robust superconductivity (superconducting transition temperature $T_c \approx 1$ K) and enhanced charge density waves (CDWs; CDW transition temperature $T_{CDW} \approx 177$ K). We further demonstrate the integration of 1L-NbSe₂ into superconducting circuits by developing oxidation-free transfer and superconducting edge-contact techniques. The 1L-NbSe₂ in these circuits feature a measured kinetic inductance $L_K \approx 0.7$ nH $\square^{-1}$, making it suitable for quantum circuits requiring elements with high kinetic inductance. This encapsulation-epitaxy methodology enables the production of air-stable 2D superconductors and van der Waals heterostructures, holding promise for wafer-scale, monolithic fabrication of superconducting quantum circuitry.

Two-dimensional (2D) superconducting transition metal dichalcogenides (SC-TMDs) are a materials system where strongly correlated physics and quantum information science converge^{1,2}. At reduced dimensionality, quantum confinement significantly modifies the behaviour of electronic phases in SC-TMDs, such as charge density waves (CDWs)⁷, quantum spin liquids⁸ and superconductivity⁹, leading to distinct phenomena compared with their bulk counterparts, making 2D SC-TMDs an ideal platform for studying correlated physics. The atomically flat interfaces and layered structure of SC-TMD heterostructures make them promising candidates for realizing quantum devices with significantly reduced footprint and crosstalk^{3–5}. For instance,

superconducting transmon qubits incorporating niobium diselenide (NbSe₂)/hexagonal boron nitride (hBN)/NbSe₂ parallel-plate capacitors have achieved coherence times comparable to those of conventional aluminium (Al)-based qubits, while occupying over 250 times less area³. However, monolayer SC-TMDs are highly air sensitive, and their superconductivity is easily disrupted by external perturbations and local defects. The growth of high-quality monolayer SC-TMDs has been a critical challenge⁶, hindering their application towards next-generation quantum devices.

Unlike semiconducting TMDs (such as MoS₂ and WSe₂), which have seen rapid progress in synthesis techniques^{10–12}, the difficulties in the

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growth of monolayer SC-TMDs come from the high melting points and low surface diffusivity of the precursors, which hinder the formation of continuous and uniform monolayers¹³. For instance, attempts to synthesize monolayer SC-TMDs using molecular-beam epitaxy^{14,15} or molten-salt assisted chemical vapour deposition (CVD)^{13,16} have yielded only micrometre-sized flakes with non-uniform layer thickness. Although electrochemical exfoliation produces monolayer flakes with larger lateral sizes, the thickness and uniformity of these films are poorly controlled in general¹⁷. In another two-step CVD approach, a metal film is first deposited on the target substrate and subsequently converted into SC-TMDs, which has enabled the synthesis of wafer-scale SC-TMDs^{18,19}. However, the average grain sizes of these films are in the nanometre regime (approximately 8.2 nm for monolayer NbSe₂ (ref. 19) and approximately 50–100 nm for multilayer SC-TMDs¹⁸), which may compromise the material quality (for example, more scattering at grain boundaries). At present, the reliable synthesis of homogeneous, highly crystalline, wafer-scale monolayer SC-TMD films remains a significant challenge.

To tackle the air-sensitivity, non-uniformity and crystallinity challenges of NbSe₂, it is instructive to consider growth strategies demonstrated in other material systems. Previous studies have shown that graphene can enable the growth of materials beneath it^{20–24}. These studies highlight the ability of using a 2D template layer to act as a mechanically robust yet permeable cover layer that enables confined growth and stabilizes otherwise difficult-to-grow phases.

In addition to superconductivity, SC-TMDs usually exhibit CDW order²⁵. A CDW is the periodic modulation of the conduction electron density and its associated lattice distortion. At atomically thin thickness, the electron–phonon interaction is strengthened and leads to more prominent CDW phenomena. For instance, monolayer NbSe₂ (1L-NbSe₂) has been reported to exhibit strongly enhanced CDWs with transition temperatures (T_{CDW}) as high as 145 K (ref. 7). However, the observation of CDW phase transition has been limited primarily to exfoliated and molecular-beam-epitaxy-grown samples^{7,14,15}, and the observation of CDW order in CVD-grown NbSe₂ has yet to be reported.

In this work, we report an ‘encapsulation epitaxy’ strategy that enables the growth of large-scale, air-stable superconducting 1L-NbSe₂ films at the interface of a ‘2D–3D hybrid substrate’, and is extendable to growing other TMD materials with high-melting-point precursors. The 2D–3D hybrid substrate comprises a 2D layer (graphene or hBN) pre-deposited on a three-dimensional (3D) amorphous substrate (silicon dioxide (SiO₂) or silicon nitride (Si₃N₄)), where the 2D layer serves both as an epitaxial growth template and as an encapsulation layer against ambient degradation of NbSe₂. Density functional theory (DFT) calculations show that the 2D–3D substrate combines the advantages of a low diffusion barrier present at 2D material surfaces and a high adsorption energy present at 3D SiO₂ surfaces, which together ensures monolayer growth in the van der Waals gap with ultra-clean interfaces. We observed robust CDW formation comparable to exfoliated samples in both CVD graphene/NbSe₂ and hBN/NbSe₂. We further demonstrate a reliable process for integrating our CVD-grown 1L-NbSe₂ into superconducting circuits, enabled by oxidation-free, wafer-scale transfer of SC-TMD thin films and by a superconducting, one-dimensional edge-contact technique that shows negligible contact resistance at low temperatures. Finally, we extract the kinetic inductance of the CVD monolayer NbSe₂ via microwave characterization, highlighting the potential of CVD-grown SC-TMDs as a material platform for developing extensive quantum computing architectures based on 2D materials.

Encapsulation epitaxy

The encapsulation-epitaxy growth process is illustrated in Fig. 1a. First, a layer of 2D material (such as graphene or hBN) is transferred to a 3D amorphous SiO₂ substrate. Then this 2D–3D hybrid substrate is placed inside a CVD furnace for the synthesis of the 1L-SC. Under the growth

conditions we have designed for this work (see Methods for details), no growth happens on the exposed 2D surface. Instead, the precursors enter the van der Waals gap at the 2D–3D interface, nucleate and form epitaxial 1L-SC flakes, and eventually merge into a continuous 1L-SC film at the interface. No multilayer growth is observed during this process (Fig. 1b). Such growth behaviour is not observed on the exposed 3D or 2D surfaces under the same growth conditions. For example, on the exposed 3D surface, only isolated thick flakes with non-uniform thickness, typically >100 nm, are detected (Extended Data Fig. 1; it is noted that this growth was performed under a metal-precursor-rich condition; see ‘Selective-area growth’ in Methods for further discussion). On the exposed 2D surface, no growth of the 1L-SC is observed, as evidenced by Raman characterization (Extended Data Fig. 2).

These growth behaviours are elucidated by DFT calculations in combination with thermodynamic and kinetic modelling^{26,27}. In CVD thin-film growth, the surface diffusion coefficient D_s determines how quickly adatoms can travel on the surface and follows an Arrhenius form²⁸:

$$D_s = D_0 \times e^{-\frac{E_{\text{diff}}}{k_B T}} \quad (1)$$

where D_0 is the pre-exponential factor, E_{diff} is the surface diffusion energy barrier, k_B is the Boltzmann constant and T is the growth temperature. Meanwhile, the residence time τ determines how long adatoms can stay on the surface before desorbing (thus the likelihood of adatom incorporation) and is defined as²⁷:

$$\tau = \nu^{-1} \times e^{\frac{E_{\text{ads}}}{k_B T}} \quad (2)$$

where ν is the surface vibration frequency and E_{ads} is the adsorption energy between the adatom and the substrate. These equations connect the surface energies (E_{diff} , E_{ads}) with the kinetic parameters (D_s , τ).

On the exposed SiO₂ surface, DFT calculations reveal a high diffusion barrier for NbSe₂ adatoms ($E_{\text{diff}} \approx 1.0$ eV), which greatly exceeds the available thermal energy at the relevant growth temperature (950 °C). According to equation (1), the diffusion coefficient is relatively low ($D_s \approx 7.6 \times 10^{-8}$ cm² s⁻¹). It severely limits the lateral adatom migration that is critical for layer-by-layer growth. Meanwhile, NbSe₂ binds strongly to SiO₂ ($E_{\text{ads}} \approx 2.6$ eV), resulting in a relatively long residence time ($\tau \approx 5.2$ ms) that promotes nucleation. These factors make vertical growth more favourable than lateral growth, leading to the formation of thick flakes on bare SiO₂. In contrast, the graphene surface presents a lower diffusion barrier ($E_{\text{diff}} \approx 0.3$ eV), enabling fast surface diffusion with a diffusion coefficient over three-orders-of-magnitude higher than on SiO₂ ($D_s \approx 5.8 \times 10^{-5}$ cm² s⁻¹). However, graphene’s weaker adsorption energy ($E_{\text{ads}} \approx 1.9$ eV) results in an approximately three-orders-of-magnitude shorter residence time ($\tau \approx 6.8 \times 10^{-3}$ ms). This increases the desorption rate and significantly suppresses nucleation, resulting in minimal growth on graphene. In short, SiO₂ favours nucleation but hinders lateral growth; graphene supports adatom mobility but fails to sustain nucleation. Encapsulation epitaxy overcomes these limitations via a 2D–3D hybrid substrate design, which combines the high adsorption energy of the 3D substrate surface to promote a sufficient residence time and nucleation, and a low diffusion barrier of 2D layer surface for efficient lateral migration. This synergistic effect enables uniform monolayer formation at the 2D–3D interface (additional DFT calculations and substrate-dependent growth results shown in Extended Data Figs. 3 and 4; additional discussion is provided in Methods).

To further elucidate the precursor diffusion and nucleation initiation locations, we have conducted extensive growth studies and characterizations on exfoliated hBN and CVD graphene under various growth times and conditions, as shown in Supplementary Figs. 1–4. The nucleation of NbSe₂ under hBN mainly initiates at edges, cracks and wrinkles, although we cannot exclude the possibility of nucleation initiating

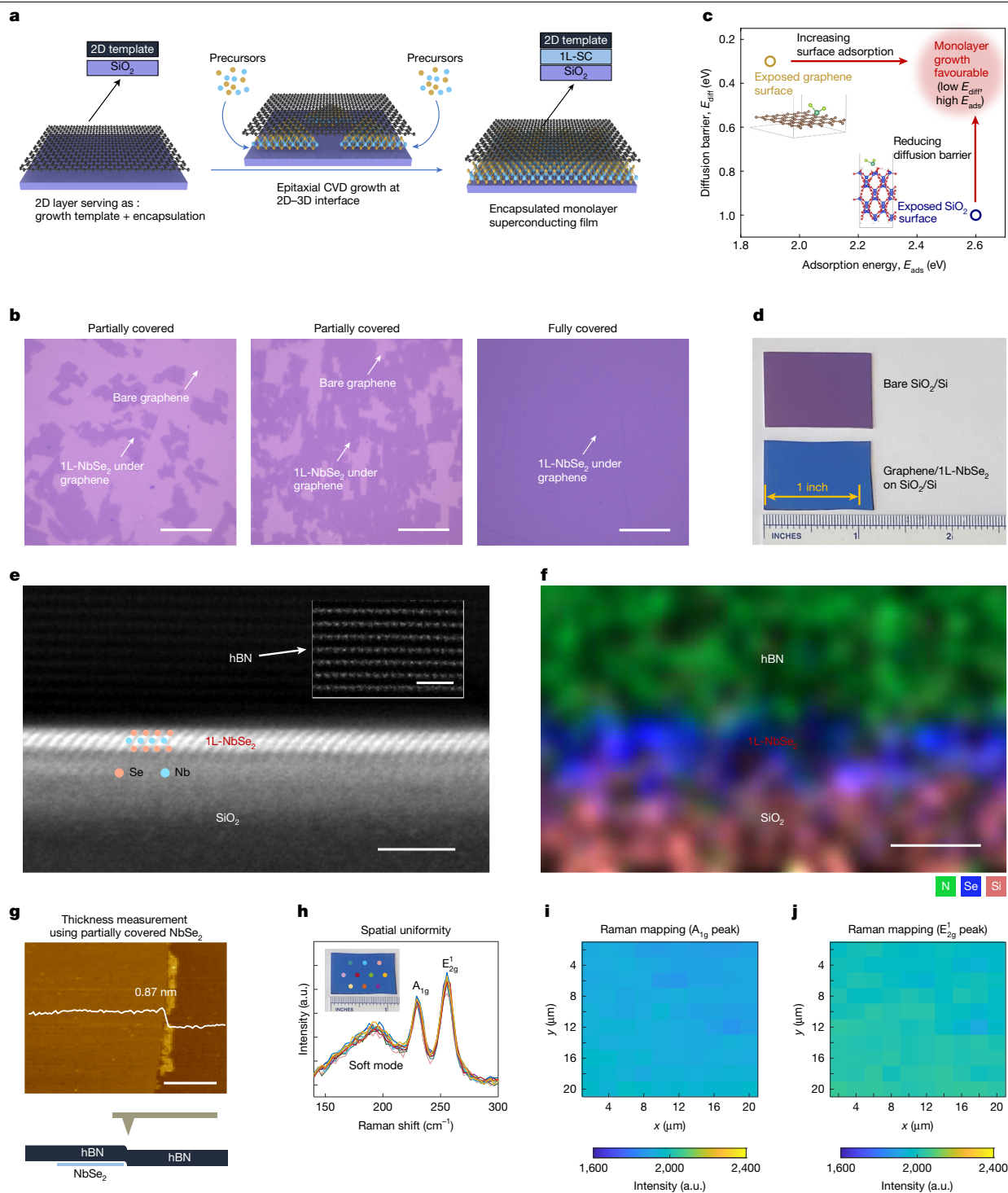


Fig. 1 | Encapsulation-epitaxy growth process and resulting 1L-NbSe₂ film.

a, Schematic illustration of the encapsulation-epitaxy growth process. The growth of 1L-SC happens in the van der Waals gap between the 2D template layer and the 3D amorphous SiO₂ substrate. **b**, Optical micrographs of CVD-grown 1L-NbSe₂ with different growth coverage, confirming the growth of 1L-NbSe₂ only with no multilayer growth. Additional characterization of the partially grown NbSe₂ can be found in Supplementary Fig. 1 and Methods. **c**, DFT-calculated surface diffusion barrier and adsorption energy of NbSe₂ on exposed graphene (yellow circle) and SiO₂ (blue circle) surfaces. The red-shaded 'Monolayer growth favourable' region is a qualitative guide, not derived from DFT. **d**, Photo showing the as-grown graphene/NbSe₂/SiO₂/Si and its comparison with bare SiO₂/Si. **e**, Cross-sectional STEM image of 1L-NbSe₂ grown at the interface between an hBN encapsulation layer and a SiO₂ substrate. The region above the 1L-NbSe₂ in the image is fully occupied by hBN, but the

contrast is low as it is adjusted for optimal visualization of NbSe₂. Inset: a zoomed-in, contrast-corrected image of the hBN area. **f**, Energy-dispersive X-ray spectroscopy mappings for the sample in e. **g**, AFM measurement of 1L-NbSe₂ grown underneath hBN via encapsulation epitaxy. A partially covered NbSe₂ sample is used, containing regions with only hBN and regions with hBN/NbSe₂, to enable the measurement of the 1L-NbSe₂ thickness from the height difference of the hBN surface, as illustrated in the bottom diagram. **h**, Spatial uniformity of NbSe₂ characterized by Raman spectroscopy at ten different locations across the sample shown in d. The measurement locations are indicated by coloured dots in the inset photo. **i, j**, Raman mapping of 20 × 20 μm² area indicating the uniformity of the NbSe₂ (optical image and all Raman spectra shown in Extended Data Fig. 5c,d). Panel i shows the A_{1g} peak intensity mapping and panel j shows the E_{2g}¹ peak intensity mapping. Scale bars, 20 μm (b), 2 nm (e,f), 1 nm (inset in e), 500 nm (g).

from defects. The monolayer CVD-grown graphene template shows a higher density of wrinkles and defects than exfoliated hBN, which further enhances the nucleation probability and lateral growth rate (see Supplementary Fig. 5 for statistical Raman characterizations of graphene defects before and after growth). However, this defect generation is limited and the graphene lattice remains largely intact, providing effective encapsulation against ambient exposure (see Methods for further discussion about the possible trade-off between growth uniformity and environmental stability). The combined edge-, wrinkle- and defect-assisted pathways enable continuous monolayer-film formation at the 2D–3D interface.

Figure 1d shows a photograph of inch-scale CVD-grown 1L-NbSe₂ at the graphene/SiO₂ interface. The growth is enabled by innovations in both substrate engineering and precursor delivery. The substrate uses the special 2D–3D hybrid approach, as discussed in the previous section. The precursor delivery is based on a modified face-to-face strategy and utilizes vapour-phase salt as growth promoters (Supplementary Fig. 6 and Methods), enabling highly scalable and uniform growth. The as-synthesized NbSe₂ films exhibit excellent growth uniformity (Fig. 1h–j and Extended Data Fig. 5c,d) and monolayer thickness, as determined by cross-sectional high-angle annular dark-field scanning transmission electron microscopy (HAADF-STEM) imaging (Fig. 1e and Supplementary Fig. 7), atomic force microscopy (AFM) measurements (Fig. 1g), and the absence of an interlayer shear mode between 20 cm⁻¹ and 30 cm⁻¹ in the Raman spectra (Fig. 2c). The material also shows a well-defined chemical structure, supported by Raman and X-ray photoelectron spectroscopy (XPS) analyses (Extended Data Fig. 5). The fact that growth occurs at the 2D–3D interface, rather than on the exposed top surface of the 2D template, is confirmed by a combination of cross-sectional STEM imaging (Fig. 1e), and the corresponding energy-dispersive X-ray spectroscopy mapping (Fig. 1f), cryogenic dry pick-up test (Extended Data Fig. 2) and air-stability characterization (Extended Data Fig. 6). Notably, the encapsulation layer significantly enhances the air stability of NbSe₂ throughout the growth and post-growth characterization and fabrication processing, including robust Raman characterization acquired in ambient after 2 weeks of air exposure (Extended Data Fig. 6). When we peeled off the encapsulation layer, which can be done for NbSe₂ grown at the hBN–SiO₂ interface (Extended Data Fig. 2), only the hBN was picked up while NbSe₂ stayed on the SiO₂. After hBN removal, the exposed 1L-NbSe₂ started to degrade immediately owing to exposure to air. Ambient Raman characterization was no longer feasible, because even low-power and short-time laser excitations (approximately 140 μW μm⁻² for 3 s) will create a hole in the un-encapsulated NbSe₂ (Extended Data Fig. 2); this is one example that clearly illustrates the ambient stability enabled by encapsulation epitaxy.

In Fig. 2a, the HAADF-STEM image shows the top-view, atomic-scale lattice structure of as-synthesized 1L-NbSe₂, with a well-defined hexagonal lattice for 1L-NbSe₂ with 1H crystal structure (see Supplementary Fig. 8 for extensive STEM characterizations). The extracted lattice constant is 0.34 nm, in good agreement with previous literature²⁹. The encapsulation graphene is not visible in the HAADF-STEM image owing to the much smaller atomic number of carbon as compared with niobium and selenium. Figure 2b shows the selected area electron diffraction (SAED) pattern of NbSe₂, which clearly shows the diffraction pattern of both the NbSe₂ and the encapsulation graphene layers. Additional SAED images are provided in Extended Data Fig. 7. These diffraction patterns are taken across a 200 × 200 μm² area, and the aligned patterns of NbSe₂ and graphene indicate that the NbSe₂ is epitaxially aligned with the graphene lattice. This is consistent with the recent ‘hypotaxy’ growth of single-crystalline MoS₂ underneath graphene³⁰. However, unlike the previous approach where metal precursors were pre-deposited on the substrate before graphene encapsulation, in our approach, the precursors are delivered in gas phase directly into the pre-formed graphene/SiO₂ or hBN/SiO₂ interface. In addition, the

top graphene (or hBN) layer remains after growth is finished, serving as an effective encapsulation layer afterwards. This approach further supports the selective-area growth of NbSe₂ using patterned graphene on a SiO₂ substrate (Extended Data Fig. 8; see also Methods for details of controlling the gas-phase metal-precursor concentration in the low-mass-flux regime), which indicates the possibility of obtaining patterned structures directly from growth to facilitate specific fabrication needs in the future. Moreover, this approach can also be extended to the growth of other TMD materials with high-melting-point precursors, such as VSe₂, VS₂, NbS₂ and WTe₂ (see Extended Data Fig. 9 for preliminary results).

CDW in CVD 1L-NbSe₂ film

The CVD-grown 1L-NbSe₂ shows robust electronic ordering, as manifested by the robust CDW signatures comparable to, or even surpassing, those of exfoliated 1L-NbSe₂. Figure 2c,d shows CDW measurements of 1L-graphene/NbSe₂ on SiO₂/Si substrate via temperature-dependent Raman characterization from 30 K to 250 K (additional characterization shown in Supplementary Fig. 9). The monolayer thickness of NbSe₂ is validated by the absence of an interlayer shear mode⁷ between 20 cm⁻¹ and 30 cm⁻¹. As the temperature decreases, we observe the clear emergence of a CDW amplitude mode and a weak, high-frequency feature (marked in the figure), two characteristics of CDWs in NbSe₂. The extracted transition temperature (T_{CDW}) for graphene/NbSe₂ is approximately 177.0 K (Fig. 2e). Similar results are observed on 1L-NbSe₂ grown at the hBN/SiO₂ interface (Extended Data Fig. 10; $T_{\text{CDW}} \approx 185.1$ K). These CDW features are in good agreement with a previous study that measured exfoliated 1L-NbSe₂ (ref. 7). The pronounced CDW features observed here indicate robust electronic ordering in the CVD-grown monolayer NbSe₂ films³¹, and, together with the structural and spectroscopic characterizations presented in Fig. 1, support their overall high quality. Both the G and 2D bands of the graphene on top of NbSe₂ show pronounced blueshifts compared with graphene on SiO₂ (Fig. 2f and Supplementary Figs. 10 and 11), indicating the introduction of both strain and doping on the graphene template after growth^{32,33}, and are respectively attributed to the heating and cooling process of CVD growth and the interlayer charge transfer effect between graphene and NbSe₂.

Superconductivity in CVD 1L-NbSe₂ film

We next investigated the low-temperature electrical transport properties of air-stable, low-defect 1L-NbSe₂ films. For this purpose, we first developed a specialized set of fabrication methodologies tailored for air-sensitive 2D materials, as illustrated in Fig. 3a–c. Following the synthesis by encapsulation epitaxy, we performed an oxidation-free transfer based on the electrical double layer (EDL) force transfer process³⁴, which enables the assembly of heterostructures with additional van der Waals materials and the larger-scale integration with other substrates and circuits (Supplementary Fig. 12). An example van der Waals heterostructure made of hBN, NbSe₂, graphene and MoS₂ is shown in Supplementary Fig. 13, which simultaneously shows the Raman peaks of each functioning layer.

