


 Cite this: *RSC Adv.*, 2026, 16, 1134

Optical anisotropy induced by ultra-strong interfacial coupling in CVD-grown WSe₂/ReSe₂ vertical heterostructures

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Integrating high-mobility isotropic transition metal dichalcogenides (TMDs) with anisotropic layered materials offers an ideal strategy to break through the application bottleneck of a single material. Van der Waals assembly has proven a direct route to realize anisotropic heterostructures for polarization-sensitive electronics and photoelectronics, yet conventional methods typically rely on artificial stacking of TMD materials with varying symmetries. This approach not only limits the efficiency of fabrication but also yields poor interfacial coupling, resulting in weak anisotropy in heterostructures. Here, we directly grow high-quality monolayer WSe₂ on low-symmetry monolayer ReSe₂ to fabricate WSe₂/ReSe₂ vertical heterostructures by NaCl-assisted chemical vapor deposition (CVD). Angle-resolved polarization Raman spectroscopy confirms that the isotropic WSe₂ exhibits pronounced in-plane optical anisotropy upon combining with low-symmetry ReSe₂, which is attributed to the intrinsic symmetry breaking of WSe₂ induced by ultra-strong interlayer coupling between WSe₂ and ReSe₂. Notably, the pronounced photoluminescence (PL) quenching and shortened exciton lifetime of WSe₂ provide direct evidence of ultra-strong interfacial coupling in the CVD-grown WSe₂/ReSe₂ vertical heterostructures. Consequently, this study demonstrates the pivotal role of interfacial coupling in achieving high-degree anisotropy in TMD heterostructures, offering a new design paradigm for polarization-sensitive electronics and optoelectronics.

 Received 24th October 2025
 Accepted 10th December 2025

DOI: 10.1039/d5ra08161k

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1 Introduction

Two-dimensional (2D) van der Waals (vdW) transition metal dichalcogenides (TMDs) have attracted particular attention owing to their tunable bandgaps, excellent electronics and optoelectronic performance, and wafer-scale manufacturability.^{1–4} However, most reported TMDs are high-symmetry crystals that exhibit in-plane isotropy, which fundamentally limits their prospects in applications demanding polarization-sensitive electronic and photonic responses.^{5–7} At the same time, low-symmetry TMDs such as ReSe₂ and ReS₂ exhibit pronounced anisotropy, making them promising for polarization applications.^{8,9} Nevertheless, the intrinsically modest carrier mobility and the formidable challenge of wafer-scale synthesis of these anisotropic TMDs are currently limiting

their transition from laboratory demonstrations to industrial applications.^{10,11}

Recently, constructing TMDs with distinct properties into heterostructures has proven an effective means to extend and modulate the intrinsic properties of individual TMDs,^{12,13} advancing the development of multifunctional integrated devices. In light of this, symmetry engineering *via* vdW heterostructures is rapidly becoming a key strategy for tailoring the anisotropic properties of TMDs,^{14,15} offering a practical route toward polarization-sensitive light emission and detection.^{16,17} Particularly, assembling heterostructures of anisotropic and high-mobility isotropic TMDs is expected to fully exploit the performance advantage of each component.^{14,18} Previous studies have shown that the anisotropic characteristics in the isotropic/anisotropic TMDs heterostructures can be both tuned and transmitted through interlayer coupling.^{19,20} However, the fabrication of such isotropic/anisotropic heterostructures typically relies on the artificial stacking of different TMDs materials through dry or wet transfer methods.^{17,19,20} On the one hand, these approaches inevitably introduces contaminants at the interface and may even damage the materials.²¹ On the other hand, this transfer-assembled heterostructures usually exhibit poor interfacial coupling,^{19,22} leading to a weak

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anisotropy. Therefore, establishing isotropic/anisotropic TMD-based heterostructures with robust interfacial coupling is the critical first step toward unlocking TMDs for high-performance polarization-sensitive applications.

In this study, we prepare high-quality $\text{WSe}_2/\text{ReSe}_2$ vertical heterostructures in a one-step chemical vapor deposition (CVD) process, and further reveal their ultra-strong interfacial coupling through temperature-dependent and time-resolved photoluminescence (PL) spectroscopy. Angle-resolved polarization Raman spectroscopy confirms the inherent in-plane optical anisotropy of as-grown monolayer ReSe_2 . As expected, the CVD-grown $\text{WSe}_2/\text{ReSe}_2$ vertical heterostructures exhibit pronounced optical anisotropy manifested through WSe_2 Raman modes, which stems from the symmetry breaking of WSe_2 induced by ultra-strong interfacial coupling with low-symmetry ReSe_2 . Meanwhile, the absence of such anisotropy in the mechanically-assembled $\text{WSe}_2/\text{ReSe}_2$ heterostructure further confirms that the robust interfacial coupling is essential for symmetry engineering. Consequently, one-step-grown isotropic/anisotropic TMD-based heterostructures offer an effective strategy to polarization-sensitive optoelectronic devices. Furthermore, our findings also provide valuable insights of interfacial coupling on anisotropy engineering.

2 Experimentals

2.1 Sample fabrication and structure characterization

Monolayer ReSe_2 and $\text{WSe}_2/\text{ReSe}_2$ vertical heterostructures were grown through NaCl-assisted CVD method. Specifically, WO_3 power was first thermally evaporated onto a sacrificial SiO_2/Si substrate to form a uniform film of 9 nm. 50 mg Re power was dispersed in 20 ml ethanol solution, and the upper suspension was dripped onto the WO_3 film surface, followed by drying at 80 °C for 5 min. Subsequently, a single NaCl crystal (~ 0.01 mg) was placed on the sacrificial substrate, while a clean SiO_2/Si growth substrate was mounted face-to-face on the sacrificial substrate. This entire stack was loaded into the central heating zone of tube furnace, while Se power (200 mg) was arranged in the edge heating zone of the quartz tube. Argon (100 sccm) was continuously supplied as the carrier gas throughout the entire growth process. The central zone was heated to 935 °C and the edge zone to 240 °C over 60 min, subsequently both zones were held at their maximum temperatures for 10 min. After reaching the set temperatures, a brief 2 min H_2 (5 sccm) was introduced, and the furnace was then allowed to cool naturally to room temperature.

The surface morphology and thickness of prepared $\text{WSe}_2/\text{ReSe}_2$ vertical heterostructures were characterized using an NX10 atomic force microscope (AFM) (Park, Korea). No-contact AFM imaging was performed with an AC160TS probe (resonance frequency 270.8 kHz) in tapping mode, using an automated step-scan at 1.0 Hz with 512×512 pixel resolution.

2.2 Optical measurements

Raman and PL measurements were conducted using a confocal Raman microscopy system (WITec Alpha 300R) under a 532 nm

laser excitation. The 1800 lines per mm and 150 lines per mm gratings were used to collect Raman and PL spectra, respectively. A $100\times$ objective lens (Zeiss, N.A. = 0.9) was employed for room-temperature measurements. For low temperature *in situ* PL measurements, samples were mounted inside the chamber of a multifunctional system with liquid helium refrigeration (CRYO Industries of America, Inc), enabling cooling from 300 K to 10 K. Particularly, the lens was switched to a $50\times$ long-focus objective (Zeiss, N.A. = 0.55), while the sample chamber was maintained at a high vacuum level of 10^{-3} Pa. For angle-resolved polarization Raman measurements, samples were installed at the center of a rotating platform and tuned from 0° to 180° , with a spectrum acquisition every 15° under both parallel and crossed polarization configurations.