The four-probe measurements confirm that the 1L-NbSe₂ films are superconducting at low temperatures after transfer and device fabrication processes. The device was fabricated in an edge-contact configuration, in which graphene/NbSe₂ heterostructures were transferred onto a new SiO₂/Si substrate by the EDL transfer method and contacted at the heterostructure edges with Al electrodes (see Supplementary Figs. 12–15 and Methods for fabrication and measurement details). As shown in Fig. 3d,e, the measured resistance in the four-probe set-up exhibits metallic behaviour ($dR/dT > 0$) at high temperatures and a superconducting transition with critical temperature $T_c \approx 1$ K, which agrees with previous literature^{14,16}. We measure the differential resistance (dI/dI)

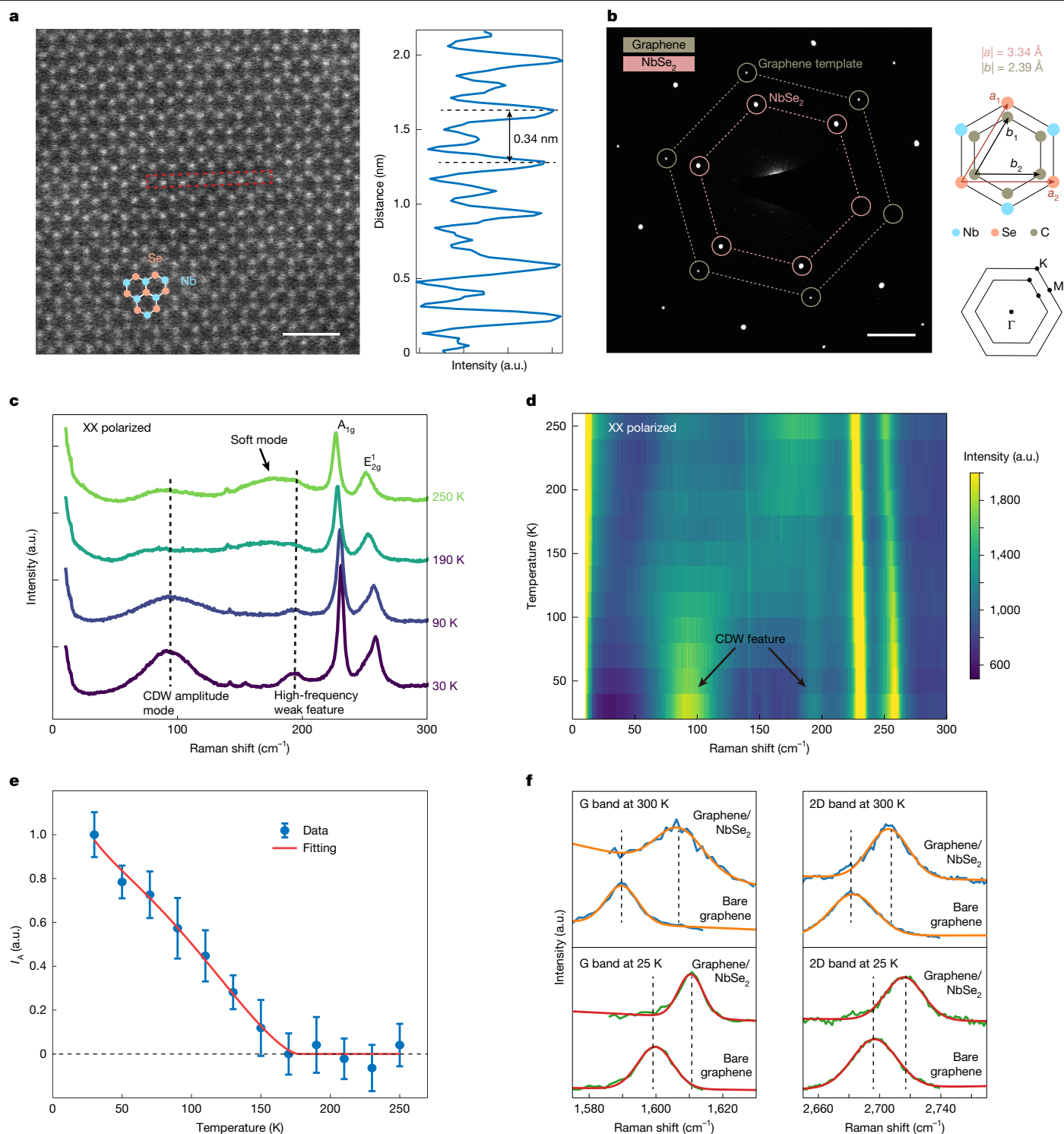


Fig. 2 | Atomic imaging and CDW characterization. **a**, Left: HAADF-STEM image of a 1L-graphene/NbSe₂ sample showing the lattice structure of NbSe₂. The graphene is not visible owing to the much lower atomic number of carbon compared with niobium and selenium. Right: line profile from the region outlined by the red box in the STEM image. **b**, SAED pattern of the 1L-graphene/NbSe₂. **c**, Temperature-dependent high-resolution Raman spectra on 1L-graphene/NbSe₂ heterostructure synthesized by encapsulation epitaxy. The XX denotes that the incident excitation laser is linearly polarized, and only the linear, parallel polarization is collected at the spectrometer entrance slit. **d**, Temperature map of the Raman scattering intensity of 1L-graphene/NbSe₂

heterostructure. **e**, Temperature dependence of the CDW amplitude mode intensity (I_A) for a 1L-graphene/NbSe₂ heterostructure (see Methods for extraction details). The error bars are calculated based on the standard deviation of the Raman scattering intensity around the amplitude mode peak and the spectral width used for integrating the mode. The solid lines are fits to mean-field theory⁷ and serve as guides to the eye. **f**, Comparison between 1L-graphene/NbSe₂ and bare 1L-graphene in terms of the graphene Raman vibrational modes at both room temperature (300 K) and cryogenic temperature (25 K). Gaussian fits with a linear background are overlaid to locate the peak positions. Scale bars, 1 nm (**a**), 2 nm⁻¹ (**b**).

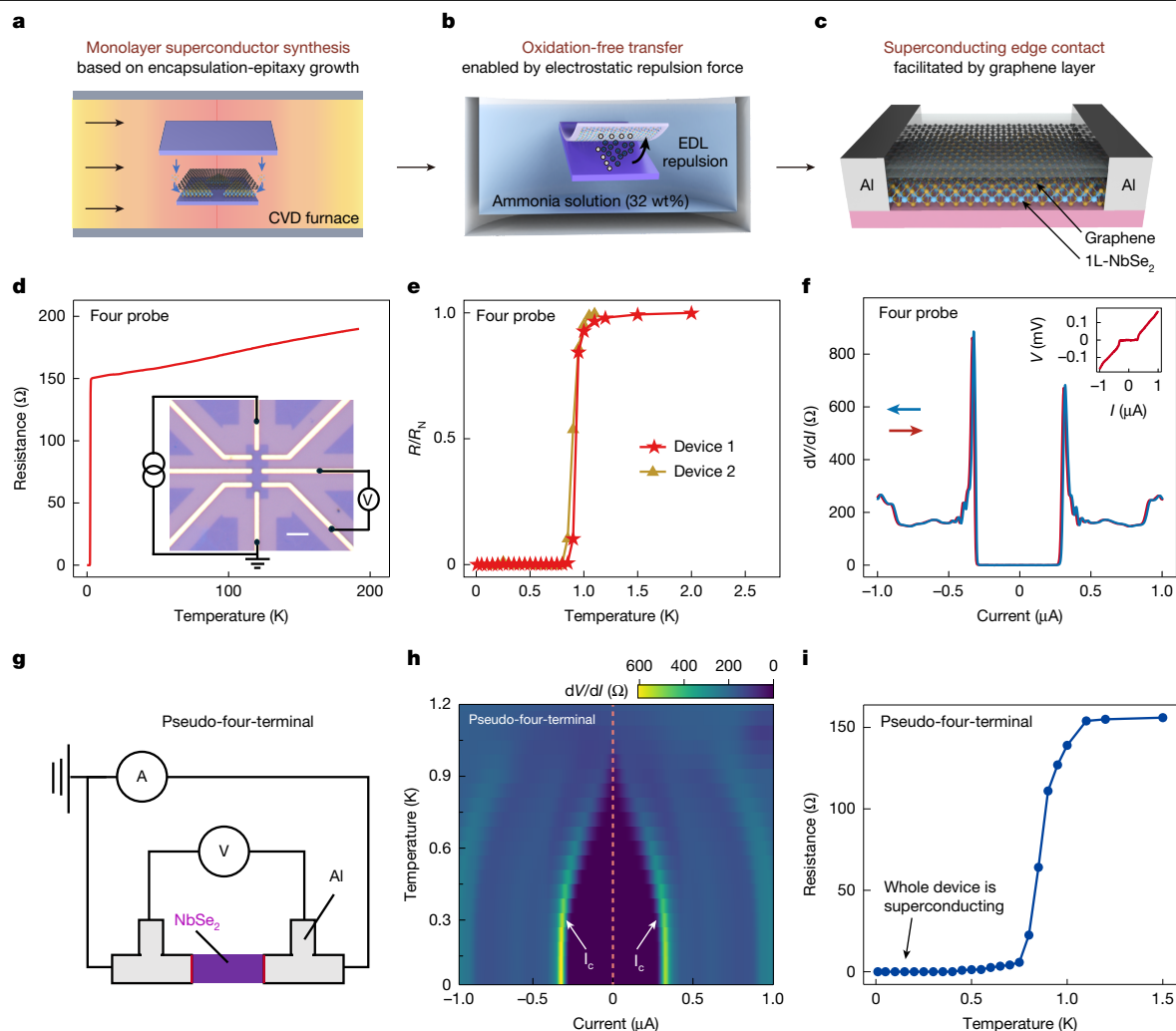


Fig. 3 | Superconductivity and superconducting edge contact.

a–c. Schematics of the fabrication process developed in this work for integration into superconducting circuits, which includes large-area synthesis of 1L-NbSe₂ using encapsulation epitaxy (**a**), oxidation-free transfer of 1L-NbSe₂ to target substrates (**b**) and a graphene-facilitated superconducting edge contact (**c**). **d.** Four-probe resistance versus temperature measurements of the 1L-graphene/NbSe₂ film with edge contacts. Inset: optical image of the fabricated device. Scale bar, 5 μm. **e.** Superconducting transition of 1L-NbSe₂ with edge-contact configurations. R , resistance; R_N , normal-state resistance. **f.** Differential resistance (dV/dI) as a function of direct-current bias current measured at 10 mK using an alternating-current excitation of 5 nA. The data

were acquired for both increasing (red) and decreasing (blue) current sweeps. Inset: current–voltage (I – V) curves measured in both sweeping directions show no discernible hysteresis within the resolution of this measurement. **g.** Schematic of the pseudo-four-probe configuration used to characterize the superconducting edge contact to Al electrodes. **h.** Differential resistance (dV/dI) measured in a pseudo-four-probe configuration as a function of temperature from 150 mK to 1.2 K. The dV/dI approaches zero in the dark blue region, indicating zero series resistance at the Al and NbSe₂ edge-contacted interfaces. **i.** Temperature dependence of the differential resistance at zero direct-current bias ($I_{DC} = 0$ μA line cut in **h**).

and current–voltage (I – V) characteristics of the edge-contacted device by sweeping the direct-current bias current (I_{DC}) from -1 μA to 1 μA, as shown in Fig. 3f, and the critical current density (J_c) for 1L-NbSe₂ is inferred to be 3.7×10^8 A m⁻². Low hysteresis is observed, indicating minimal joule heating and efficient thermal dissipation within the device³⁵.

Pseudo-four-terminal measurements³⁶ further confirm that superconducting edge contacts with zero contact resistance are achieved in edge-contacted graphene/NbSe₂ devices at low temperature, as shown in Fig. 3g–i (see Methods for more discussion). Superconducting contacts are essential for integrating 1L-SCs into quantum circuits through ohmic interfaces, thereby preserving high internal quality factors Q_i and minimizing dissipation. Our results demonstrate that encapsulation epitaxy enables the direct growth of heterostructures with clean, bubble-free interfaces, making them suitable for a wide range of applications. In this system, the graphene layer not only protects

NbSe₂ from ambient air but also may assist in forming high-quality superconducting contacts.

Integration with superconducting circuit

We demonstrate that encapsulation-epitaxially grown 1L-NbSe₂ can be integrated with conventional Al superconducting circuits. Figure 4a,b shows the superconducting circuit used to characterize 2D superconductors in the microwave regime^{37,38}. The measured NbSe₂ heterostructure comprises MoS₂/graphene/NbSe₂/hBN layers. All the components are continuous monolayer films grown by CVD (graphene, NbSe₂) or metal–organic CVD (MoS₂), except for hBN, which is exfoliated multilayer flakes (Supplementary Figs. 12 and 13). The chip is cooled to 10 mK in a dilution refrigerator (see Supplementary Fig. 16 for the wiring schematic), and the microwave transmission coefficient S_{21} is measured as a function of probe frequency using a vector

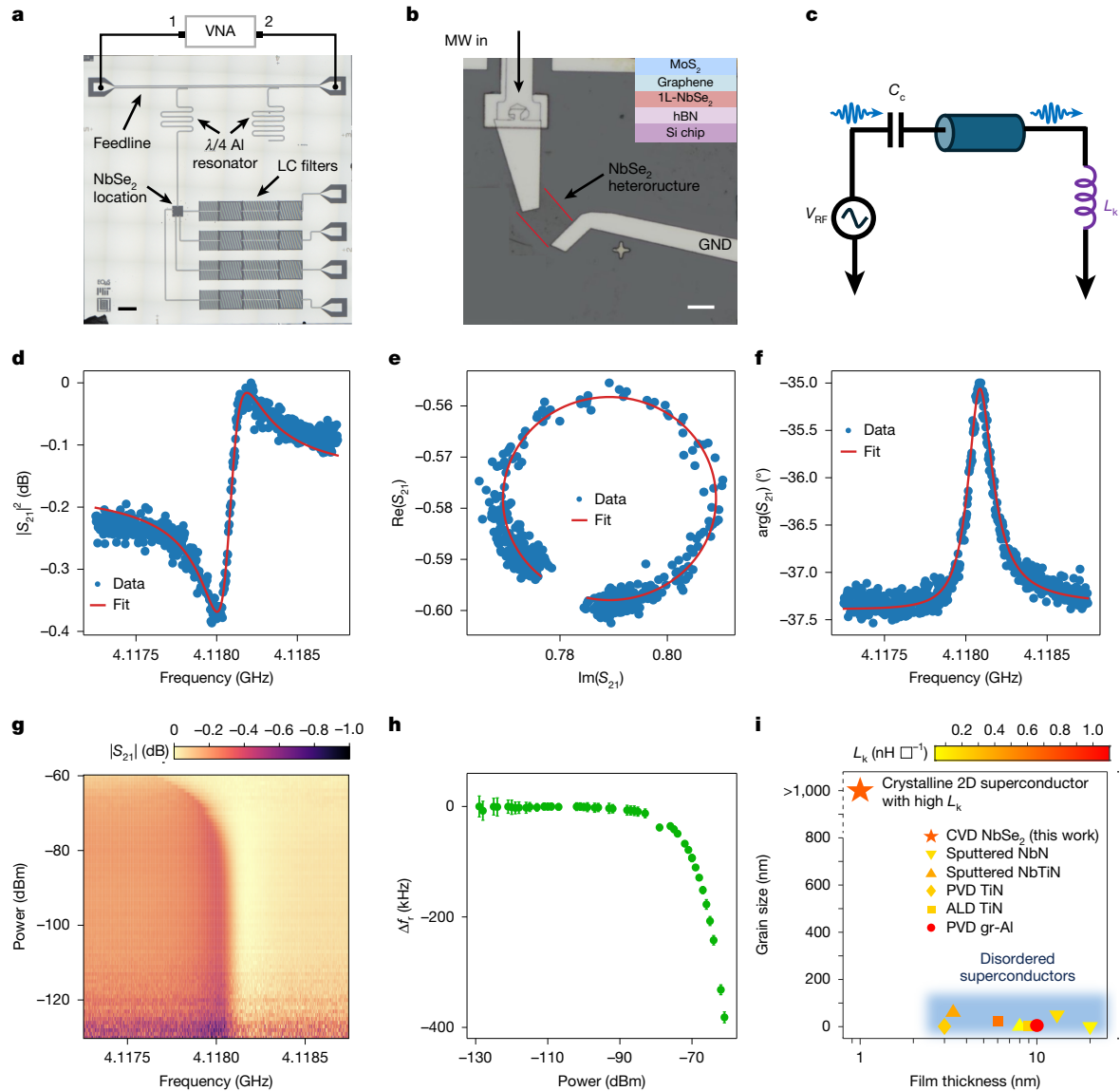


Fig. 4 | Integrating 1L-NbSe₂ into a superconducting circuit. **a**, Optical image of the superconducting circuit used to characterize the kinetic inductance of 1L-NbSe₂ films. Two $\lambda/4$ Al coplanar waveguide resonators are capacitively coupled to a common feedline connected to a vector network analyser (VNA). Port 1 sends the microwave input, and port 2 measures the transmitted output for the S_{21} measurement. The inductor–capacitor (LC) filters are connected to the DC-bias lines. The LC filters and associated DC-bias lines were included in the chip design but were not used for the measurements reported here. **b**, Optical image of 1L-NbSe₂ fully encapsulated by hBN (bottom) and graphene and MoS₂ (top) integrated onto the Si chip. The MoS₂, graphene and NbSe₂ are all monolayer continuous films synthesized via wafer-scale CVD techniques. The hBN is obtained by mechanical exfoliation. ‘MW in’ represents the input microwave signal; GND denotes the electrical ground of the chip. **c**, An equivalent circuit model for a $\lambda/4$ resonator terminated to ground via an inductor (1L-NbSe₂) and coupled capacitively to a feedline via coupling capacitor C_c . V_{RF} represents the voltage of the microwave input labelled ‘MW in’ in **b**. **d**, Resonator

network analyser. Figure 4d–f shows the magnitude, phase and complex I – Q data of the 1L-NbSe₂-terminated $\lambda/4$ resonator. A Lorentzian fit³⁹ determines the resonant frequency of the 1L-NbSe₂-terminated Al resonator $f_r = 4.118$ GHz and the internal quality factor $Q_i = 3.15 \times 10^4$ at a radio-frequency power of ~ 110 dBm at the device. The Q_i for the single-photon limit is 2.29×10^4 (at -125 dBm; Supplementary Fig. 17).

To extract the kinetic inductance of 1L-NbSe₂ heterostructure, we compare the microwave response of the NbSe₂-terminated resonator

spectroscopy (transmission coefficient $|S_{21}|^2$) of a $\lambda/4$ Al coplanar waveguide terminated ground with 1L-NbSe₂ film, with an applied radio-frequency power of ~ 110 dBm at the sample. **e**, Complex I – Q plot of the transmission coefficient (S_{21}), with blue dots showing the measured response and the red trace the Lorentzian fit. A resonant frequency $f_r = 4.118$ GHz and internal quality factor $Q_i = 3.15 \times 10^4$ are extracted from the circle fit (applied radio-frequency power of ~ 110 dBm). **f**, Phase of S_{21} as a function of frequency. **g**, Microwave power dependence of the NbSe₂-terminated $\lambda/4$ resonator. **h**, The resonant frequency shift, defined as $\Delta f_r = f_r(P) - f_r(P = -130 \text{ dBm})$, plotted as a function of the applied microwave power P . The observed downwards shift at higher powers indicates the onset of a nonlinear inductance in the superconducting film.

i, Comparison of high-kinetic-inductance films synthesized by extensible production techniques, focusing on film thickness, crystallinity (grain size) and kinetic inductance (L_k). ALD, atomic layer deposition; PVD, physical vapour deposition. Scale bars, 0.2 mm (**a**), 10 μm (**b**).

with that of an all-Al $\lambda/4$ resonator with identical geometry. From an analytical model³⁸, we extract the sheet kinetic inductance of the CVD 1L-NbSe₂ to be $0.71 \text{ nH } \square^{-1}$. It is noted that this value probably represents the lower bound for the kinetic inductance of CVD-grown 1L-NbSe₂ in isolation owing to the presence of the graphene layer. In addition, we sweep the microwave power from -130 dBm to -60 dBm at the device (Fig. 4g) and observe a resonant frequency shift of 400 kHz (Fig. 4h), which can be attributed to the Kerr nonlinearity of the

film. Finally, we benchmark our CVD-grown 1L-NbSe₂ film against other high-kinetic-inductance materials reported in the literature that are synthesized using wafer-scale-compatible approaches^{40–47}, as shown in Fig. 4i. Current state-of-the-art inductors primarily rely on amorphous or polycrystalline films with nanometre-scale grain sizes (disordered superconductors). The grain size of 1L-NbSe₂ is >1 μm, showing much better crystallinity than other materials such as NbN, NbTiN, TiN and granular aluminium (gr-Al). We note that the measured flake size in a partially grown 1L-NbSe₂ is about 1 μm; however, as the growth is epitaxial, grain size can be much larger than 1 μm assuming that the grain stitching is seamless. Therefore, the continuous, air-stable 1L-NbSe₂ film synthesized in this work is a highly crystalline monolayer superconductor with high kinetic inductance. This opens opportunities to construct quantum circuits based on wafer-scale superconducting 2D heterostructures using monolithic fabrication processes, making it well suited for superconducting quantum circuits in applications such as metrology⁴⁸, sensing⁴⁹ and quantum computing⁵⁰.

In conclusion, the encapsulation-epitaxy method developed in this work enables wafer-scale growth of monolayer superconducting films at the interface of a 2D vdW material (for example, graphene or hBN) and a 3D substrate (for example, SiO₂ or Si₃N₄) with enhanced air stability, demonstrating strong potential for use in fabricating superconducting quantum devices. The approach is applicable to other 2D material systems (Extended Data Fig. 9) and provides a route to directly synthesize van der Waals heterostructures, particularly benefiting air-sensitive monolayers. We envision that this encapsulation-epitaxy method will enable a broad class of monolayer materials for advancing technologies beyond quantum computing.

We became aware of a complementary but distinctive work²⁴ during the peer review of this paper.

Online content

Any methods, additional references, Nature Portfolio reporting summaries, source data, extended data, supplementary information, acknowledgements, peer review information; details of author contributions and competing interests; and statements of data and code availability are available at <https://doi.org/10.1038/s41586-026-10865-1>.

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Methods

Encapsulation epitaxy for growing monolayer NbSe₂

The encapsulation epitaxy was conducted using a hot-wall CVD system at ambient pressure with a fused quartz tube. We used niobium oxide films (obtained by first depositing 10 nm Nb on SiO₂/Si substrates by magnetron sputtering, and then oxidizing them in air at 500 °C for 5 h) and Se powder as the growth precursors. Before growth, the 2D–3D hybrid substrate (for example, graphene/SiO₂) was prepared by transferring or exfoliating a 2D layer onto SiO₂/Si (285 nm SiO₂) substrate. The 2D layers used in this work were either a layer of CVD-grown monolayer graphene film transferred from copper substrate or directly exfoliated hBN flakes. Supplementary Fig. 6 shows the set-up of the CVD reaction chamber. The 2D–3D hybrid substrate and niobium oxide film were placed face-to-face at the centre of the hot-wall furnace, with a distance of 1–3 mm between them. The face-to-face design ensures uniform precursor delivery to the growth substrate⁵¹, enabling uniform film growth. Sodium chloride (NaCl, 1.8 g) was placed at the upstream in the middle between the furnace centre and the Se boat. The Se powder (2.5 g; 99.5%, Sigma-Aldrich) was placed at the upstream edge. The carrier gas was 60 cubic centimetres per minute (sccm) argon (Ar) and 6 sccm hydrogen (H₂). The furnace was heated to 950 °C in 15 min and maintained at that temperature for 9 min for the growth of NbSe₂ at the 2D–3D interface. It is noted this is just one of the optimal growth conditions for obtaining large-scale continuous monolayer NbSe₂ at the interface of monolayer graphene and SiO₂. The growth time, temperature and precursor amount can also be adjusted depending on the target TMD materials and growth substrate. Then the top cover of the furnace was opened to allow fast cooling of the sample.