3 Results and discussions

Fig. 1 presents the fabrication process of the $\text{WSe}_2/\text{ReSe}_2$ vertical heterostructures by space-confined CVD technique. Fig. 1a shows the schematic diagram of the CVD tube furnace, where Se precursor is placed in the left heating zone, while WO_3 and Re precursors are placed in the right heating zone. Considering the ultra-high melting point of Re (>3000 °C), a NaCl-assisted strategy was employed to achieve monolayer ReSe_2 growth. Fig. S1 provides the comparative optical images of $\text{WSe}_2/\text{ReSe}_2$ heterostructures grown with and without NaCl assistance. Specifically, during the CVD growth process, the introduction of NaCl auxiliaries not only lowers the reaction temperature, but also simultaneously increases the precursor supersaturation and the density of nucleation points.^{23,24} Fig. 1b and c depict the growth mechanism of the $\text{WSe}_2/\text{ReSe}_2$ vertical heterostructure. When the Se and Re precursors are heated to their target temperatures, the ReSe_2 molecular clusters first nucleate and grow into irregular shapes on the SiO_2/Si substrate surface. Subsequently, as the WO_3 precursor reaches its set temperature, the WSe_2 molecular clusters further nucleate on the surface of the existing ReSe_2 and gradually grow into triangular shapes. Fig. 1d and e show the optical images of the as-grown ReSe_2 film and $\text{WSe}_2/\text{ReSe}_2$ heterostructures, respectively. Particularly, the successful preparation of the heterostructures combining isotropic WSe_2 with anisotropic ReSe_2 , offering an idea platform to investigate the intrinsic anisotropic properties in vertically stacked heterostructures.

We first conducted structural and optical characterization for the vertical $\text{WSe}_2/\text{ReSe}_2$ heterostructures. Fig. S2 provides the AFM scanning images and corresponding height profiles. Here, the boundary between the underlying ReSe_2 and the substrate is clearly visible, while the triangular shape of the top-layer WSe_2 can also be observed. The ReSe_2 flake shows a thickness of approximately 1.3 nm, consistent with a monolayer characteristics on SiO_2/Si substrates. The top WSe_2 layer was scanned more precisely, exhibiting a height of approximately 0.8 nm, which is highly close to the theoretical thickness of monolayer WSe_2 . These results confirm the successful formation of $\text{WSe}_2/\text{ReSe}_2$ bilayer vertical heterostructures. Furthermore, we also extracted the RMS roughness from the high-resolution AFM image of the as-grown $\text{WSe}_2/\text{ReSe}_2$