It is emphasized that the implementation of ‘face-to-face’ precursor delivery was made possible by a key innovation in NaCl vapour delivery. At the growth temperature of 950 °C, niobium oxide alone cannot evaporate; the presence of NaCl is essential, as it lowers the melting point of niobium oxide and facilitates its evaporation. However, when NaCl solids are physically mixed with the metal oxide precursor, NaCl tends to evaporate significantly faster than the metal precursor and becomes depleted before the growth is completed, which results in an unstable vapour environment and hinders the ability to achieve continuous monolayer growth. To overcome this problem, we developed a vapour-phase-NaCl-based salt delivery method. As illustrated in Supplementary Fig. 6a, NaCl is placed upstream in a lower-temperature zone, physically separated from niobium oxide. Upon heating, NaCl melts and evaporates at a nearly constant rate, which is determined by the local temperature (about 750–850 °C), throughout the growth. When NaCl vapour is carried by the Ar/H₂ carrier gas to the growth region, it interacts with niobium oxide that is placed face to face with the substrate, effectively lowering its melting point and enabling continuous and controlled evaporation of the niobium oxide precursor. All the precursors are therefore volatilized and transported to the substrate in the gaseous phase, and the nucleation process is thus consistent with vapour-phase CVD systems, although transient liquid-like intermediates at the buried interface cannot be entirely excluded. This vapour-phase salt delivery approach ensures a steady Nb precursor flux, which is critical for the uniform and reproducible growth of continuous 1L-NbSe₂ films.

Selective-area growth

The selective-area growth of NbSe₂ shown in Extended Data Fig. 8 is achieved using patterned graphene on a SiO₂ substrate. A key requirement for such selective-area growth is the suppression of unwanted nucleation on the exposed SiO₂ regions. We found that this suppression is strongly dependent on the local mass flux of metal precursors. Under high precursor concentrations, thick NbSe₂ flakes tend to form even on SiO₂, probably owing to elevated supersaturation that facilitates nucleation. However, when the Nb precursor concentration is reduced,

such parasitic growth on SiO₂ is effectively suppressed. Importantly, growth at the 2D–3D interface still persists, as the graphene edges act as favourable nucleation sites. In practice, this reduction in precursor concentration is achieved by increasing the distance between the niobium oxide source and the substrate from 1 mm to 2–3 mm, while keeping other growth parameters unchanged. The graphene patterns are defined using standard lithography and etching processes, allowing full customization of the pattern and shape.

DFT calculations

The first-principles calculations based on DFT are performed using the Vienna Ab initio Simulation Package (VASP)⁵². The exchange–correlation interactions are incorporated by the generalized gradient approximation in the form of Perdew–Burke–Ernzerhof. Core and valence electrons are treated by projector augmented wave method and plane-wave basis functions, respectively. The nudged elastic band⁵³ method is used to determine the diffusion barriers.

Calculation of diffusion coefficient and surface residence time

The diffusion coefficient follows an Arrhenius form $D_s = D_0 \times e^{-E_{\text{diff}}/k_B T}$, where D_0 is the diffusion prefactor, E_{diff} is the surface diffusion barrier extracted from DFT calculations, k_B is the Boltzmann constant and T is the growth temperature²⁷. The prefactor D_0 was taken to be $1 \times 10^{-3} \text{ cm}^2 \text{ s}^{-1}$, a typical value for surface diffusion of adatoms on solid surfaces²⁸. The surface residence time τ was estimated using: $\tau = v^{-1} \times e^{E_{\text{ads}}/k_B T}$, where E_{ads} is the adsorption energy and v is a characteristic surface vibration frequency, taken to be approximately 10 THz, which is frequently used for thermally activated surface processes^{26,27,54}. The adsorption energy E_{ads} is approximated as the magnitude of the DFT-calculated binding energy E_{binding} , which is defined as $E_{\text{binding}} = E_{\text{total}} - E_{\text{molecule}} - E_{\text{surface}}$, where E_{total} , E_{molecule} and E_{surface} are the energies of the adsorbed system, isolated molecule and isolated substrate, respectively. The growth temperature is $T = 950 \text{ °C}$. For NbSe₂ on SiO₂, with $E_{\text{diff}} \approx 1.0 \text{ eV}$ and $E_{\text{ads}} \approx 2.6 \text{ eV}$, we obtained $D_s \approx 7.6 \times 10^{-8} \text{ cm}^2 \text{ s}^{-1}$ and $\tau \approx 5.2 \text{ ms}$. For NbSe₂ on graphene, where $E_{\text{diff}} \approx 0.3 \text{ eV}$ and $E_{\text{ads}} \approx 1.9 \text{ eV}$, we obtained $D_s \approx 5.8 \times 10^{-3} \text{ cm}^2 \text{ s}^{-1}$ and $\tau \approx 6.8 \times 10^{-3} \text{ ms}$. These contrasting values illustrate the interplay between residence time and surface mobility, and help rationalize the substrate-dependent growth behaviours observed in our experiments.

Substrate-dependent growth test for understanding the growth mechanism

In addition to SiO₂, we further evaluated the applicability of encapsulation epitaxy using a 2D template on other amorphous 3D substrates, including Si₃N₄, HfO₂ and Al₂O₃. This assessment combined both DFT calculations and experimental growth tests, as presented in Extended Data Figs. 3 and 4. It turned out that encapsulation epitaxy can be successfully performed using Si₃N₄, with the as-grown NbSe₂ showing high uniformity (Supplementary Fig. 18). However, it was not suitable for HfO₂ and Al₂O₃, on which only thick flakes and black dense film were observed. DFT calculations (Extended Data Fig. 3) indicate that Si₃N₄ exhibits E_{adsorb} and E_{diff} values comparable to those of SiO₂, supporting both the nucleation and lateral growth conditions required by encapsulation epitaxy. In contrast, HfO₂ and Al₂O₃ show a significantly higher magnitude of E_{diff} , which hinders the ability of the 2D template to mediate the surface diffusion barrier. These findings are consistent with both our theoretical understanding and experimental observations.

Shape of the partially grown NbSe₂ flakes

It is noted that the apparent shape of the partially grown NbSe₂ shown in Fig. 1b does not reflect the true morphology of individual NbSe₂ crystals. As revealed in Supplementary Fig. 1, each optical-domain NbSe₂ region is actually composed of many small crystalline flakes with grain size of about 1 μm (as the growth is epitaxial, the real grain size can be much larger than 1 μm assuming that the grain stitching is

seamless). The larger shapes observed in the optical image are defined by graphene wrinkles. This suggests that NbSe₂ growth can be initiated at the wrinkle sites, which is reasonable considering the fact that the wrinkle location has a larger gap size and thus serves as the entrance for the precursors to enter the van der Waals gap. Another entry point for precursors is the edge of the graphene layers.

Verification of growth at 2D–3D interface by cryogenic dry pick-up test

The unusual growth behaviour at the 2D–3D interface in encapsulation epitaxy is evidenced by the cryogenic dry pick-up test and Raman characterization shown in Extended Data Fig. 2. In this test, a cryogenic dry pick-up technique based on cold polydimethylsiloxane (PDMS) (less than –90 °C) was utilized to pick up the as-grown hBN/NbSe₂ from the growth substrate (SiO₂). The pick-up process is detailed in a previous study⁵⁵. It turned out that only the hBN was picked up while NbSe₂ stays on the original substrate, which means that hBN is on top of NbSe₂. This, combined with the STEM cross-section image and superior air stability, confirms that the 1L-NbSe₂ is grown at the 2D–3D interface, instead of on top of the 2D surface.

Oxidation-free wafer-scale transfer of NbSe₂ by EDL force

As shown in the cryogenic pick-up test, we found it challenging for dry transfer methods to detach the CVD-grown 1L-NbSe₂, a difficulty that also applies to many other CVD-grown 2D materials. To enable device fabrication and circuit integration, we developed an oxidation-free, wafer-scale wet transfer method based on EDL forces utilizing ammonia solutions. A detailed discussion of the EDL transfer method is provided in our recent study³⁴. The key modification in the present study is that both the detachment from the growth substrate and attachment to the target substrate are conducted in an inert environment, as the detachment temporarily exposes the bottom surface of NbSe₂ to ambient, making it vulnerable to oxidation.

The transfer process began by spin-coating the graphene/NbSe₂ stack (also applicable to hBN/NbSe₂) with poly(methyl methacrylate) (PMMA) A4 (molecular weight 950,000) at 3,000 rpm for 60 s. The sample was then baked at 70 °C for 5 min. Both the spin-coat and bake steps were performed in air, as the top graphene layer provides sufficient protection for the NbSe₂. The PMMA/graphene/NbSe₂ film was subsequently moved into a nitrogen box (oxygen free but not water free, owing to the presence of ammonia) for delamination from the growth substrate and transfer to the target substrate. Inside the nitrogen box, the PMMA/graphene/NbSe₂/SiO₂/Si was slowly put into the ammonia solution surface, which enabled immediate detachment of NbSe₂ from SiO₂ by EDL force. The detached PMMA/graphene/NbSe₂ was rinsed with water, transferred from the water surface onto the target substrate and subsequently dried naturally in air for approximately 1 hour. The sample was then taken out of the glovebox for PMMA removal. The as-transferred NbSe₂ showed minimal damage, as confirmed by optical microscopy, Raman spectroscopy and preservation of superconductivity.

Characterization of graphene defects and the impact on long-term air stability

When monolayer graphene is used as a growth template, additional defects can be potentially generated in the graphene layer owing to the high growth temperature. However, our growth time is very short (only 9 min to grow a continuous monolayer NbSe₂), and the defect generation is limited. To quantify the defect density of the graphene template layer, we conducted statistical Raman spectroscopy of graphene before and after NbSe₂ growth. As shown in Supplementary Fig. 5, the post-growth spectra show only a slight increase in the D-to-G peak intensity ratio (I_D/I_G) and a decrease in the 2D-to-G peak intensity ratio (I_{2D}/I_G). This indicates that defect generation is limited and the lattice structure remains largely intact, and the graphene can still

provide sufficient protection against air degradation for its practical applications.

The graphene encapsulation is highly effective for general storage and NbSe₂ can survive most device fabrication steps. According to our long-term air stability measurements, we did not observe any discernible changes in NbSe₂ within the timescale of our experiments (Extended Data Fig. 6). Furthermore, after undergoing typical fabrication steps such as spin-coating, electron-beam lithography, reactive ion etching, and development, and even baking the graphene–NbSe₂ heterostructure at 180 °C in air for 3 min, no degradation was observed (Supplementary Fig. 14). These results indicate that graphene provides effective encapsulation against ambient exposure.

However, when the samples undergo the full set of standard clean-room fabrication processes, we find that NbSe₂ devices can still be damaged during the e-beam evaporation step (Supplementary Fig. 14). This limitation may be attributed to defects that act as vulnerable sites and reduce the robustness of the material under energetic deposition conditions. To address this issue, for microwave device fabrication, we introduce an additional protective layer (such as graphene or MoS₂) on top of the structure to further enhance its robustness during processing. This post-deposition protection can be readily added in ambient conditions without concern for NbSe₂ degradation, as the material is already well protected by the initial graphene encapsulation.

Importantly, the encapsulation-epitaxy growth mechanism, together with pre-growth protective layers such as graphene or hBN, is key to enabling the integration of as-grown monolayer NbSe₂ into superconducting qubit circuits, which had not been feasible previously.

Additional characterization of the graphene template surface

After the growth, we find that the graphene surface is clean with minimal polymer contaminations. The high growth temperature is likely to have annealed away any potential polymer residues during the ramp process. Indeed, according to some studies, polymers may heal some graphene defects at elevated temperatures⁵⁶. Supplementary Fig. 19 show extensive AFM characterizations of graphene/NbSe₂. The graphene surface looks smooth and clean after the CVD growth. The height of NbSe₂ is about 0.63 nm, close to the measurement results on hBN–NbSe₂ shown in Fig. 1g. We note there are some particles on the edge of graphene flakes and wrinkles on the graphene film, which are probably unreacted precursors introduced during the cooling process. These images also show that the nucleation and growth of NbSe₂ initiates from the edges and wrinkles of the template graphene.

Discussion of grain boundaries

Grain boundaries may still be present in our CVD-grown NbSe₂ films. However, the grain boundary density here is expected to be much lower than that reported in previous studies^{18,19}. The grain size of NbSe₂ in this work exceeds 1 μm, whereas previously reported large-scale NbSe₂ films have grain sizes in the nanometre regime, representing a reduction in grain boundary density of several orders of magnitude. Furthermore, as NbSe₂ is epitaxially grown in this work, NbSe₂ under a single graphene (or hBN) domain is constrained to only two possible orientations: either 0° or 60°. Grains of the same orientation merge without forming grain boundaries whereas 60°-rotated grains form mirror twin boundaries, meaning only a subset of grain coalescence events give rise to boundaries. This is in stark contrast to non-epitaxial films, where grains adopt random orientations and nearly every grain coalescence results in a grain boundary.

When polycrystalline CVD graphene is used as the growth template, an additional source of grain boundaries arises when NbSe₂ grains grown beneath adjacent graphene domains of different orientations meet. However, as the CVD graphene domain size (tens to hundreds of micrometres) greatly exceeds the NbSe₂ grain size (about ~1 μm), such graphene-domain-induced NbSe₂ grain boundaries are expected to be sparse. These boundaries are distinct from the

inversion-symmetry-broken induced mirror twin boundaries, which arise when two NbSe₂ domains of opposite orientation meet during epitaxial growth.

Investigation of mirror twin boundaries

Owing to the breaking of in-plane inversion symmetry in monolayer NbSe₂, domains with opposite orientations may potentially form during growth. To examine the possible presence of such domains, we performed extensive HAADF-STEM imaging over multiple regions of the sample (Supplementary Fig. 8). Within the areas we examined, we did not observe domains with opposite orientations. However, we cannot completely rule out their presence outside the regions accessible to our measurements, which is the reason why we describe the film as epitaxially grown rather than framing it to be strictly single crystalline. Another possibility is that such domain boundaries may have been reduced or eliminated during growth through self-alignment mechanisms. Such mechanisms have recently been reported in the CVD growth of 2D TMDs, where initially misoriented domains can merge into a single orientation during growth^{30,57}.

Thermodynamic analysis of the NbSe₂ formation compared with other TMDs

To understand the difficulty of NbSe₂ formation and clarify the role of the 2D–3D hybrid substrate, we conducted thermodynamic analysis. Supplementary Fig. 20 presents DFT-calculated formation energies for various TMDs from the Materials Project database⁵⁸. For sulfur-based TMDs, NbS₂ (–1.349 eV per atom) has a more negative formation energy than MoS₂ (–1.308 eV per atom). However, for NbSe₂ (–0.839 eV per atom), where sulfur is replaced by selenium, the formation energy is significantly less negative than that for MoS₂, indicating that NbSe₂ formation is thermodynamically less favourable. WTe₂ (–0.106 eV per atom) has the least negative formation energy among all TMDs studied, making it the hardest to form. These trends are physically meaningful, reflecting the monotonically decreasing electronegativity of the chalcogen atoms from S to Se to Te, which weakens the metal–chalcogen bond accordingly.

In this work, we have demonstrated growth on a range of TMDs including NbSe₂, NbS₂, VS₂, VSe₂ and WTe₂. These materials differ substantially in their thermodynamic properties, as they span sulfur-based, selenium-based and tellurium-based systems and cover a wide range of formation energies (Supplementary Fig. 20). As encapsulation epitaxy succeeds across all of them on the hybrid substrate, whereas monolayer growth fails on either graphene or SiO₂ alone, this strongly supports the conclusion that the hybrid substrate, not the TMD material-specific properties, is the dominant factor for the growth.

Further discussion about superconducting edge contact

Figure 3g shows the measurement configuration where we measure the resistance of the electrodes and the device without incorporating line resistance. The differential resistance (dI/dI) of the two terminals goes to zero (within instrument noise) below the critical current transition, which indicates that the contact resistance (or any residual resistance at the interface of Al–NbSe₂) is very low when the film is superconducting. The line cut from Fig. 3h at zero bias current shows that the resistance becomes zero (Fig. 3i) when 1L–NbSe₂ is superconducting, which confirms superconducting transparent contact between the Al electrodes and 1L–NbSe₂.

Comparison between CVD 1L–NbSe₂ with conventional high-kinetic-inductance materials

We emphasize that the primary advance of this work is not to outperform existing materials solely in terms of L_k or Q_i , but rather to demonstrate a platform for studying high-impedance crystalline superconductors, in contrast to the amorphous materials more frequently used in superconducting circuits.

Conventional high- L_k superconductors, including gr-Al, NbN and TiN, rely largely on disorder engineering to obtain large kinetic inductance. Disorder increases the normal-state sheet resistance and suppresses the superfluid stiffness, thereby enhancing L_k . Although these materials have shown promising performance in superconducting quantum circuits, their disorder-assisted enhancement of L_k can impose a fundamental limit on the internal quality factor Q_i (ref. 59). Supplementary Fig. 21 summarizes this trend for high-impedance superconductors, including gr-Al, NbN, Nb_xSi_{1–x} and TiN (refs. 45,60), showing that higher kinetic inductance is often accompanied by reduced Q_i .

Our work explores crystalline, atomically thin NbSe₂ as an alternative high- L_k platform, where large kinetic inductance arises from reduced dimensionality and low superfluid density rather than strong structural disorder. As shown in Fig. 4i, the CVD NbSe₂ offer the thinnest body thickness (about 1 nm) and largest grain size (>1 μm) among high- L_k superconducting materials, while having kinetic inductance comparable to gr-Al or NbN thin films (Supplementary Fig. 21) in the low-thickness (<10 nm) regime. The measured $Q_i \approx 3.15 \times 10^4$ is lower than the best optimized conventional films, but it already falls within the range of several reported high- L_k superconducting materials. Importantly, we note that the reported Q_i in this work corresponds to the entire NbSe₂-terminated Al resonator, which should not be deemed as the intrinsic quality factor of NbSe₂ itself. A dedicated device geometry isolating the intrinsic NbSe₂ response via an all-NbSe₂ resonator represents a natural next step, which is beyond the scope of this work and will be pursued in the future.

The CVD NbSe₂ in this work represents a 2D superconductor-based platform that simultaneously offers: (1) atomically thin thickness, (2) high crystallinity with large grain size, (3) high kinetic inductance and (4) an atomically flat surface. This combination of properties is inaccessible to conventional amorphous thin films, and positions CVD NbSe₂ as a complementary and potentially extensible pathway for superconducting quantum circuits, rather than a direct replacement of existing high-impedance materials.

Material characterization

Raman spectra were measured using WITec alpha300 apyron Confocal Raman with an excitation wavelength of 532 nm and a grating of either 300 or 1,800 grooves per mm. AFM images were taken using Asylum Cypher VRS in tapping mode. SEM images were obtained using a high-resolution SEM (ZEISS Merlin; 3 kV) with an in-lens detector.

XPS spectra were acquired using a PHI VersaProbe II system equipped with a monochromatic Al K α source (1,486.6 eV) and a 200-μm spot size. The measurements were performed at a gun power of 50 W and an operation voltage of 15 kV. To neutralize surface charging during data acquisition, the sample was flooded using electron and argon ion guns. All spectra were calibrated against the C 1s peak at 284.8 eV.

TEM samples used for cross-sectional STEM images were prepared by Helios 460F1 Dual Beam milling. The lamella was thinned to 50–100 nm thickness at an accelerating voltage of 30 kV with a current decreasing from 0.78 nA to 40 pA, followed by a fine polish at lower voltage with a current of 21 pA to clean the sample surface.

The cross-sectional HAADF-STEM images were captured with a Thermo Fisher Scientific Titan Themis 80–300 aberration-corrected S/TEM at 300 kV. The HAADF-STEM images were post-processed using DigitalMicrograph software (Gatan). To filter the image noise, the lattice contrast of HAADF-STEM images was enhanced by Gaussian smooth and band-pass filter. Energy-dispersive X-ray spectroscopy mappings were acquired with a Super-X detector (4 silicon drift detectors) at 300 kV. Nb and Se signals are selected from the L line to increase the intensity.

The atomic lattice imaging in Fig. 2a was conducted using a high-resolution STEM (Titan Themis Z G3 Cs-corrected S/TEM; 60 kV). The SAED images were obtained using multipurpose transmission electron microscope (FEI Tecnai G2 Spirit TWIN; 120 kV). The graphene/NbSe₂ free-standing samples were prepared by transferring

as-grown graphene/NbSe₂ from SiO₂/Si substrate onto a TEM grid, which is enabled by electrostatic repulsion force in ammonia solution (EDL transfer)³⁴.

CDW characterization

The optical detection of CDW in NbSe₂ was conducted using temperature-dependent Raman spectroscopy measurements, which were performed on a Horiba Labram HR Evolution spectrometer equipped with a liquid-nitrogen-cooled charge-coupled device and 600 l mm⁻¹ and 1,800 l mm⁻¹ spectrometer gratings for medium- and high-resolution spectra, respectively. All measurements were performed using an excitation wavelength of $\lambda = 532$ nm with an incident power of approximately 100 μ W on the sample. Bragg grating notch filters were used to filter the elastic line, permitting measurements down to 5 cm⁻¹ energy transfer. The incident excitation laser was linearly polarized and passed through an achromatic half-wave plate (Thorlabs) placed before a $\times 50$ objective (Olympus) to vary the angle of incident polarization on the sample, with a $1/e^2$ focus spot diameter of approximately 4 μ m. The scattered light polarization was resolved into linear parallel (X') and perpendicular (Y') components using a linear polarizer placed before the entrance slit of the spectrometer. Variable temperature measurements were performed using a Montana Instruments S50 closed-cycle optical cryostat.

The CDW transition temperature, T_{CDW} , was determined from the intensity of the amplitude mode I_A , which is reported in the literature⁷. Raman spectra $S_i(\omega)$ collected at temperature T were first normalized by a reference spectrum $S_0(\omega)$ measured at a temperature $T_0 > T_{\text{CDW}}$. Subsequently, 1 was subtracted from this ratio to eliminate temperature-independent contributions unrelated to the CDW transition. The integrated intensity of the amplitude mode was then extracted as a function of temperature (Fig. 2e). The transition temperature T_{CDW} was identified as the point where I_A crosses zero. The solid curves in Fig. 2e and Extended Data Fig. 10c are fits based on mean-field theory, used as guides to the eye (the equation used here is provided in the Supplementary Information of ref. 7).

Device fabrication and characterization

After the 1L-NbSe₂-based heterostructure had been transferred to the target substrate, it went through standard device fabrication process in ambient conditions. It was first patterned and etched into a designed shape using electron-beam lithography and reactive ion etching. After removing the resist, another electron-beam lithography step was conducted to define metal contact patterns. After developing the pattern, the 1L-NbSe₂-based heterostructure went through another etching to expose the edges, followed by 160 nm Al deposition. In situ ion milling was not required in this process as the graphene serves as an ideal layer for constructing the edge contact, which overcomes the challenge of oxidation of exposed NbSe₂ edges. After completing the metal deposition and lift-off processes, the sample was mounted and cooled to a base temperature of 10 mK using a dilution refrigerator (Bluefors LD400).

During most of the device fabrication process, the NbSe₂ can be handled, transported and stored in ambient conditions without a post-growth encapsulation layer—a unique advantage provided by encapsulation epitaxy. For devices that involve the step of metal deposition, an additional post-growth protection layer was applied to ensure that NbSe₂ is not damaged by the high-energy metal atoms (we use monolayer MoS₂ as it is readily accessible to us). Supplementary Fig. 14 shows a comprehensive comparison of graphene/NbSe₂ with and without an extra capping layer during the fabrication steps.

For the pseudo-four-terminal measurement in Fig. 3g–i, the measured resistance represents the total series resistance of 1L-NbSe₂, the NbSe₂–Al interface and Al leads between the interface and voltage probe. It drops to zero at low temperature, confirming that the superconducting edge contact is achieved with zero resistance between 1L-NbSe₂ and a conventional 3D superconductor (Al).

Transport measurement set-up

Each direct-current line inside the dilution refrigerator is equipped with cryogenic filters at 4 K stage (RC π -filter with a cut-off frequency of 53 kHz) and a low-pass filter (RC-RF filter from QDevil) at the mixing chamber plate to get rid of high-frequency noise. Supplementary Fig. 15 shows the wiring diagram for direct-current transport measurement from room temperature to the device. For measuring the resistance of our devices as a function temperature, an alternating-current signal was provided by an SRS 860 lock-in amplifier and was applied through the BNC cables. An SRS 560 low-noise voltage amplifier ($\times 1,000$) was used to amplify the signal, which was then read by the lock-in amplifier. For differential measurements, both alternating-current and direct-current signals were applied with the current bias scheme using the lock-in for small alternating-current current and a QDAC, which is a multi-channel, ultralow-noise voltage source/digital-to-analogue converter (from QDevil), as a direct-current voltage source. Two resistors with resistance much larger compared with the device's resistance were used to create a current bias from the voltage sources: $R_{\text{AC}} = 10$ M Ω and $R_{\text{DC}} = 100$ k Ω . The amplified voltages were measured by the lock-in input channel and Keithley 2100 digital multimeter (DMM). All instruments mentioned here and the fridge share a common electrical ground.

Microwave measurement set-up

The microwave measurement set-up is shown in Supplementary Fig. 16, where the device is mounted at the mixing chamber stage with via coax cables. The input line was attenuated by 20 dB at the 4 K stage, 10 dB at the still stage (1 K), and 40 dB at the mixing chamber stage to minimize thermal noise from higher-temperature stages. The output line was filtered with 3-GHz high-pass and 12-GHz low-pass filters at the mixing chamber stage. We had a Josephson travelling wave parametric amplifier at the mixing chamber stage, and low-noise high-electron-mobility transistor amplifiers at the 4 K stage and room-temperature stages to amplify the output signal. A vector network analyser from Keysight (N5232A) was used to send and receive microwave signals.

Discussion of residual resistance

The residual resistance observed in Fig. 3d is 150 Ω and if we convert it to sheet resistance (our device has width 5 μ m and length 3 μ m), it becomes approximately 250 $\Omega \square^{-1}$, which is comparable to values typically reported for exfoliated NbSe₂ fully encapsulated by hBN (approximately 240 $\Omega \square^{-1}$)⁶¹ and graphene (top) and a hBN (bottom) encapsulated device (450 $\Omega \square^{-1}$)⁶. Given that our sheet resistance closely matches that of high-quality exfoliated NbSe₂, we think that the excessive defect scattering is not dominant in the transport in our devices. Although some residual resistance is expected from finite disorder, and domain boundaries, its magnitude suggests that defect-induced scattering is not unusually strong. Regarding device performance, any remaining defect scattering would primarily affect mobility and normal-state dissipation. However, as our measured resistance values remain within the range of high-quality NbSe₂ devices, we expect the impact on overall device operation to be limited.

Data availability

All data needed to evaluate the conclusions herein are present in the paper.

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Author contributions J.K., J.Ĵ-j.W. and W.D.O. supervised the project. X.Z. and S.Z. conceived of the experiments. X.Z. developed the encapsulation-epitaxy growth method. S.Z. performed the superconducting device fabrication and measurements. X.Z., K.Z. and Z.W. carried out the material synthesis and characterization. C.A.O. and L.G.P.M. conducted low-temperature Raman measurements, under the supervision of R.C. H.X. performed DFT calculations, under the supervision of J.L. X.L. conducted the cross-sectional STEM imaging, under the supervision of Y.H. F.L. conducted additional cross-sectional STEM imaging, under the supervision of C.S. Z.W. and Y.Z. provided the CVD graphene films. S.L. conducted the cryogenic dry pick-up tests, under the supervision of P.K. T.Z. contributed to the graphic illustration. T.H.Y. and S.P. contributed to device fabrication. J.W. and Z.H. contributed to the data analysis. S.V., K.T. and S.M. prepared the Nb films. X.Z., S.Z., J.Ĵ-j.W. and J.K. wrote the paper; all authors read and revised the paper.

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Competing interests X.Z. and J.K. are co-inventors on a patent application (provisional filing number number 63/941,016) related to the research presented in this paper.

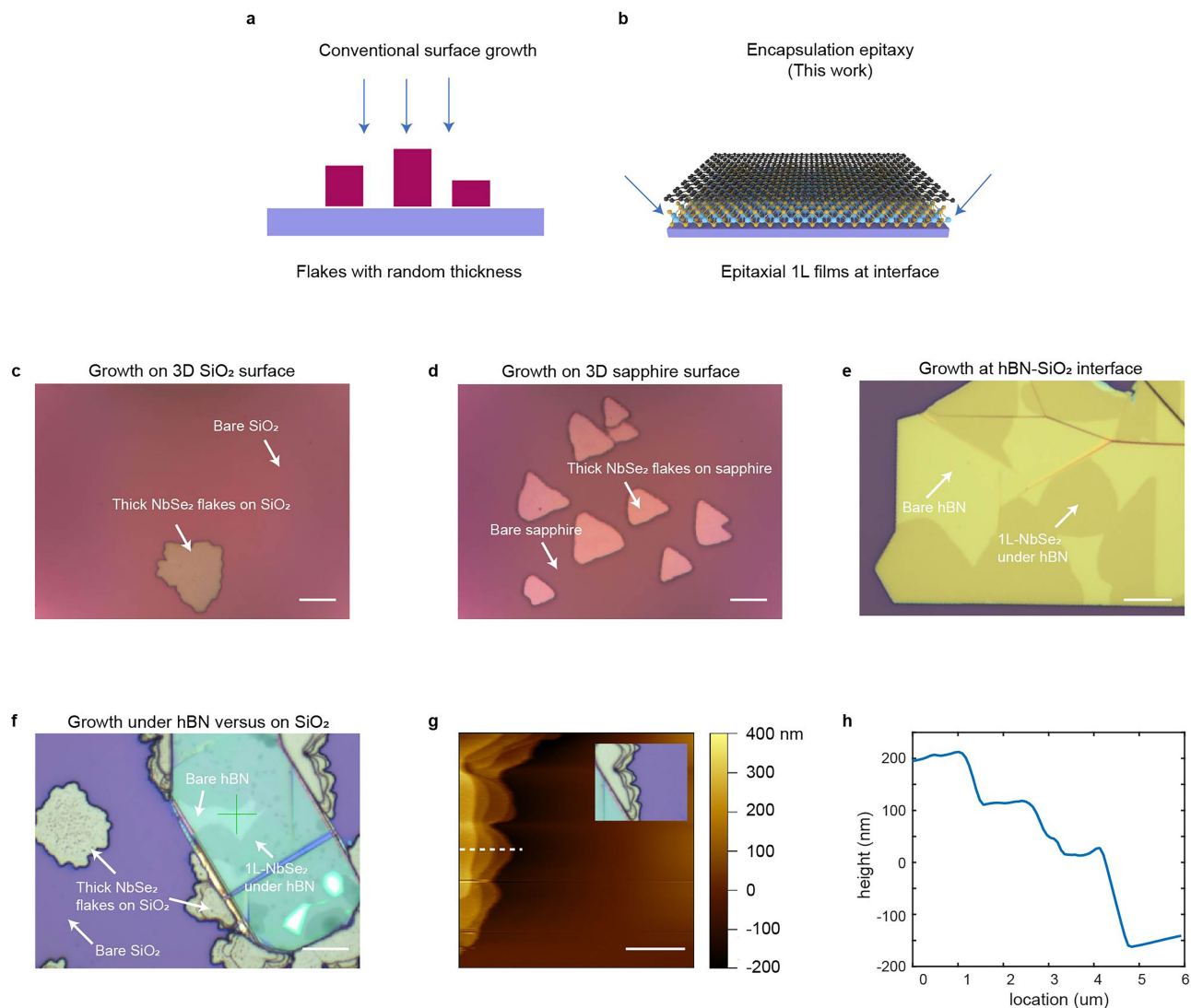
Additional information

Supplementary information The online version contains supplementary material available at <https://doi.org/10.1038/s41586-026-10865-1>.

Correspondence and requests for materials should be addressed to William D. Oliver, Joel Ĵ-j. Wang or Jing Kong.

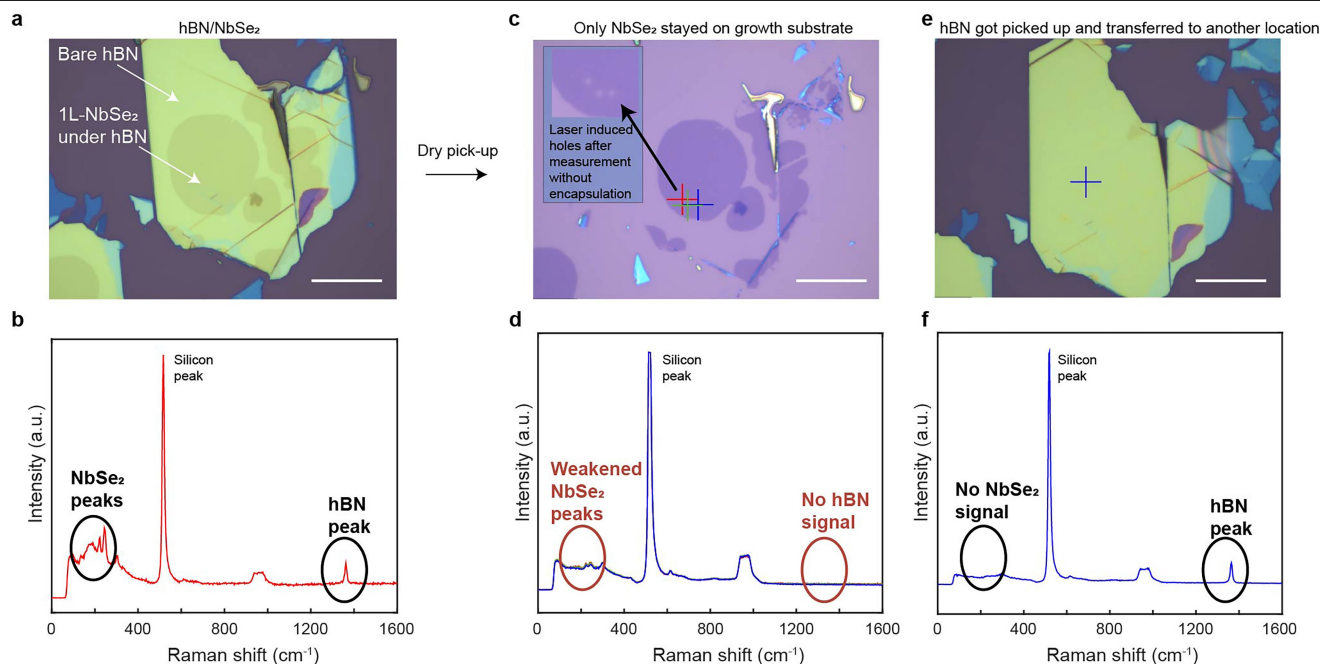
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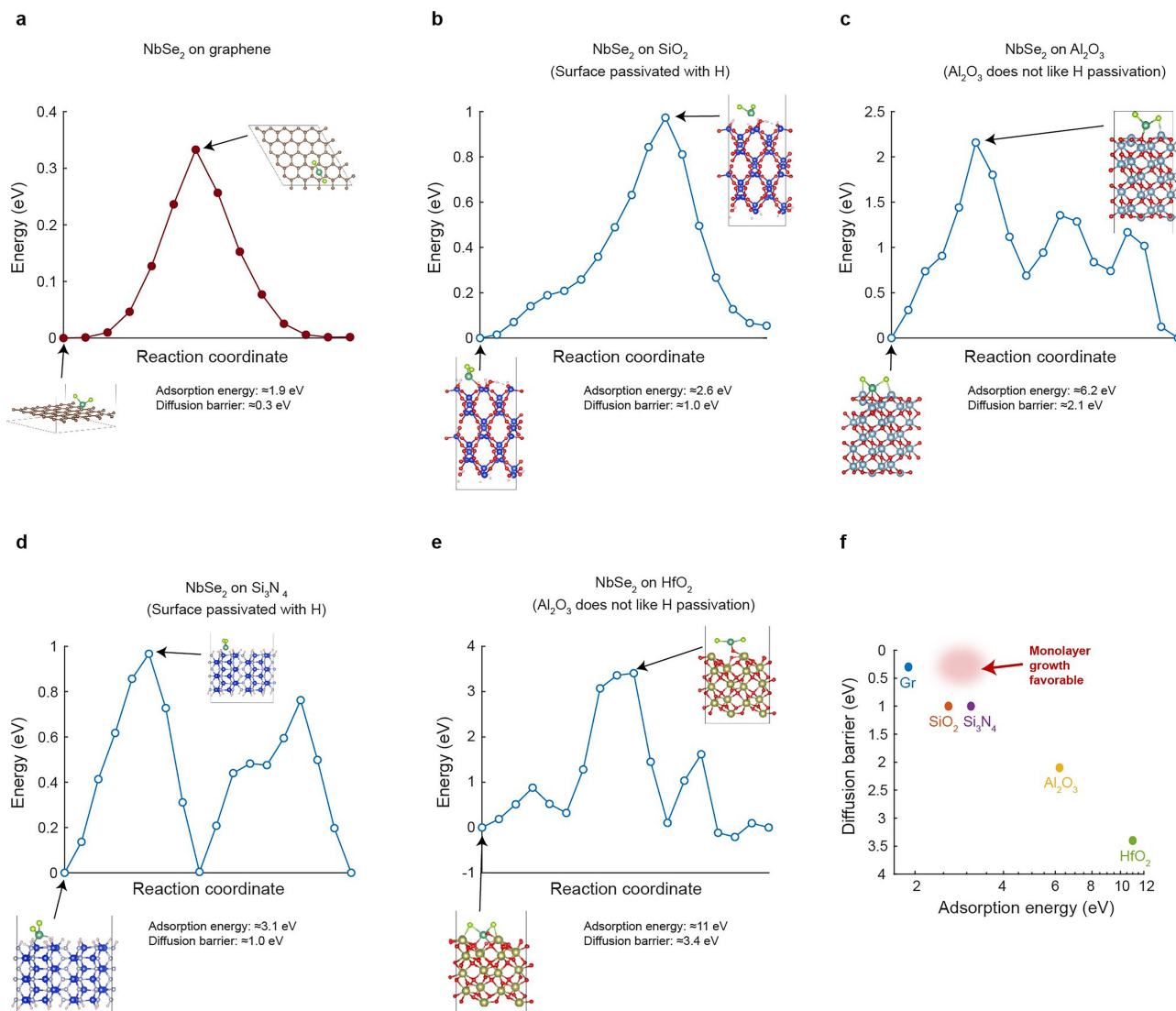
Extended Data Fig. 1 | Comparison between conventional surface growth versus encapsulation epitaxy. **a-b**, Schematic showing the difference between (a) conventional surface growth on 3D surface and (b) encapsulation epitaxy at 2D-3D interface. **c-e**, Optical image showing the growth of NbSe₂ on (c) SiO₂ substrate surface, (d) sapphire substrate surface, and (e) exfoliated hBN flakes on SiO₂. Scale bars, 10 μ m. **f**, Additional example showing the difference between SiO₂ surface and hBN-SiO₂ interface. This growth utilizes a high Nb

precursor concentration, enabling the simultaneous observation of island growth on SiO₂ surface (conventional surface growth) and 1L-NbSe₂ growth at hBN-SiO₂ interface (encapsulation epitaxy). Scale bar, 10 μ m. **g**, AFM image measuring the thickness of the thick NbSe₂ flake on SiO₂ surface. Inset, optical image showing the measured location. Scale bar, 1 μ m. **h**, Corresponding height profile of the AFM image shown in panel **g** (dashed white line).

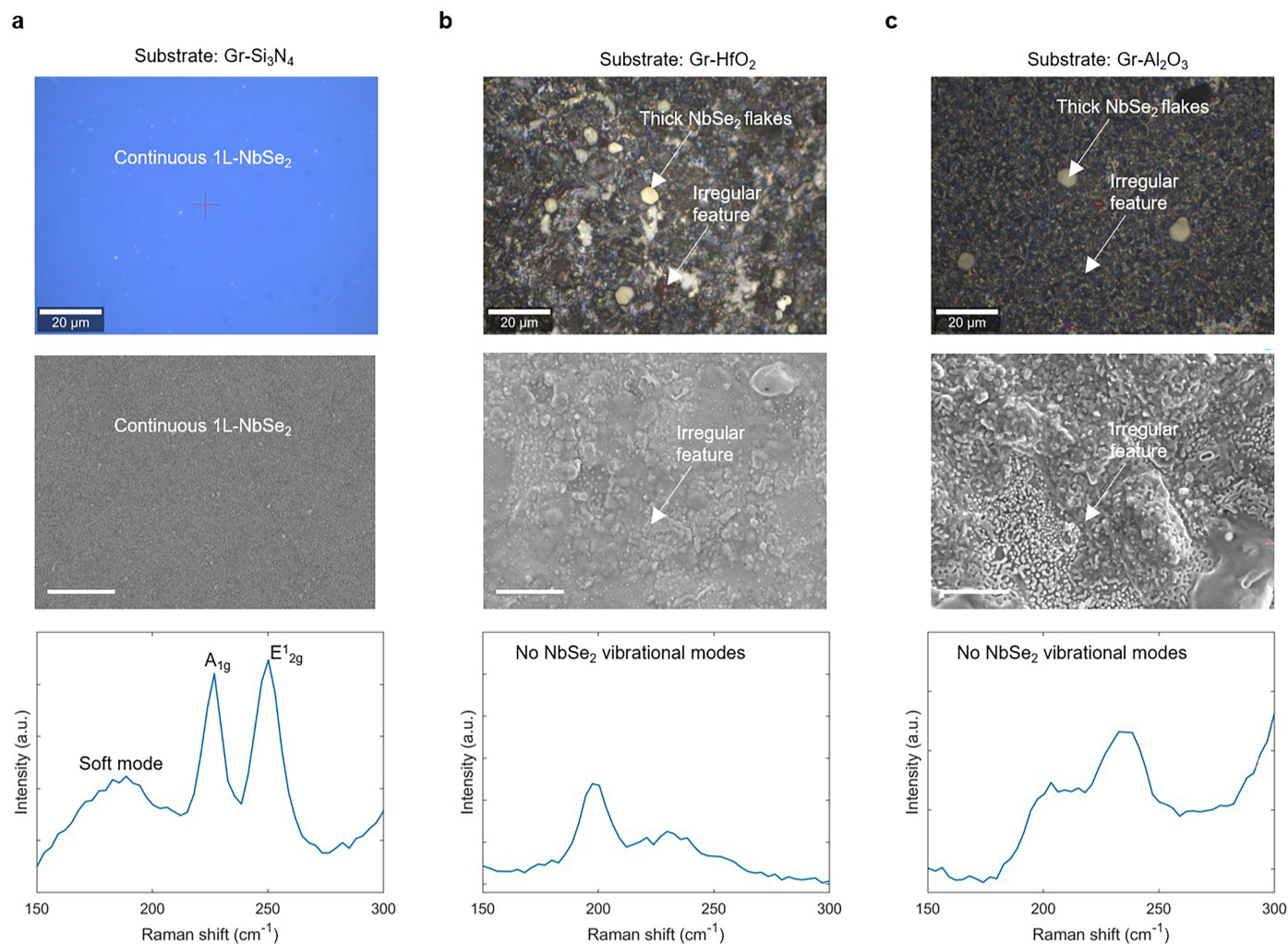


Extended Data Fig. 2 | Cryogenic dry pick-up test demonstrating that NbSe₂ growth occurred exclusively at 2D-3D interface, with no growth observed on the exposed 2D surface. **a**, Optical image showing the 1L-NbSe₂ grown by encapsulation epitaxy at the hBN/SiO₂ interface. **b**, Raman spectra of as-grown hBN/NbSe₂ from panel **a**. **c**, Optical image of the same sample shown in panel **a** after cryogenic dry pick-up (see Methods). The NbSe₂ remains on the growth substrate while the hBN flakes were lifted off, confirming that NbSe₂ was grown beneath hBN at the 2D-3D interface. After removal of the hBN layer, further laser exposure will cause damage to the unprotected NbSe₂ surface. The inset shows the optical image of laser-induced holes on exposed NbSe₂ surface,

underscoring the importance of 2D encapsulation. **d**, Raman spectra of the exposed NbSe₂ after removal of the top hBN layers. The NbSe₂ peaks are markedly weakened due to the absence of hBN encapsulation during laser exposure. The three spectra correspond to measurements taken at the positions indicated by the three crosses in panel **c**. **e**, Optical image of the hBN flake being picked up in **a** and transferred onto another new substrate. **f**, Raman spectra measuring the cross-marked location in panel **e**. No NbSe₂ signal was detected, together with the uniform color contrast under the optical microscope image, indicating no growth happened on top of the hBN surface. Scale bars, 20 μm.