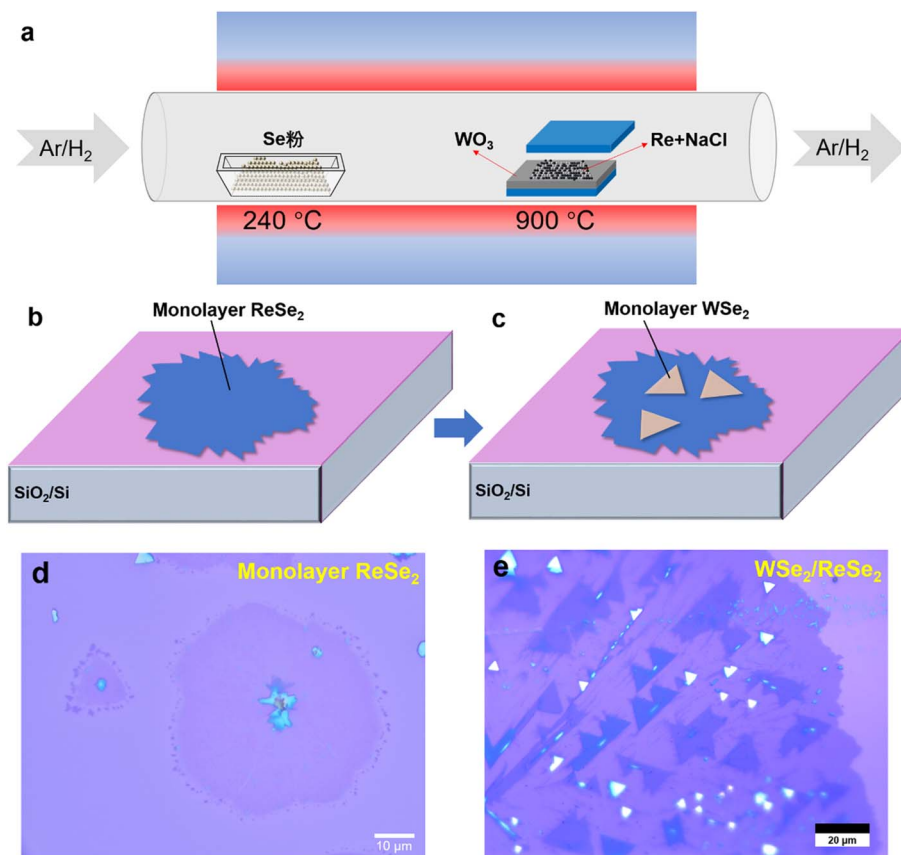


Fig. 1 Fabrication process of the WSe₂/ReSe₂ vertical heterostructures by NaCl-assisted CVD method. (a) The schematic diagram of the CVD tube furnace for WSe₂/ReSe₂ heterostructures growth. (b) and (c) The growth mechanism of the WSe₂/ReSe₂ vertical heterostructure. (d) Optical image of the as-grown monolayer ReSe₂ film. (e) Optical image of the as-grown WSe₂/ReSe₂ vertical heterostructures.

heterostructure, obtaining $S_q = 0.0007 \mu\text{m}$ and $S_a = 0.0006 \mu\text{m}$. Such sub-nanometer roughness suggests an exceptionally flat interface. Fig. 2a shows the Raman spectrum of the monolayer ReSe₂ under a 532 nm laser excitation. Due to the low-symmetry crystal structure, theory predicts that there are 18 Raman modes in monolayer ReSe₂,²⁵ which is rarely observed in most transition metal dichalcogenides.²⁶ To obtain the accurate vibration frequency of each Raman mode, we performed detailed Lorentzian fitting for all Raman peaks, as shown by the colored dotted lines in Fig. 2a. According to the fitted results, 17 Raman modes can be successfully identified in the as-grown monolayer ReSe₂. Table S1 summarizes the fitted frequencies of the 17 Raman modes and compares them with the calculated values. The minute differences between experiment and theory confirm the excellent crystal quality of the as-grown monolayer ReSe₂. On the other hand, we also notice that only 16 Raman modes were experimentally observed in the ref. 25. This suggests that the finite spectral resolution may limit the extraction of all 18 Raman modes. Fig. 2b further compares the Raman spectra of as-grown monolayer ReSe₂, monolayer WSe₂, and WSe₂/ReSe₂ vertical heterostructure. Two prominent Raman modes E_{2g}^1 (247.8 cm⁻¹) and A_{1g} (256.7 cm⁻¹) of WSe₂ are observed in the heterostructure region, which are individually assigned to the in-plane and out-of-plane vibration of W and Se atoms.²⁶

The inset shows a uniform Raman image of the heterostructure obtained by integrating the intensity of E_{2g}^1 mode, indicating that the top-layer WSe₂ retains a pristine lattice quality. Compared with standalone monolayer WSe₂, the E_{2g}^1 mode exhibits a slight redshift, whereas the A_{1g} mode undergoes a pronounced redshift in the heterostructure region. Particularly, in TMDs heterostructures, the in-plane vibration (E_{2g}^1) is known to be relatively insensitive to the interfacial interaction, while A_{1g} mode is highly sensitive to interlayer coupling due to its out-of-plane vibrational characteristic. Consequently, the marked shift of the A_{1g} peak provides compelling evidence for effective interfacial coupling within the WSe₂/ReSe₂ heterostructure. In addition, these weak Raman peaks originating from underlying ReSe₂ are observed as well, which further confirms the formation of hetero-stacked structure.

In typical bilayer TMDs heterostructures, the interaction between two components mainly depends on the interfacial coupling,^{22,27} for which strong interlayer coupling is highly desired. Therefore, we further conducted PL spectroscopy measurements for the as-grown WSe₂/ReSe₂ heterostructures to evaluate the interlayer coupling strength, as shown in Fig. 2c. It is well known that monolayer WSe₂ exhibits strong PL emission at room temperature due to its direct bandgap feature (~1.6 eV).²⁸ In stark contrast, the PL emission of WSe₂ is completely



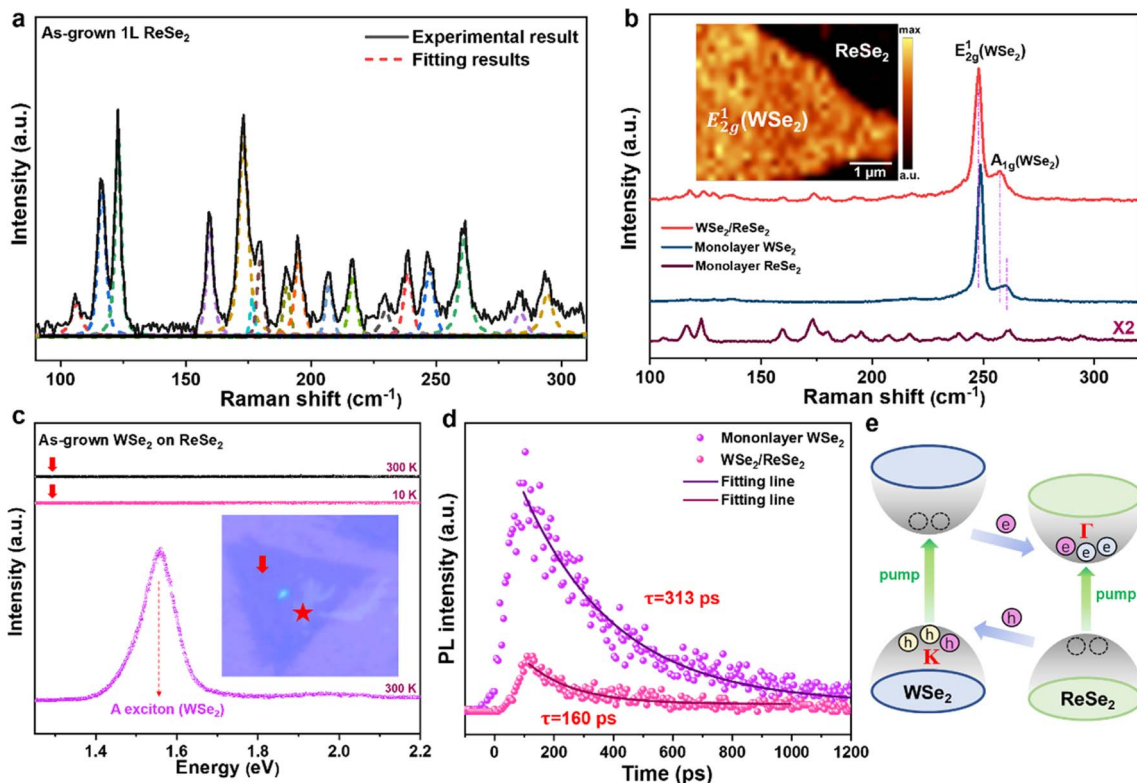


Fig. 2 Optical characterization of the CVD-grown monolayer ReSe₂, monolayer WSe₂, and WSe₂/ReSe₂ vertical heterostructures. (a) Raman spectrum of the as-grown monolayer ReSe₂ with detailed Lorentzian fitted peaks (colored dotted lines). (b) Comparison of Raman spectra between the monolayer ReSe₂, monolayer WSe₂, and WSe₂/ReSe₂ vertical heterostructure, the inset shows a spatial intensity map generated by integrating the E_{2g}¹(WSe₂) peak of WSe₂. (c) PL spectra of the WSe₂/ReSe₂ vertical heterostructure measured at 300 K and 10 K, together with the spectrum (300 K) obtained after 1064 nm laser irradiation treatment (The red star marks the laser-irradiated region). (d) Time-resolved PL measurements of the standalone monolayer WSe₂ and WSe₂/ReSe₂ heterostructure. (e) The schematic diagram of interfacial charge transfer in the WSe₂/ReSe₂ vertical heterostructures.