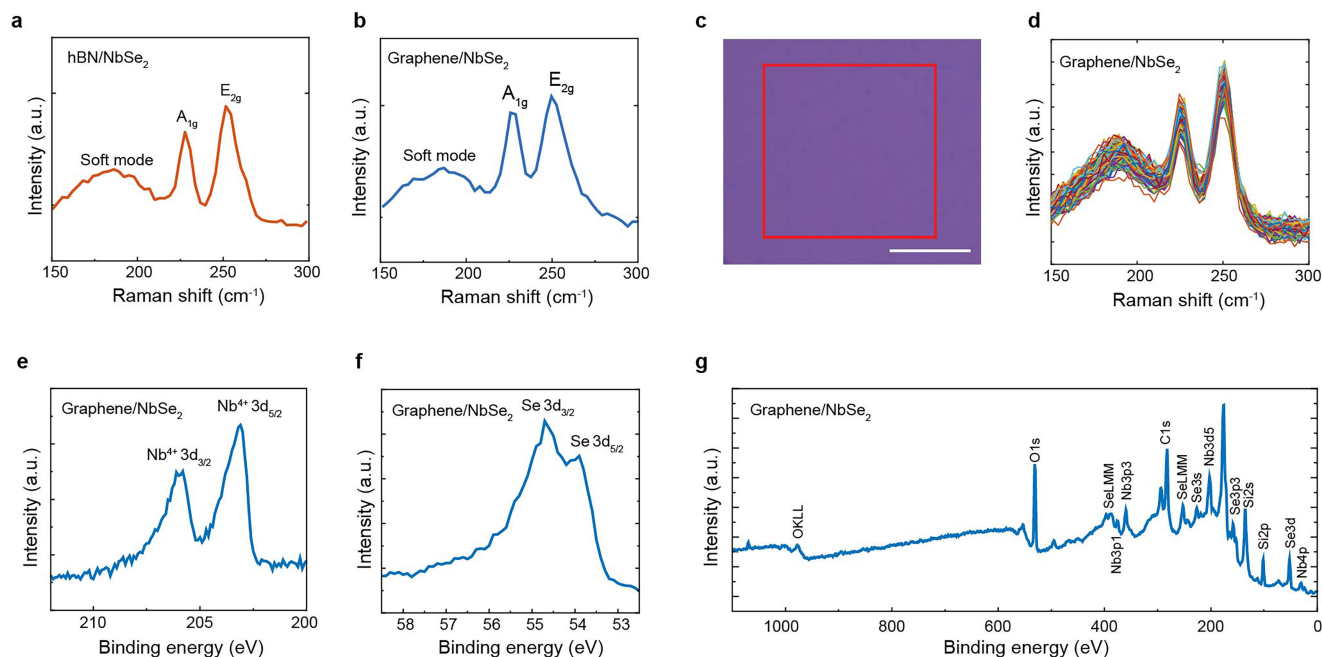
Extended Data Fig. 3 | More comprehensive DFT calculations. a-e, DFT-calculated energy along the diffusion path of NbSe₂ on (a) graphene, (b) SiO₂, (c) Al₂O₃, (d) Si₃N₄, and (e) HfO₂. **f**, Calculated diffusion energy vs. adsorption energy of NbSe₂ on various surfaces.



Extended Data Fig. 4 | NbSe₂ growth test on different substrates.

a-c, Optical images, SEM images and Raman spectra showing the NbSe₂ growth result on (a) Graphene-Si₃N₄ substrate (b) Graphene-HfO₂ substrate and (c) Graphene-Al₂O₃ substrate. The continuous 1L-NbSe₂ is successfully grown at graphene-Si₃N₄ interface. But in the case of HfO₂ and Al₂O₃, only thick NbSe₂

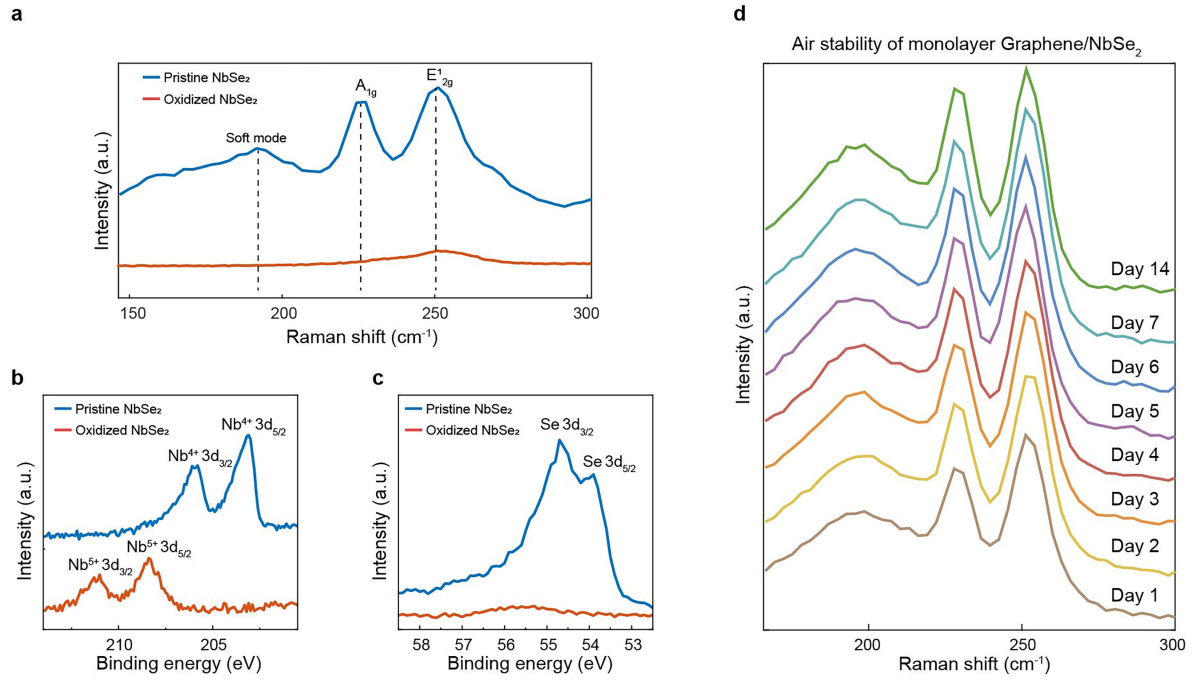
flakes and irregular black features (no NbSe₂ Raman signal) were formed. The Si₃N₄ was grown by PECVD, and HfO₂ and Al₂O₃ were grown by ALD. The graphene films were then transferred on top of them. Scale bars for optical images, 20 μ m. Scale bars for SEM images, 2 μ m.



Extended Data Fig. 5 | Raman and XPS characterization of as-grown NbSe₂.

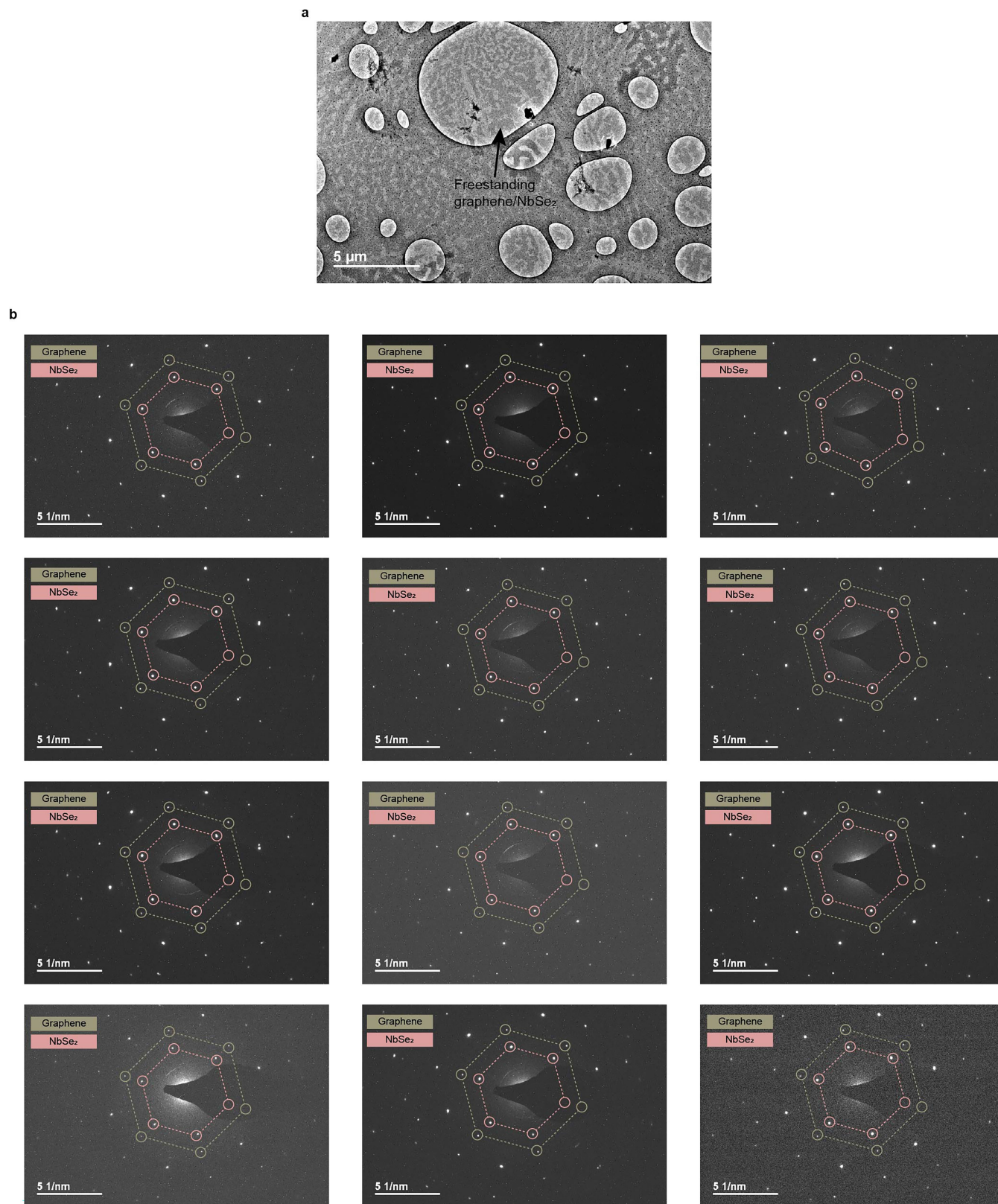
a, Raman spectrum of as-grown hBN/NbSe₂. **b**, Raman spectrum of as-grown graphene/NbSe₂. **c**, Optical image showing the area of the graphene/NbSe₂ sample used for Raman mapping in Fig. 1i–j. Scale bar: 10 μ m. **d**, Raman spectra collected from the mapped area shown in **c**. **e**, XPS spectrum of as-grown graphene/NbSe₂ showing Nb 3d peaks. Two peaks at 203.4 eV and 206.3 eV

correspond to Nb⁴⁺ 3d_{5/2} and 3d_{3/2} binding energies, consistent with literature values. **f**, XPS spectrum of as-grown graphene/NbSe₂ showing Se 3d peaks. Peaks at 53.2 eV and 54.2 eV correspond to Se 3d_{5/2} and 3d_{3/2} binding energies, also in good agreement with literature¹⁶. **g**, XPS Survey spectrum of the graphene/NbSe₂ sample.



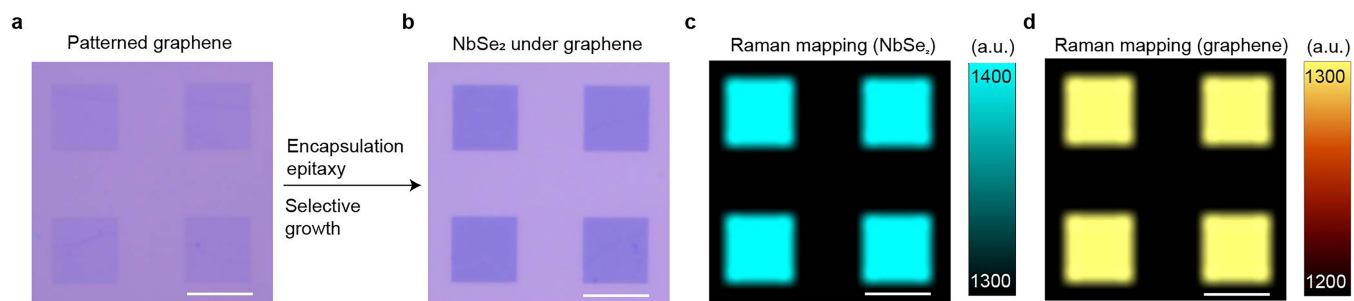
Extended Data Fig. 6 | Air-stability characterization of 1L-NbSe₂ grown by encapsulation epitaxy. **a**, Raman spectra comparing pristine NbSe₂ with intentionally oxidized NbSe₂. The disappearance of NbSe₂ vibrational modes upon oxidation highlights Raman spectroscopy as an effective tool for rapid air-stability assessment. **b-c**, X-ray photoelectron spectroscopy (XPS) of the same pristine and oxidized NbSe₂ samples shown in **a**. Upon oxidation, the Nb valence changes from 4⁺ to 5⁺, and the Se signal disappears, revealing that the oxidized NbSe₂ is composed of Nb₂O₅. Noted this Nb₂O₅ might be amorphous

considering no Raman characteristic peak is detected. **d**, Raman spectra of a 1L-graphene/NbSe₂ sample grown by encapsulation epitaxy during continuous exposure to ambient air. Minimal oxidation is observed, highlighting the high air-stability of NbSe₂ in this work. This also provides indirect evidence that NbSe₂ is grown beneath the 2D encapsulation layer, as the unencapsulated monolayer NbSe₂ is expected to rapidly oxidize in air and exhibit significantly weakened or diminished Raman signals (See Extended Data Fig. 2).



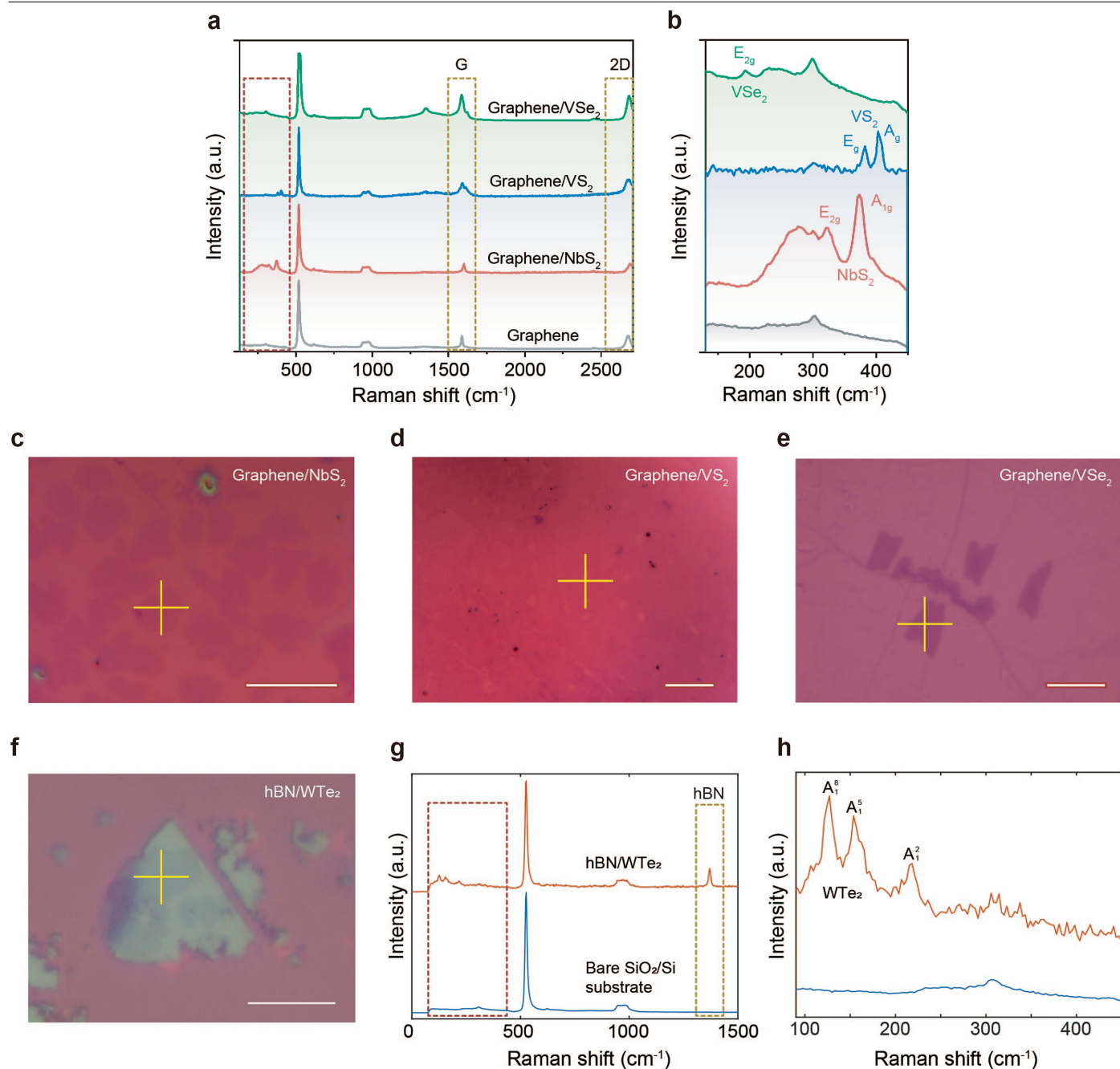
Extended Data Fig. 7 | More SAED characterizations of graphene/NbSe₂.
a, Planar-view SEM image of graphene/NbSe₂ on TEM grid, showing one of the regions that SAED was taken. Graphene is a continuous monolayer while the

NbSe₂ are partially grown flakes. The holes are free-standing areas. Scale bar, 5 μm. **b**, SAED measurements of graphene/NbSe₂ taken at random locations across a 200 × 200 μm² area. Scale bar, 5 nm⁻¹.



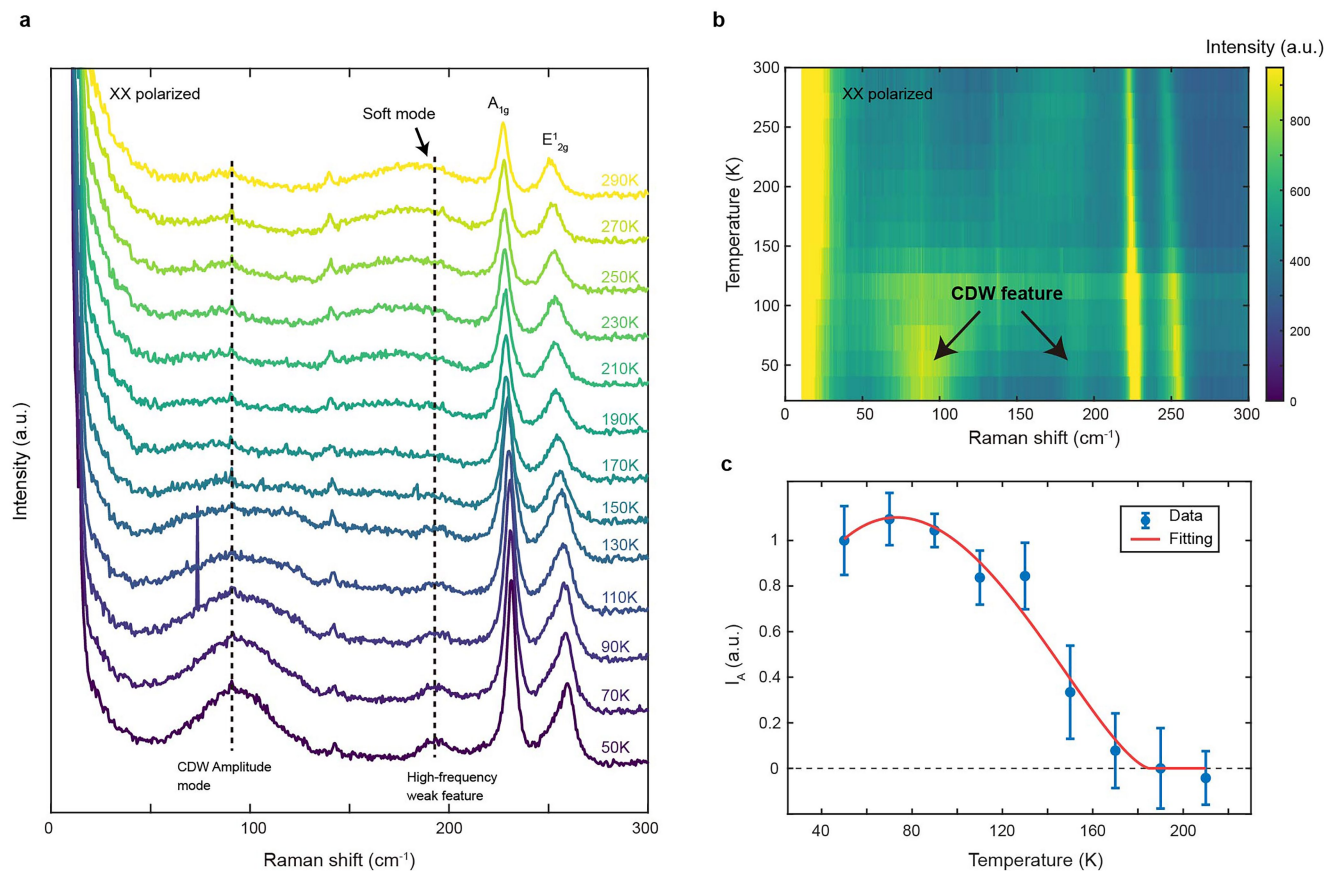
Extended Data Fig. 8 | Selective-area growth of NbSe₂ utilizing patterned graphene. **a**, Optical image of patterned graphene on SiO₂ substrate. **b**, Optical image of NbSe₂ grown underneath the patterned graphene shown in **a**. **c-d**,

Raman intensity mapping images of (c) NbSe₂ E_{2g} peak and (d) graphene 2D peak. The Raman mapping was taken on the optical image shown in **b**. Scale bars, 10 μm.



Extended Data Fig. 9 | Growth of other high-melting-point TMDs by encapsulation epitaxy. **a**, Raman spectra of as-grown graphene/NbS₂, graphene/VS₂, and graphene/VSe₂, with the spectrum of bare graphene included as a reference. Raman peaks located around 1590 and 2680 cm⁻¹ correspond to G and 2D vibration modes of graphene. Raman peaks at 300 and 520 cm⁻¹ arise from the silicon substrate. **b**, Zoomed-in view of panel **a** in the range 130–450 cm⁻¹ (the region within red dotted rectangle). The Raman peaks of graphene/NbS₂ are observed at 324 and 375 cm⁻¹, corresponding to the E_{2g} and A_{1g} modes of H-phase NbS₂⁶². The Raman peaks of graphene/VS₂ appear at 384 and 404 cm⁻¹, corresponding to the E_g and A_g modes of H-phase VS₂⁶³. The

Raman peak of graphene/VSe₂ appears at 193 cm⁻¹, corresponding to the E_{2g} vibrational mode of H-phase VSe₂⁶⁴. **c–e**, Optical images of as-grown graphene/NbS₂, graphene/VS₂ and graphene/VSe₂, respectively. The markers on the optical images indicate the measurement positions for the Raman spectra in panel **a**. Scale bars, 10 μm. **f**, Optical image of as-grown hBN/WTe₂. The marker indicates the Raman measurement location. Scale bar, 10 μm. **g**, Raman spectra of as-grown hBN/WTe₂, with bare SiO₂/Si substrate included as a reference. **h**, Zoomed-in view of panel **g** in the range 90–450 cm⁻¹ (the region within red dotted rectangle). The Raman peaks of hBN/WTe₂ are observed at 127, 157, and 218 cm⁻¹, corresponding to the A₁^g, A₁^s and A₂^g modes of Td-WTe₂⁶⁵.



Extended Data Fig. 10 | CDW measurement of hBN/NbSe₂. **a**, Temperature dependent Raman measurement (range showing vibration modes of NbSe₂) with XX polarized laser on hBN/1L-NbSe₂ heterostructure synthesized by encapsulation epitaxy. The curves are vertically offset for clarity. **b**, Color plot of the temperature dependent Raman measurement shown in **a**. **c**, Temperature

dependence of the CDW amplitude mode intensity (I_A) for hBN/1L-NbSe₂ heterostructure. The error bars are calculated based on the standard deviation of the Raman scattering intensity around the amplitude mode peak and the spectral width employed for integrating the mode. Solid lines are fits to mean field theory⁷.

Supplementary information

Encapsulation epitaxy of air-stable 2D superconductors for quantum circuits

In the format provided by the
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Supplementary Information for

Encapsulation epitaxy of air-stable 2D superconductors for quantum circuits

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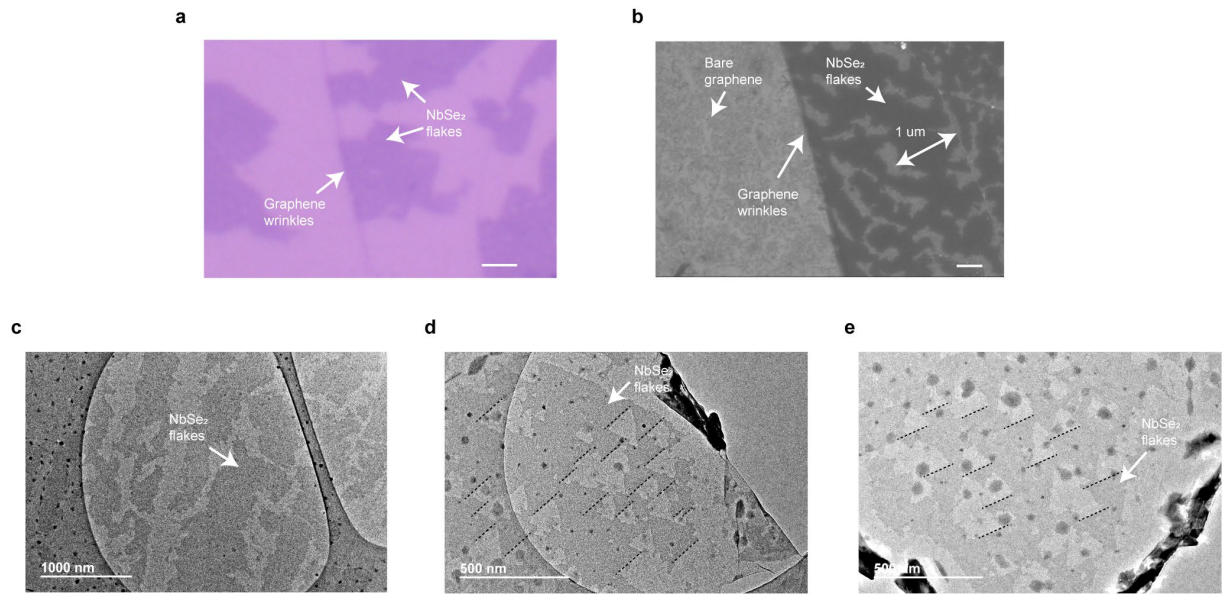
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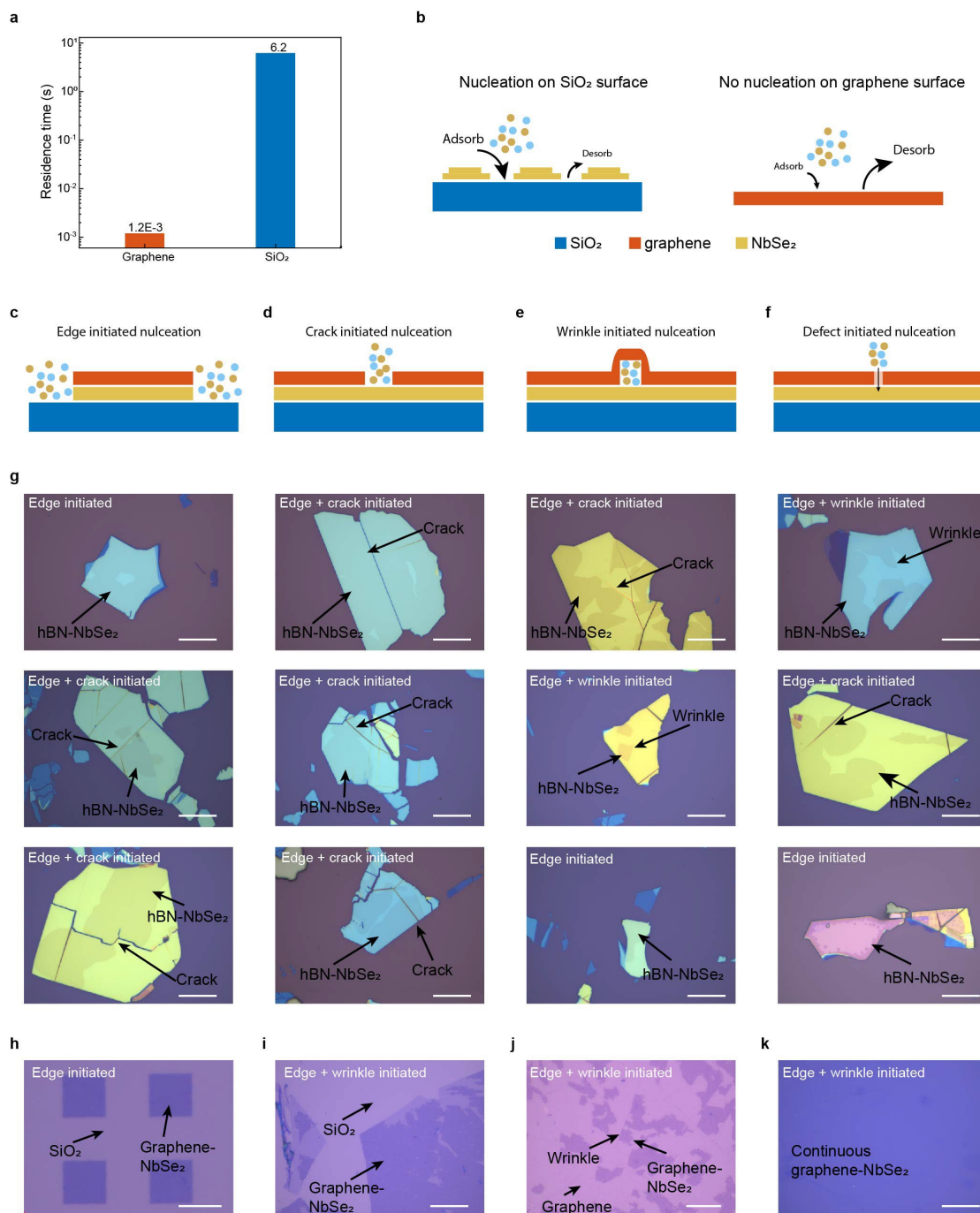
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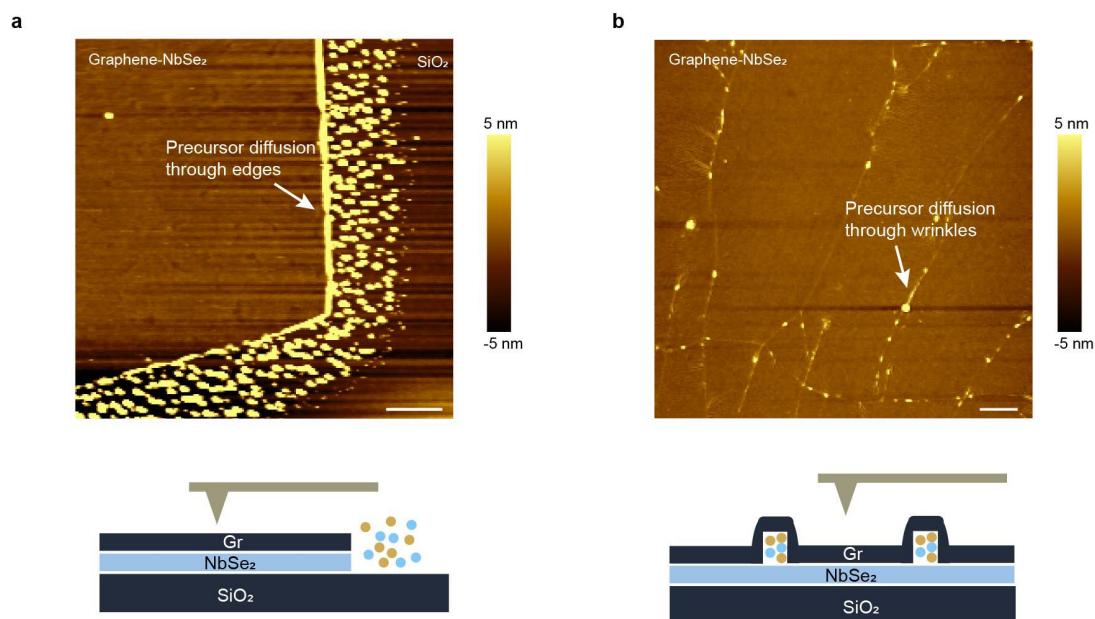
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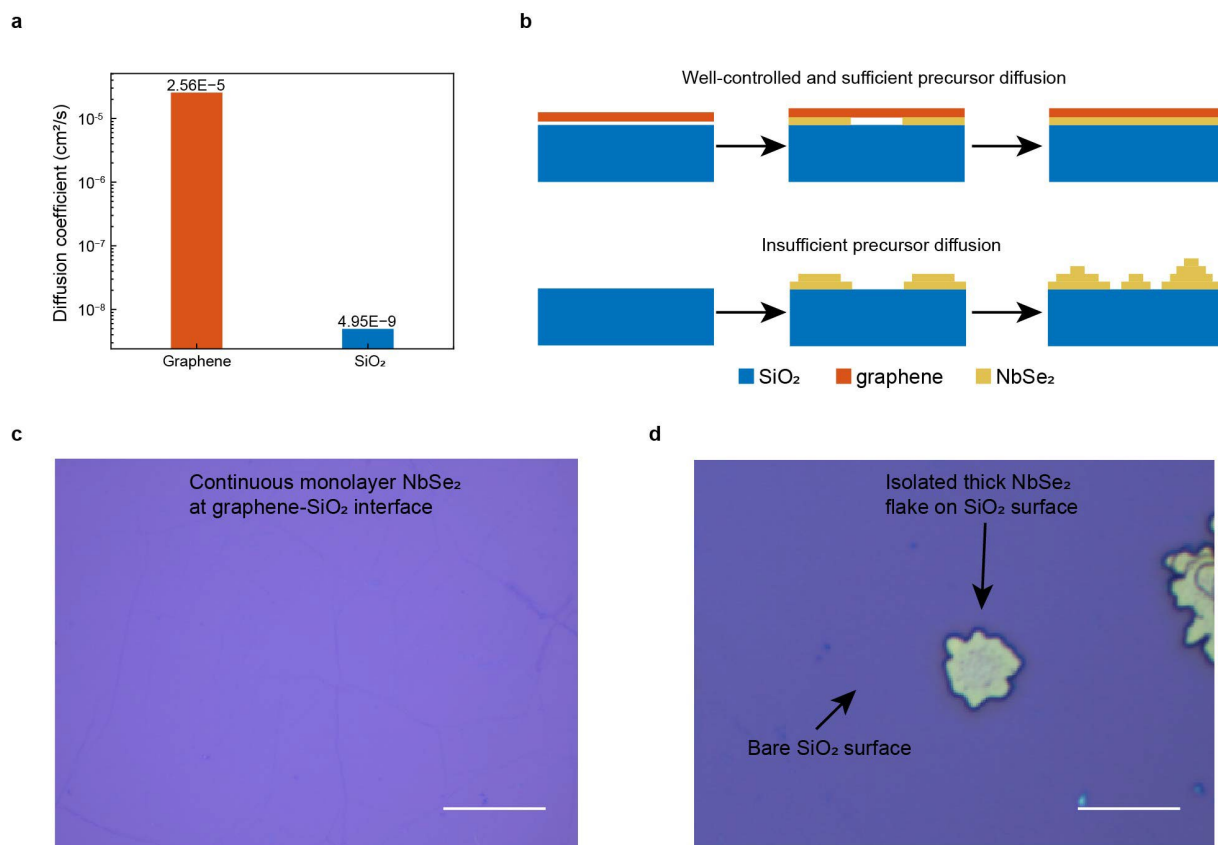




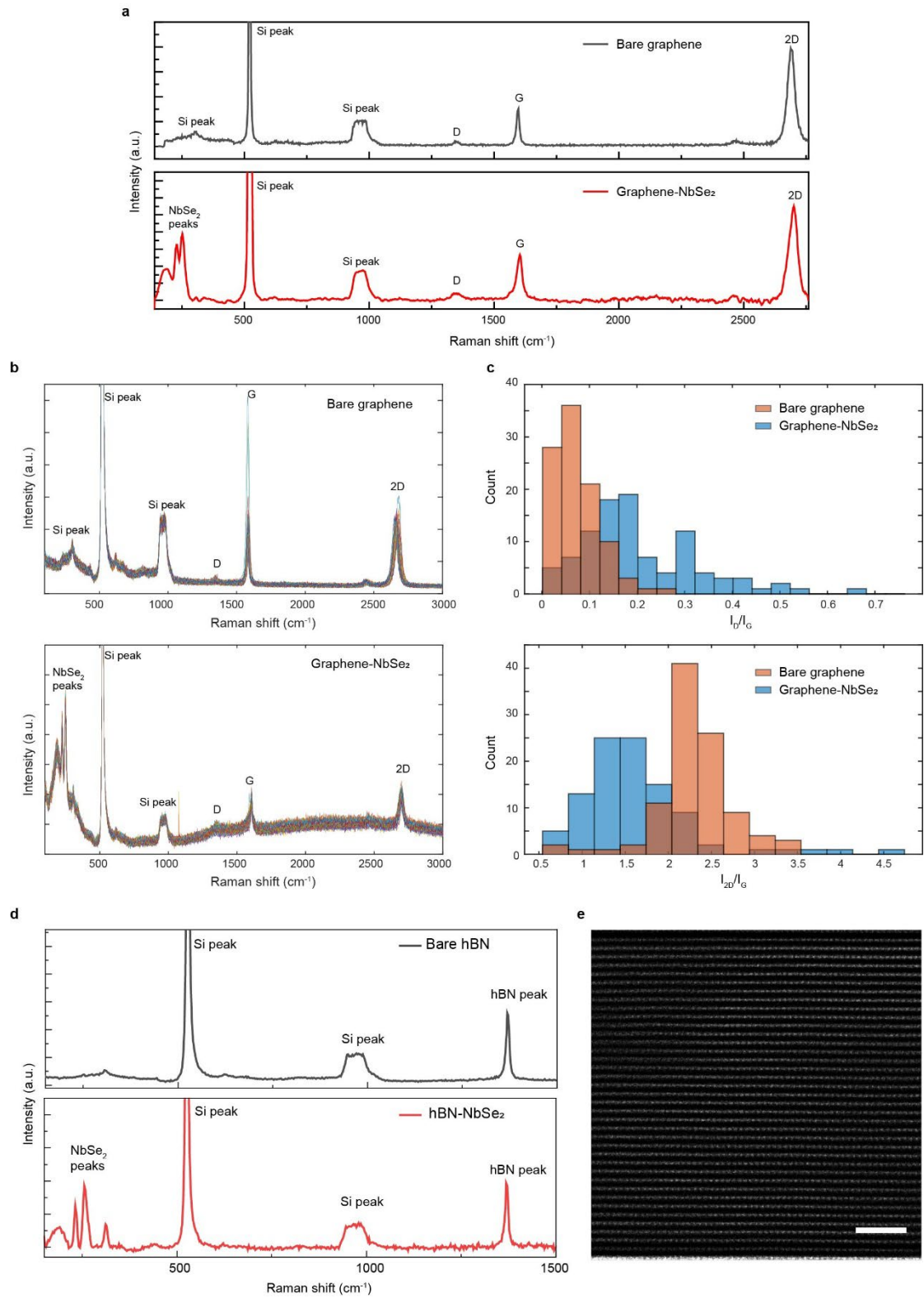
Supplementary Fig. 2 | Nucleation mechanism. **a**, Residence time of precursors on exposed graphene and SiO₂ surface calculated by DFT. **b**, Schematic comparing the nucleation behavior of precursors on exposed SiO₂ versus exposed graphene substrates. **c-f**, Schematic illustrating the nucleation starting locations of. **g**, Optical microscope images of NbSe₂ grown at hBN-SiO₂ interfaces with varying growth time and precursor concentrations. NbSe₂ regions and the likely nucleation mechanism are indicated in the graph. Scale bars, 20 μm. **h-k**, Optical microscope images of NbSe₂ grown at graphene-SiO₂ interfaces under varying growth times and precursor concentrations. **h**, Graphene patterned into square shapes. Scale bar, 10 μm. **i**, Graphene flakes used as the growth template. Scale bar, 20 μm. **j**, Continuous graphene used, with NbSe₂ partially grown along the interface. Scale bar, 10 μm. **k**, Continuous graphene used, showing continuous monolayer NbSe₂ growth at the interface. Scale bar, 20 μm.



Supplementary Fig. 3 | Nucleation mechanism characterization by AFM. a, AFM image measuring the edge of graphene-NbSe₂, revealing an edge-initiated nucleation mechanism. **b**, AFM images measuring continuous graphene-NbSe₂ film, revealing a wrinkle-initiated nucleation mechanism. Scale bars, 1 μ m.

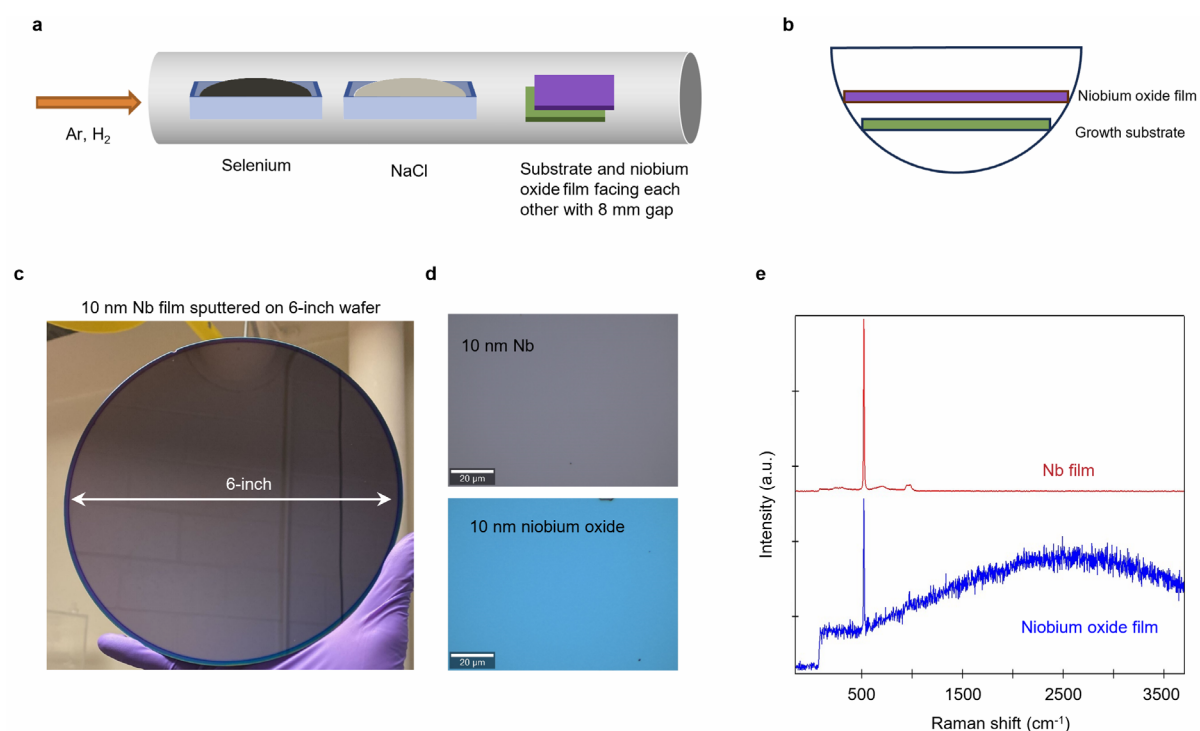


Supplementary Fig. 4 | Precursor diffusion. **a**, Diffusion coefficient of graphene and SiO₂ calculated by DFT. **b**, Schematic comparing the growth mode when using hybrid graphene-SiO₂ as substrate versus using bare SiO₂ as substrate. **c**, Optical microscope image of NbSe₂ grown at graphene-SiO₂ interface. **d**, Optical microscope image of NbSe₂ grown on bare SiO₂ surface. Scale bars, 20 μ m.

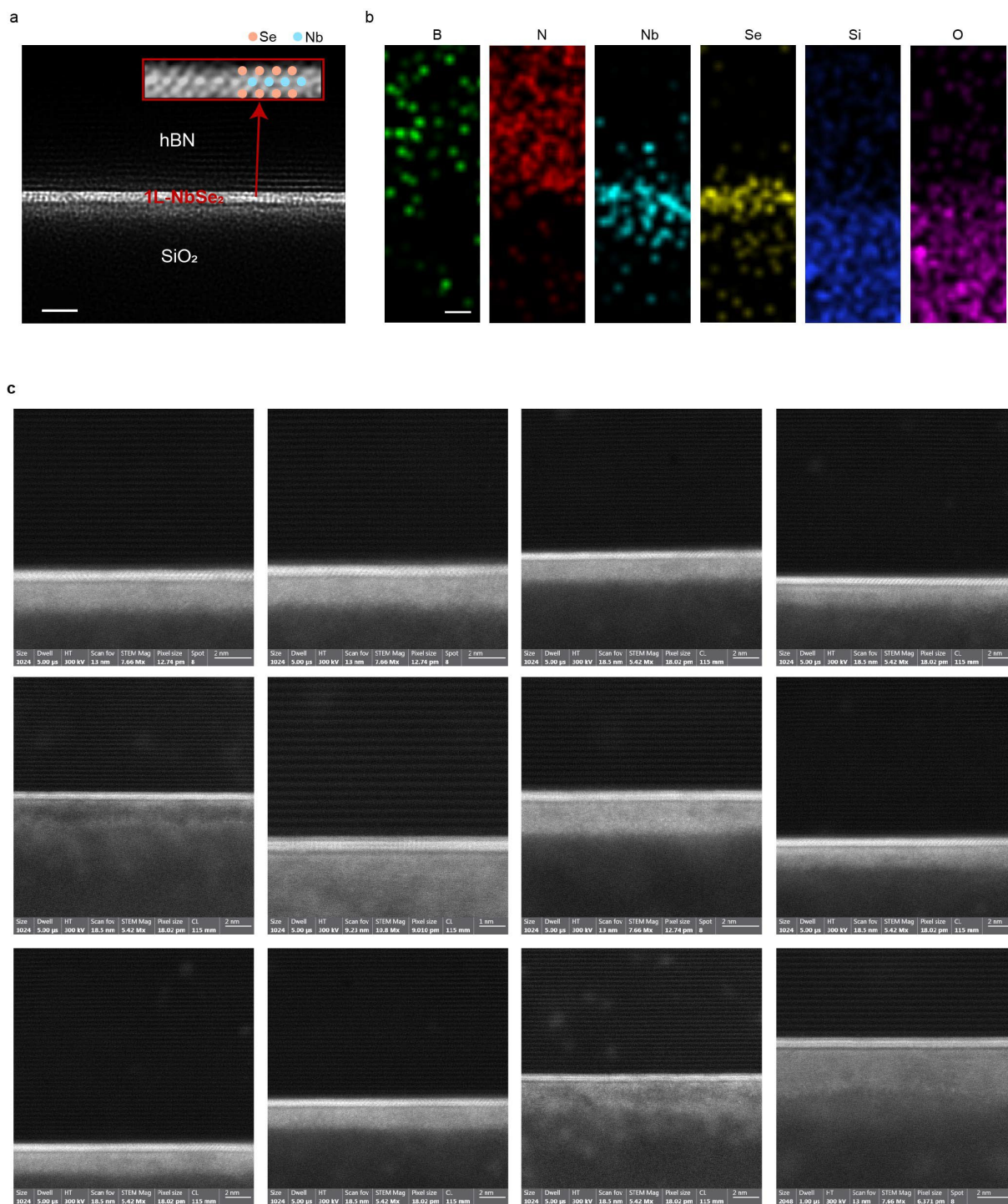


Supplementary Fig. 5 | Raman characterization of graphene and hBN before and after NbSe₂ growth. **a**, Representative Raman spectra of the graphene template before and after NbSe₂ growth, with the graphene-NbSe₂ spectrum baseline-subtracted. **b**, Statistical Raman spectra of bare graphene and graphene-NbSe₂ over 100 points. **c**, Statistical distributions of the Raman peak intensity ratios for graphene measured before and after growth, including (b) I_D/I_G and (c) I_{2D}/I_G . **d**, Comparison of hBN

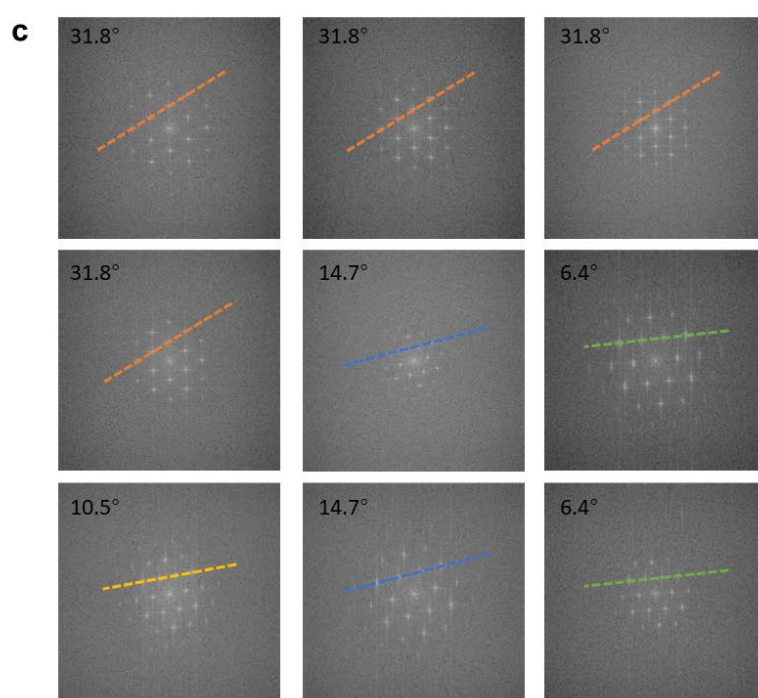
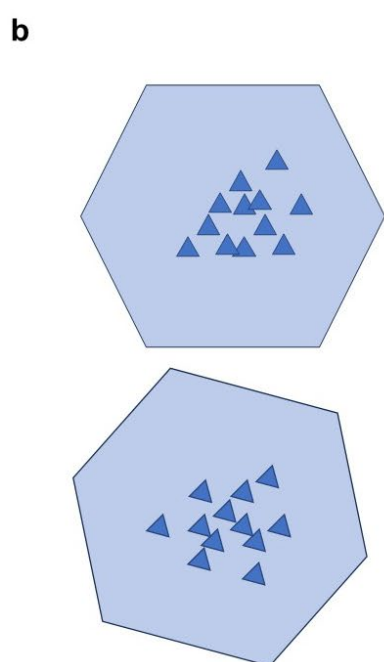
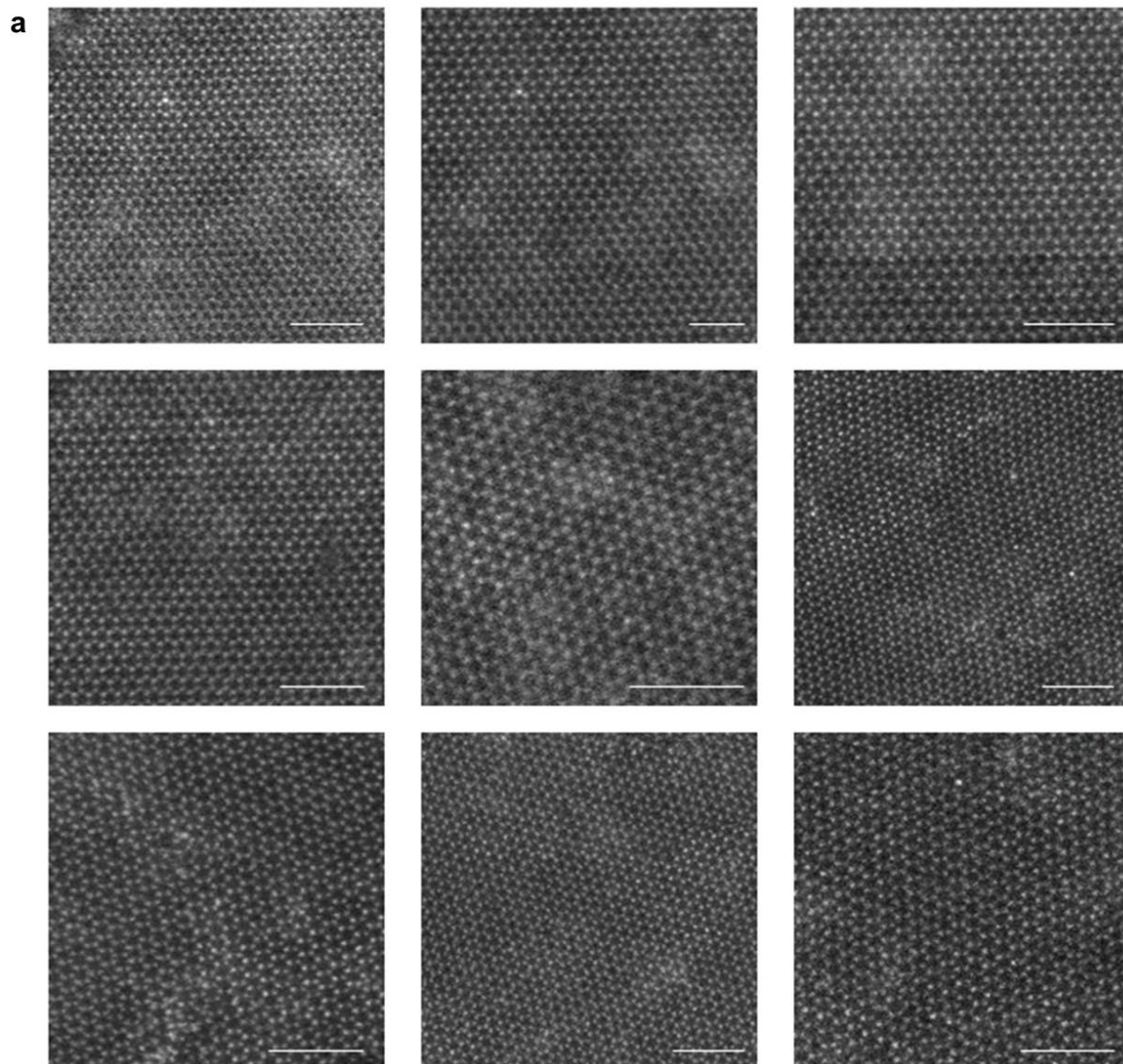
template before and after growth of NbSe₂, with the hBN-NbSe₂ spectrum baseline-subtracted. **e**, High-resolution cross-sectional STEM image of hBN after growth of NbSe₂. Scale bar, 2 nm.



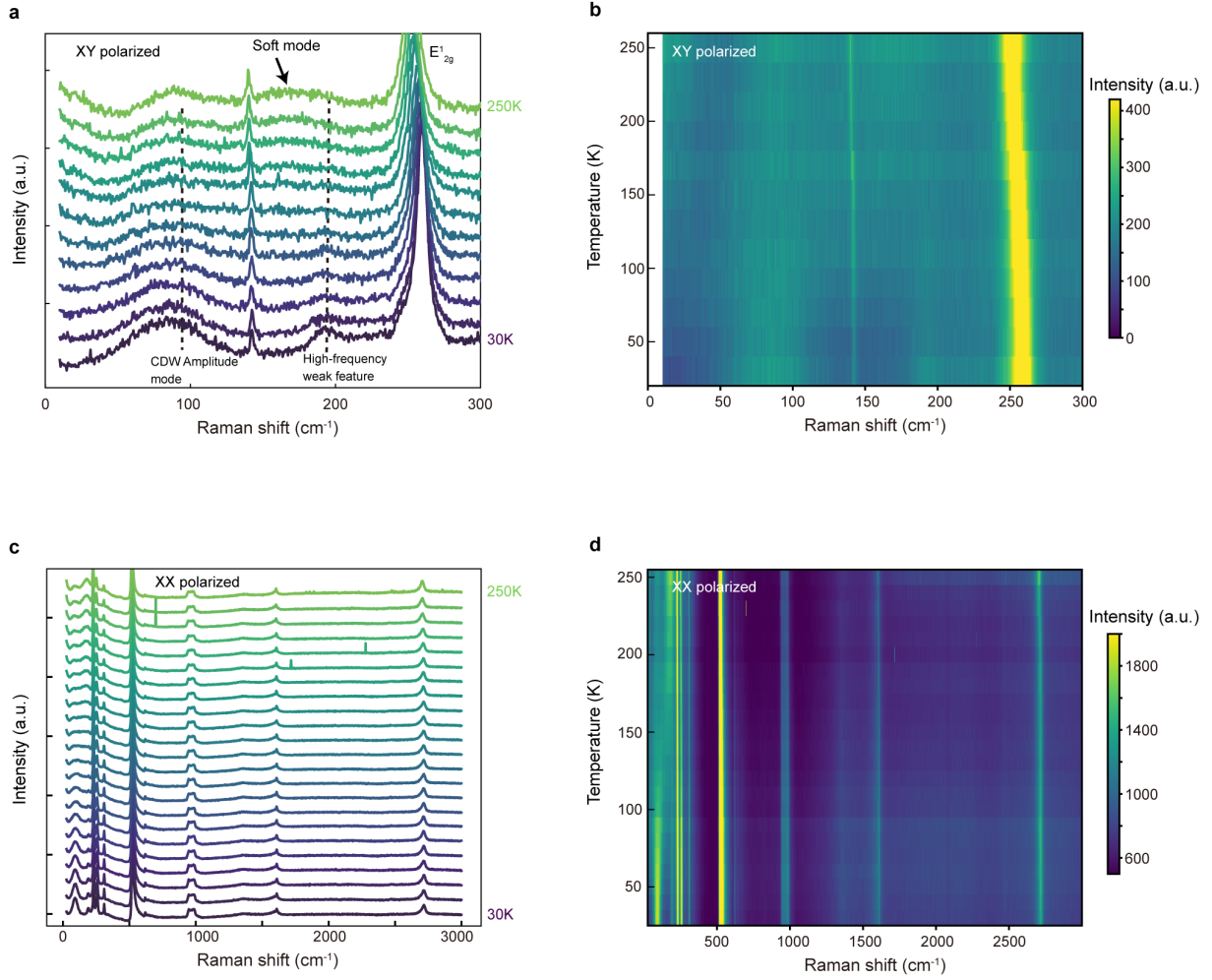
Supplementary Fig. 6 | Growth setup. **a**, Schematic showing the growth setup in the CVD system. **b**, Cross-sectional schematic illustrating the positioning of the substrate and precursor in **a**. **c**, Photo showing the as deposited Nb film (10 nm thick) by sputtering. **d**, Optical microscope image showing the as-deposited Nb film and the niobium oxide film obtained by oxidation in air. Scale bars, 20 μm . **e**, Raman spectra of the Nb film and niobium oxide film that show no discernable vibration modes from Nb or niobium oxide, which indicate an amorphous-like character.



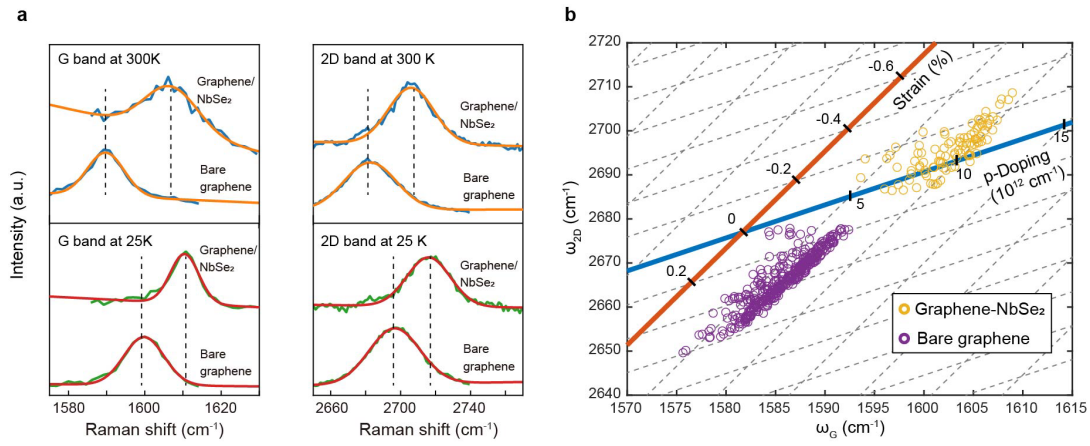
Supplementary Fig. 7 | Additional cross-sectional STEM characterization of monolayer NbSe₂ grown by encapsulation epitaxy at the interface of hybrid 2D-3D substrate. a, Cross-sectional STEM image of 1L-NbSe₂ grown at the interface between an hBN encapsulation layer and a SiO₂ substrate. The region on the top of the 1L-NbSe₂ in the image is fully occupied by hBN, but the contrast is low as it is adjusted for optimal visualization of NbSe₂. Scale bar, 2 nm. Inset, a zoomed-in image of the 1L-NbSe₂ cross section grown underneath hBN. **b**, EDS mappings for the sample in panel 1a. Scale bar, 2 nm. **c**, Extensive STEM images of NbSe₂ at hBN-SiO₂ interface.



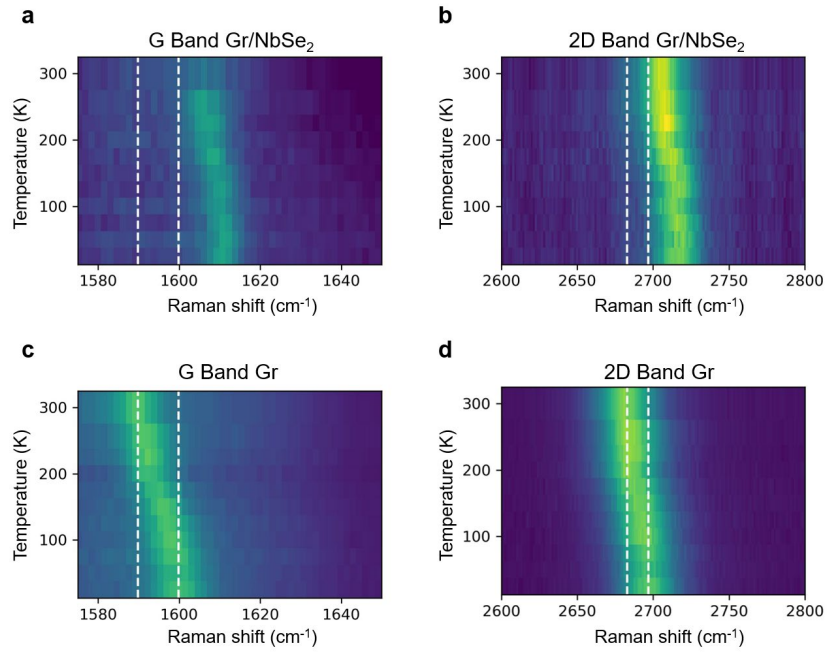
Supplementary Fig. 8 | Extensive HAADF-STEM characterizations. **a**, HAADF-STEM images of monolayer NbSe₂ grown by encapsulation epitaxy at the interface of a hybrid 2D-3D substrate, acquired across multiple regions of a millimeter-scale sample. Scale bars, 2 nm. Note there are small number of point defects in the images, which were caused by e-beam exposure (200 kV was used) during our HAADF-STEM imaging process. The varying NbSe₂ orientations across different HAADF-STEM imaged regions arise from the polycrystalline nature of the CVD-grown monolayer graphene we used, whose individual domains are not mutually aligned. Nevertheless, NbSe₂ grows epitaxially beneath the graphene within each domain, as confirmed by the selected area electron diffraction (SAED) characterization in Fig. 2b and Extended Data Fig. 7, which simultaneously capture the orientations of both NbSe₂ and overlying graphene. **b**, Schematic illustration of NbSe₂ grains (dark blue) growing under different CVD graphene domains (light blue). NbSe₂ grains within the same graphene domain share the same orientation, while grains from different domains adopt different orientations following the local graphene template. **c**, Fast Fourier Transformation (FFT) patterns extracted from the 9 HAADF-STEM images in panel a (arranged in the same order), revealing four distinct NbSe₂ orientations: images 1–4 share one orientation, images 5 and 8 share a second, images 6 and 9 share a third, and image 7 shows a fourth, consistent with imaging across approximately four CVD graphene domains of different orientations. The angle written in the top left of each figure corresponds to the angle between the dotted line and the horizontal.



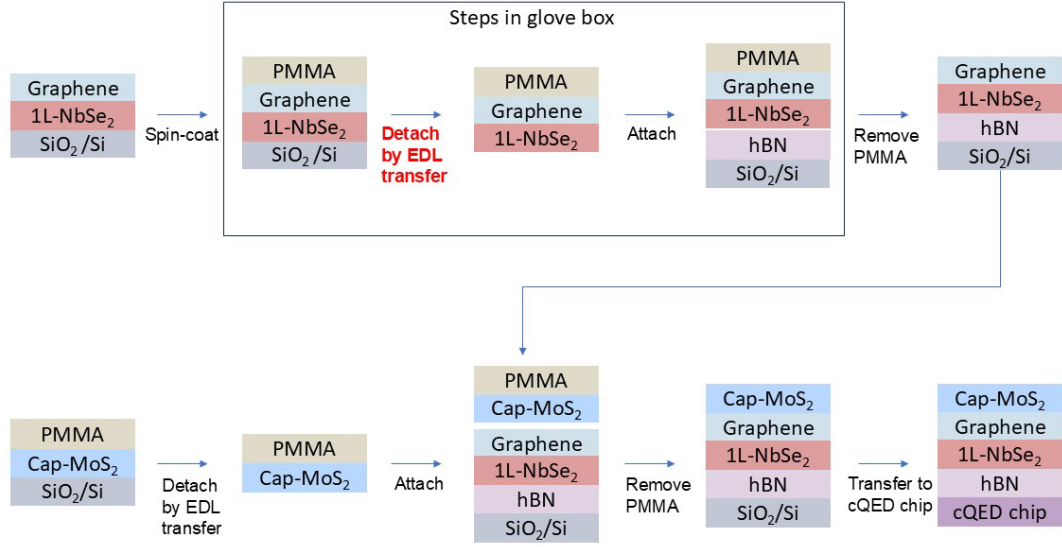
Supplementary Fig. 9 | Additional temperature dependent Raman measurement of 1L Graphene/NbSe₂ sample. **a**, Temperature dependent Raman measurement (from 30 K to 250 K with step size of 20 K). XY denotes that the incident excitation laser is linearly polarized, and the scattered light is resolved into linear perpendicular polarization before the spectrometer entrance slit. **b**, Corresponding Color map plot of **a**. **c**, Temperature dependent Raman measurement (from 30 K to 250 K with step size of 10 K). XX denotes that the incident excitation laser is linearly polarized, and the scattered light is resolved into linear parallel polarization before the spectrometer entrance slit. **d**, Corresponding Color map plot of **c**.



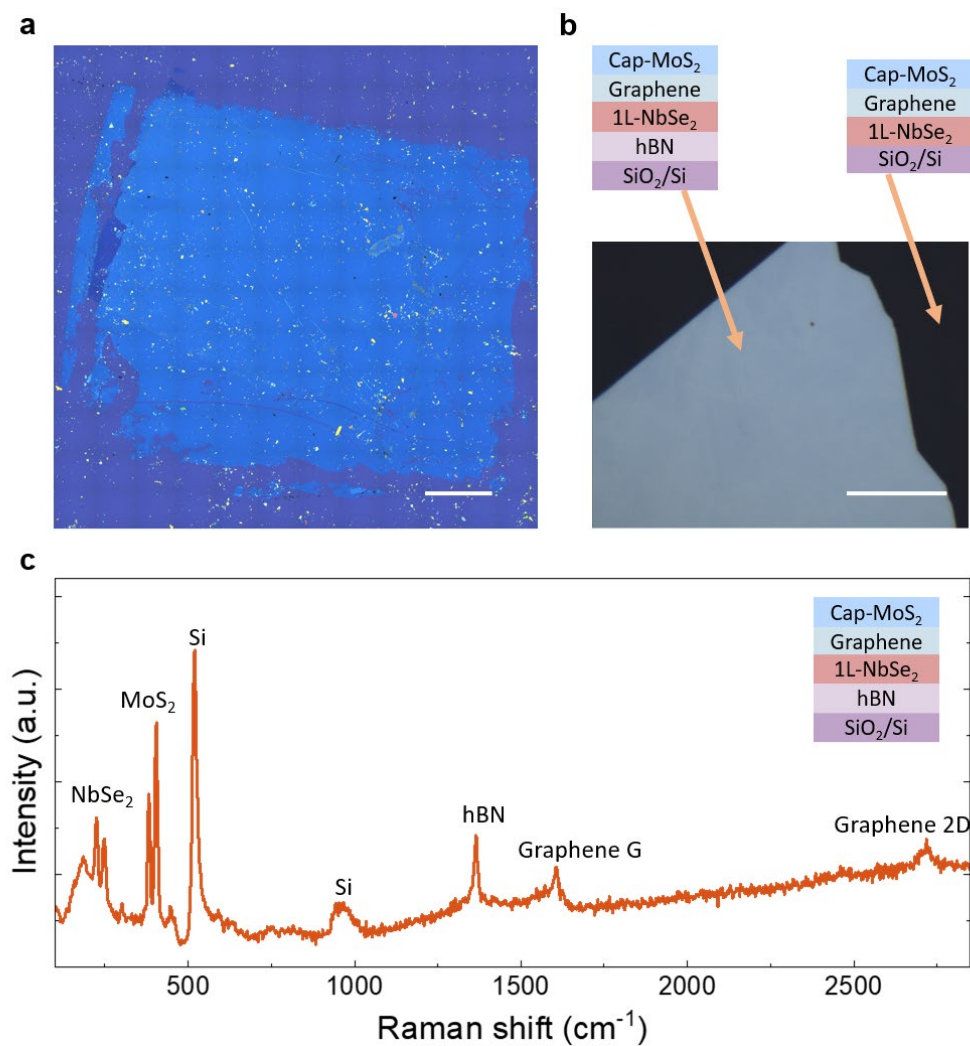
Supplementary Fig. 10 | Strain and doping analysis of graphene. a, Raman spectra of graphene before and after growth. **b,** Scatter plot of graphene G and 2D peaks.



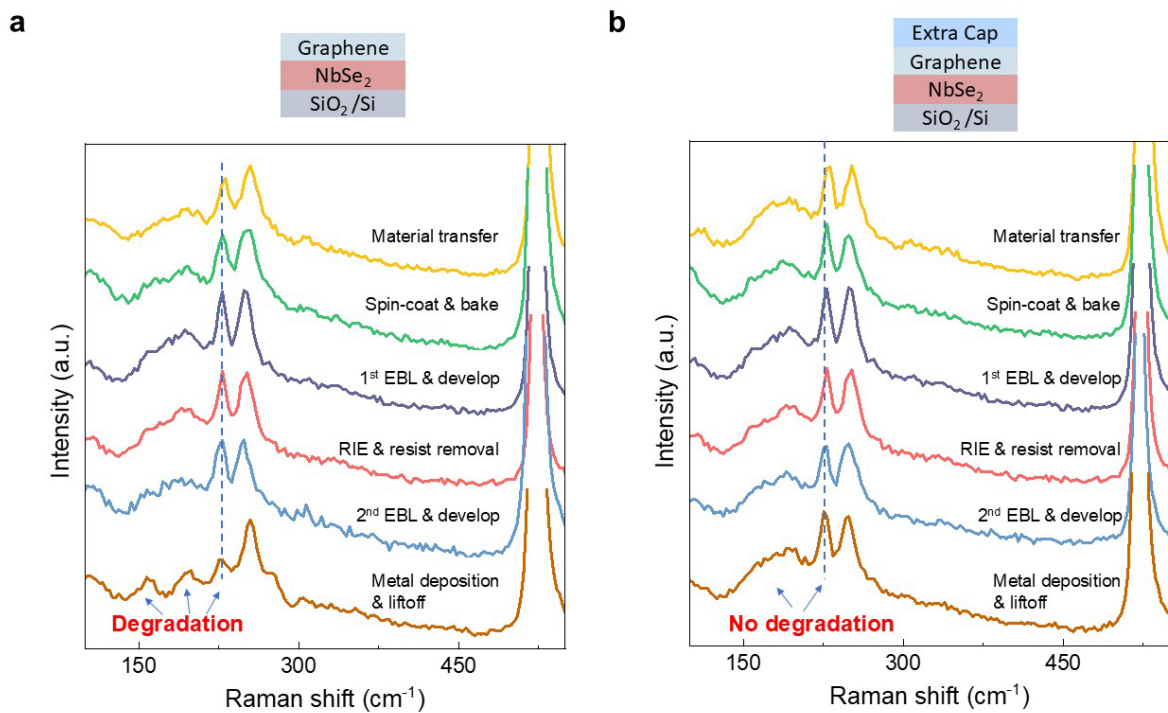
Supplementary Fig. 11 | Comparison between Graphene/NbSe₂/SiO₂ and pristine graphene/SiO₂ in terms of the evolution of Raman vibrational modes with temperature. Both G and 2D band of graphene on NbSe₂ (a-b) are clearly blue shifted as compared to bare graphene (c-d).



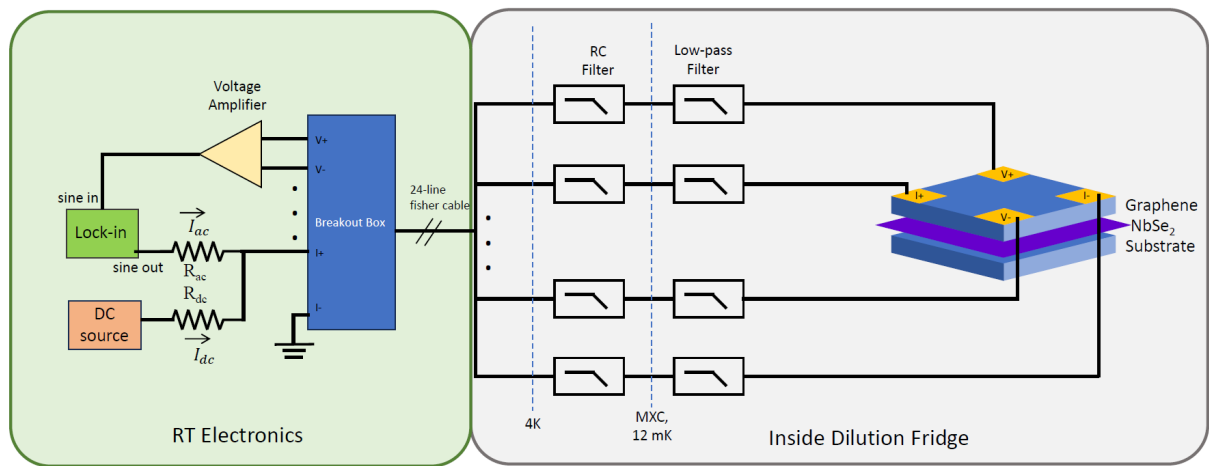
Supplementary Fig. 12 | Schematic illustration of assembling NbSe₂-based van der Waals heterostructures and integration to quantum circuits. Large-scale growth and transfer techniques are employed. In this work, the MoS₂, graphene and NbSe₂ are all monolayer continuous films synthesized via CVD techniques. The hBN is currently exfoliated, it can be replaced with CVD-grown hBN for scalable device integration. The MoS₂ layer serve as a protective barrier during metal evaporation (see Supplementary Fig. 6 for metal induced damage) and can be substituted with arbitrary 2D or thin films.



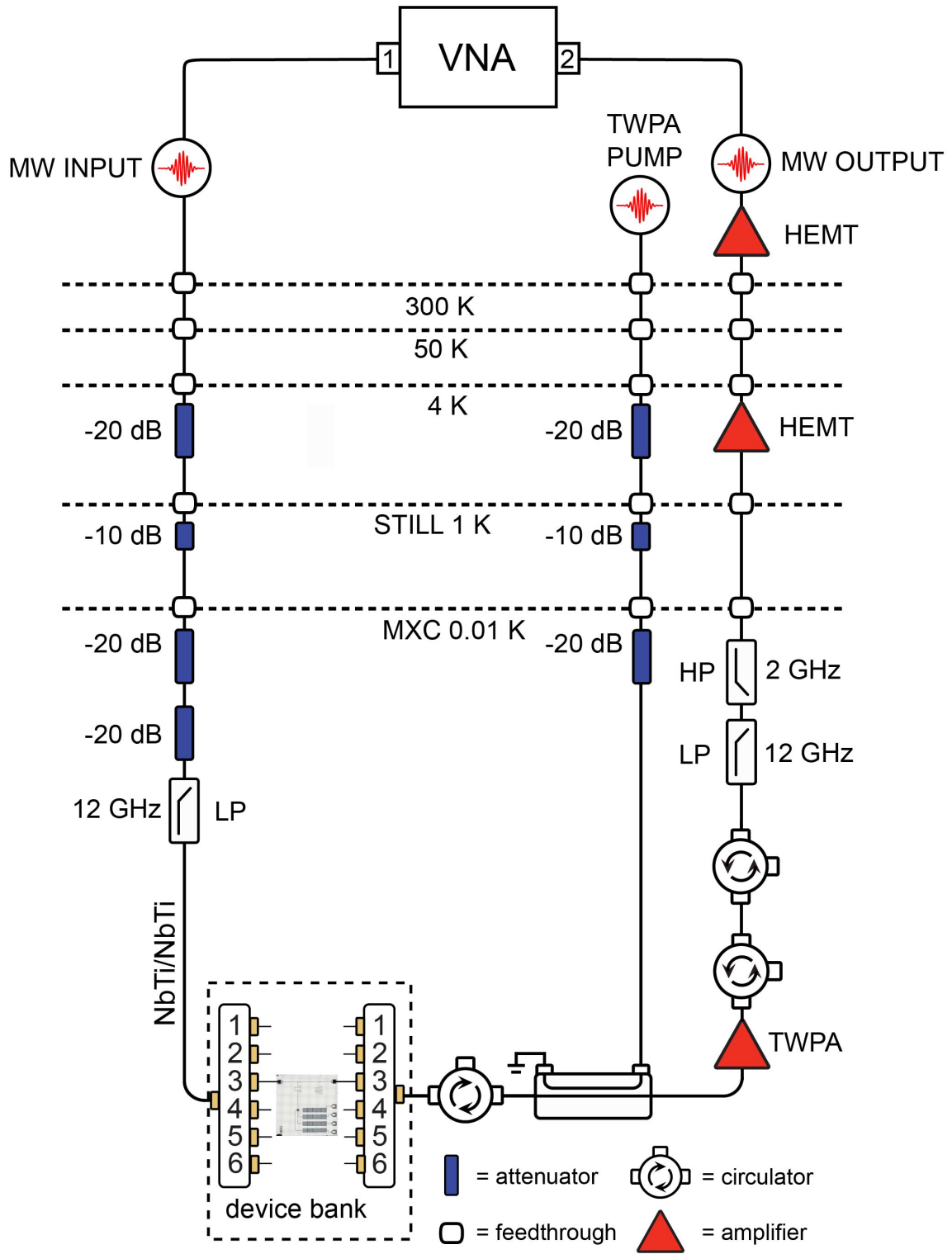
Supplementary Fig. 13 | Optical microscope images and Raman spectrum of the as-constructed van der Waals heterostructures. **a**, Stitched optical microscope images showing the Van der Waals heterostructure of MoS₂/graphene/NbSe₂/hBN on SiO₂/Si substrates. The many small flakes on the images come from the contrast of exfoliated hBN flakes. The blue rectangular areas are the regions covered by continuous MoS₂, graphene and NbSe₂ films, are all grown by scalable CVD methods and then stacked together. The outer purple color corresponds to bare SiO₂/Si substrate. Scale bar, 1 mm. **b**, Zoomed-in images of the heterostructure. The schematics show the places with and without hBN. Scale bar, 20 μm . **c**, Raman spectra measuring the heterostructure. The vibrational modes for NbSe₂ (soft mode, A_{1g}, E_{2g}), MoS₂ (E_{2g}, A_{1g}), silicon, hBN and graphene (G band and 2D band) are all simultaneously observed, confirming the successful assembly of the heterostructure. The structure also exhibits proper electrical functionality (see Fig. 4 in the main manuscript).



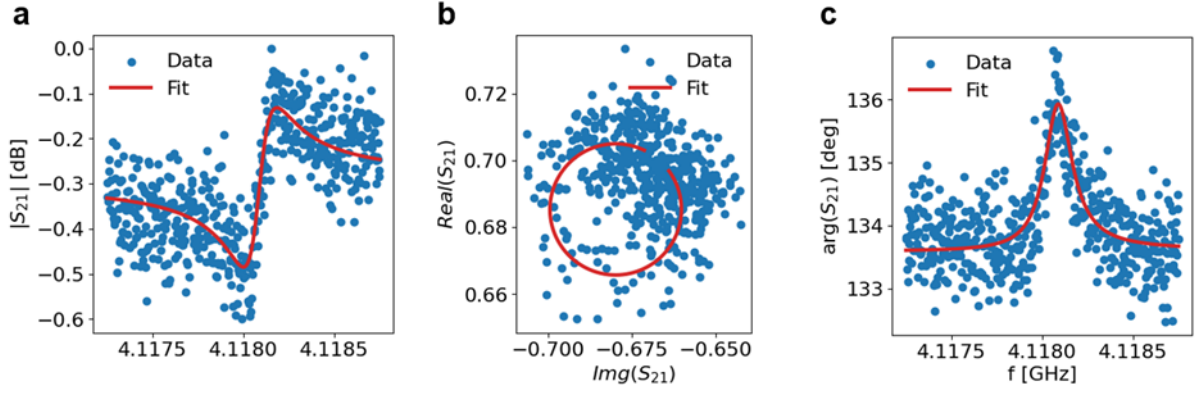
Supplementary Fig. 14 | Comparison of Graphene/ NbSe_2 with and without extra capping layer. a, Raman spectra measuring the as-grown NbSe_2 sample under graphene by encapsulation epitaxy after each fabrication step. With the graphene encapsulation layer, the monolayer NbSe_2 remain unoxidized throughout the fabrication steps in air except the second metal deposition step. An extra capping layer is needed to protect the NbSe_2 and ensure fully successful fabrication. **b,** Raman spectra measuring 1L graphene/ NbSe_2 sample that is protected by an extra cap layer (in this case, 1L-graphene). Raman remains intact throughout all fabrication steps.



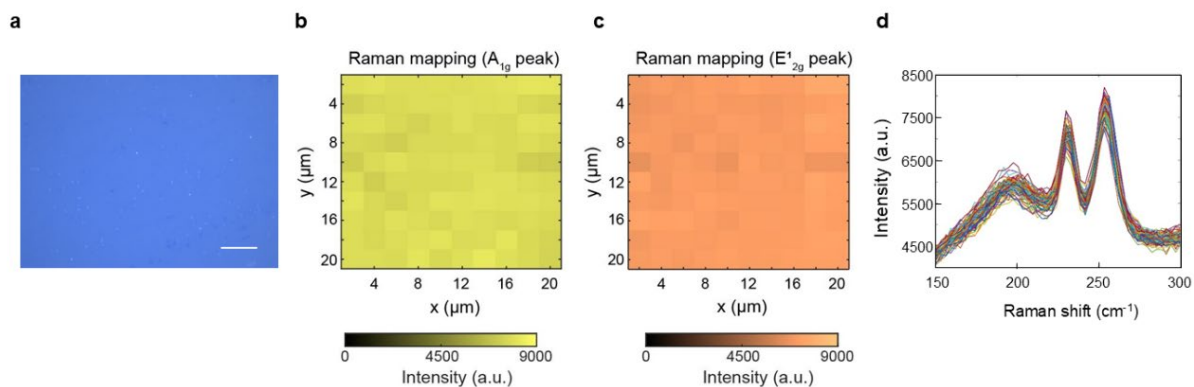
Supplementary Fig. 15 | Wiring diagram for the transport measurement. The room-temperature measurement setup is shown along with the DC wiring scheme inside the dilution refrigerator.



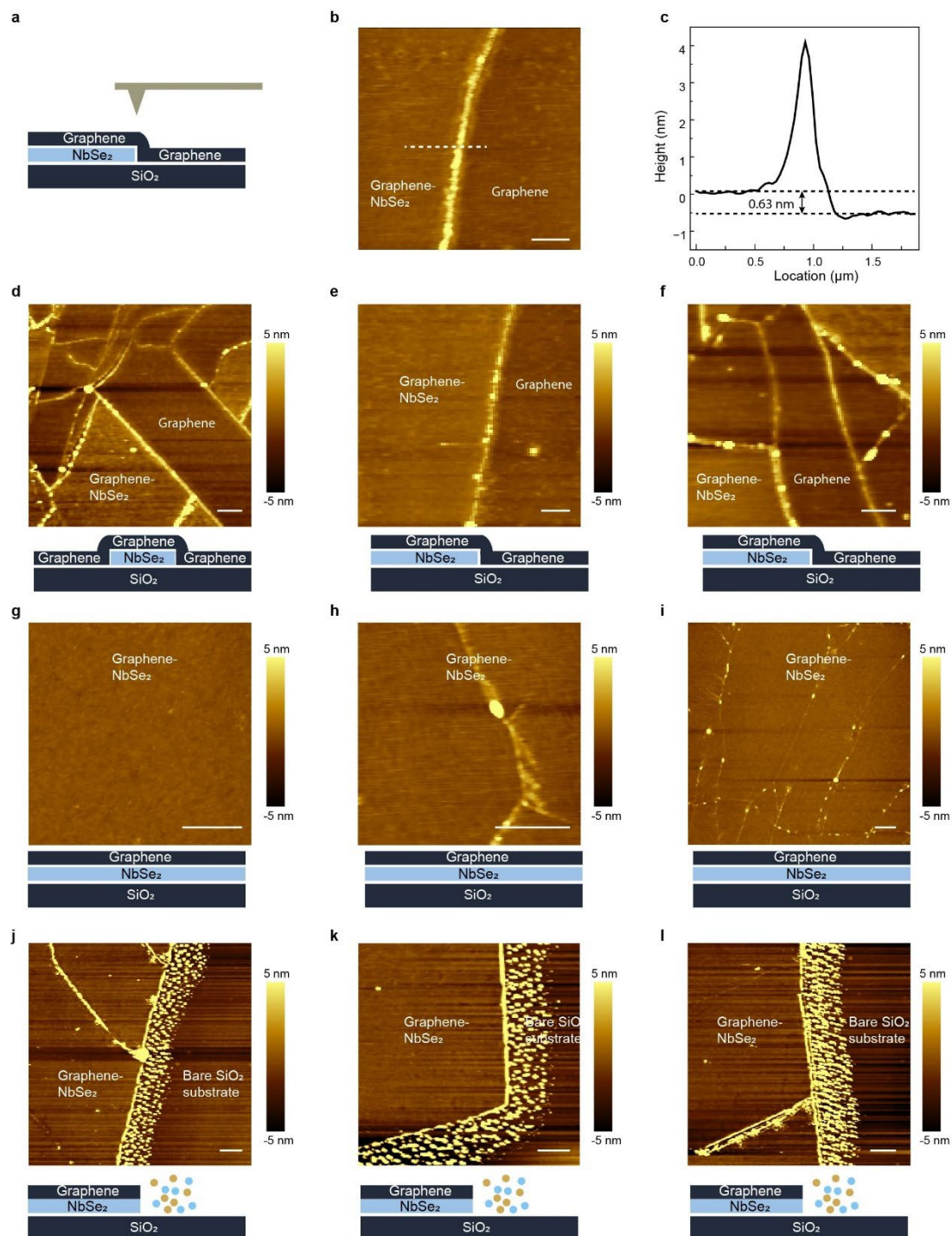
Supplementary Fig. 16 | Wiring diagram for the microwave characterization. Wiring diagram of the microwave measurements with all electronic components used in this experiment.



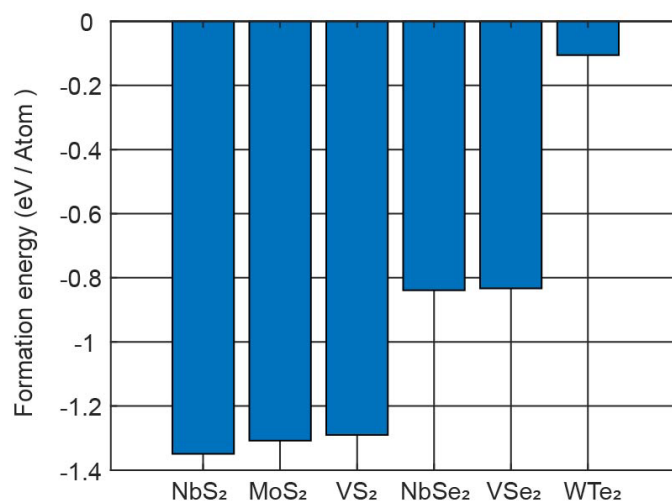
Supplementary Fig. 17 | Measurement at single-photon limit (-125 dBm). **a**, Resonator spectroscopy (transmission coefficient $|S_{21}|^2$) of a $\lambda/4$ aluminum CPW terminated ground with 1L-NbSe₂ film, with an applied RF power of -125 dBm at the sample. **b**, Complex I-Q plot of the transmission coefficient (S_{21}), with blue dots showing the measured response, and the red trace the Lorentzian fit. **c**, Phase of S_{21} as a function of frequency.



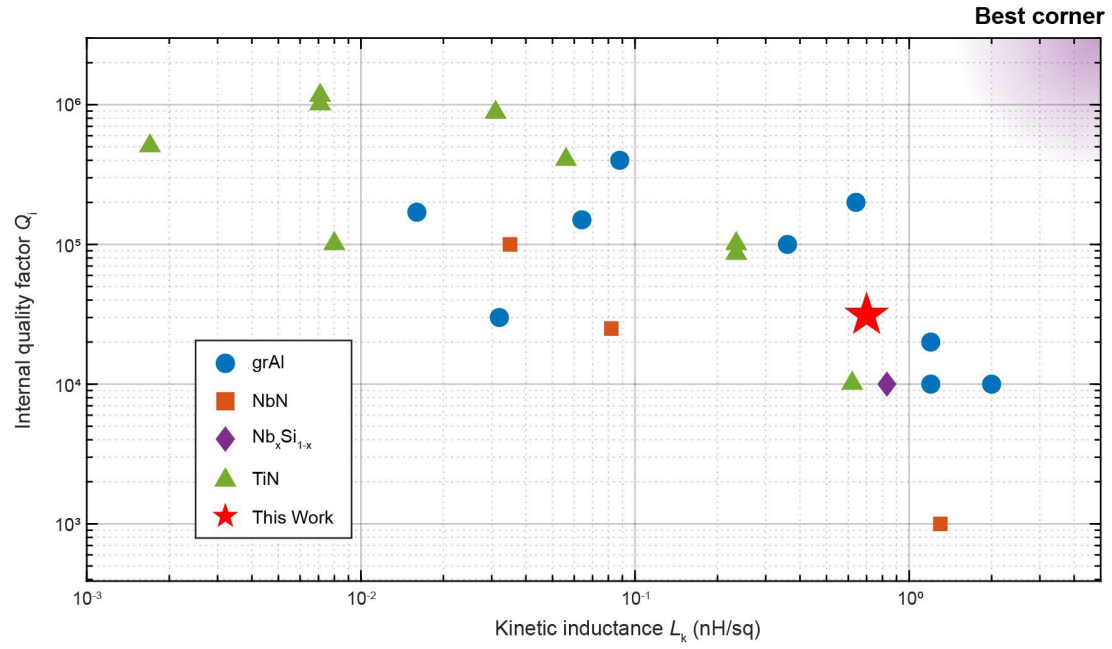
Supplementary Fig. 18 | Additional Characterization of NbSe₂ on SiN_x. **a**, Optical microscope image of NbSe₂ grown at graphene-SiN_x interface. Scale bar, 20 μm. **b-c**, Raman mapping of 20×20 μm² area indicating the uniformity of the NbSe₂. **d**, Overlay of all Raman spectra collected from the mapped area shown in panels **b-c**.



Supplementary Fig. 19 | Extensive AFM images of graphene-NbSe₂ on SiO₂ grown by encapsulation epitaxy. **a-c**, Schematic (panel **a**), AFM image (panel **b**) and height line profile (panel **c**) of a partially grown NbSe₂ under graphene. **d-l**, Additional AFM images and corresponding illustrations of NbSe₂ grown under graphene. Panels **d-f** show partially grown NbSe₂, panels **g-i** show continuous monolayer NbSe₂, and panels **j-l** show measurements taken at the graphene edge. Scale bars, 1 μm .



Supplementary Fig. 20 | DFT formation energies (eV/atom) for various TMD materials, obtained from the Materials Project database. Sulfide-based TMDs (NbS₂, MoS₂, VS₂) show significantly more negative formation energies than selenide- and telluride-based systems (NbSe₂, VSe₂, WTe₂), reflecting the decreasing electronegativity of the chalcogen from S to Se to Te.



Supplementary Fig. 21 | Benchmark kinetic inductance L_k with internal quality factor Q_i for superconducting thin films. The top right corner represents the ideal regime where high kinetic inductance and high internal quality factor are simultaneously achieved. “grAl” denotes granular aluminum.