quenched in the as-grown WSe₂/ReSe₂ heterostructures at both 300 K and 10 K, as indicated by the black and pink lines in Fig. 2c. We preliminarily attribute this pronounced quenching to two possibilities: (i) in as-grown monolayer WSe₂, a high density of defects (such as vacancies and dislocations) may act as non-radiative recombination centers, significantly reducing the radiative recombination efficiency of photo-generated carriers; (ii) ultrafast interlayer charge transfer occurs due to the type-II band alignment formed between WSe₂ and ReSe₂, causing significant PL quenching.²⁹ To verify the above deduction, we employed a 1064 nm infrared laser to perform localized heating for the CVD-grown WSe₂/ReSe₂ heterostructure. Through high-power irradiation, the underlying ReSe₂ was effectively removed (see the inset in Fig. 2c), which enables *in situ* monitoring of the WSe₂ PL emission before and after ReSe₂ removal. Interestingly, once the ReSe₂ layer is removed, the isolated monolayer WSe₂ recovers strong PL emission even at room temperature, as depicted by the purple trace. This finding strongly indicates that the as-grown monolayer WSe₂ possesses excellent lattice quality, thus ruling out the possibility that high density defects are responsible for the observed PL quenching in WSe₂/ReSe₂ heterostructures. Furthermore, we also provided more direct experimental evidence, as depicted in Fig. 2d. We

conducted time-resolved PL measurements to compare the A-exciton lifetime of WSe₂ in the heterostructure with that of the standalone monolayer WSe₂. According to the fitting results, the A-exciton of monolayer WSe₂ exhibits a lifetime of ~313 ps, whereas the lifetime is reduced to ~160 ps in the heterostructure region. This pronounced shortening provides direct evidence of efficient interlayer charge transfer at the WSe₂/ReSe₂ interface.

Consequently, we can reasonably conclude that the PL quenching behavior occurred in WSe₂/ReSe₂ heterostructures originates from high-efficiency interlayer charge transfer.^{29,30} As illustrated in Fig. 2e, upon 532 nm laser excitation, the photo-generated electrons are first excited to the conduction band minimum (CBM) of WSe₂, and subsequently transfer to the CBM of ReSe₂ in an extremely short time scale. Simultaneously, the photo-generated holes transfer from the valence band maximum (VBM) of ReSe₂ to that of WSe₂, leading to electron-hole separation across the two layers. Notably, such efficient interfacial charge transfer usually necessitates strong interlayer coupling between the constituent layers of heterostructures.^{22,31} The experimental findings collectively support the conclusion that the as-grown WSe₂/ReSe₂ heterostructures possess robust interfacial coupling strength.



It should be noted that in most cases the PL quenching induced by interlayer charge transfer is incomplete, intralayer exciton emission from the constituent layers can still be detected at low temperatures.^{32,33} In this present study, however, no intralayer exciton signal from WSe₂ is observed even at 10 K, providing further evidence for exceptionally strong interfacial coupling. In addition, the absence of interlayer exciton emission at low temperature is mainly ascribed to momentum mismatch.³⁴ Specifically, in the WSe₂/ReSe₂ heterostructure, interlayer exciton emission originates from the recombination of electrons at CBM of ReSe₂ and holes at VBM of WSe₂. Theoretical studies show that the CBM of monolayer ReSe₂ is located at the Γ point of the Brillouin zone,³⁵ whereas the VBM of monolayer WSe₂ lies at the K point.³⁶ This momentum-space mismatch greatly suppresses the probability of radiative recombination between spatially separated electrons and holes.³⁴

As is well known, monolayer WSe₂ is intrinsically intralayer isotropic, whereas monolayer ReSe₂ is a typical intralayer anisotropic material. The strong interlayer coupling demonstrated here in the CVD-grown WSe₂/ReSe₂ heterostructures therefore provides an ideal platform for probing how ReSe₂ anisotropy modulates WSe₂ properties. Fig. S3a presents the polarized Raman spectra of the monolayer ReSe₂ collected in both parallel (XX) and crossed (XY) configurations. Notice that some Raman modes exhibit pronounced intensity difference in two configurations. This is due to the Raman intensity being determined by the Raman tensors of vibration modes and the incident and scattered light polarization vectors.^{37,38} Particularly, each ReSe₂ vibration mode possesses a definite symmetry that dictates its Raman tensor non-zero elements, while the contribution of these elements to the scattered intensity is selectively weighted by the polarization configurations.^{37,38} Fig. S3b shows the polarized Raman spectra of the WSe₂/ReSe₂ heterostructure recorded in two configurations. Clearly, two vibration modes of WSe₂ exhibit distinct polarization responses: the E_{2g}^1 intensity drops markedly in the crossed configuration, whereas A_{1g} intensity remains almost unchanged.

Moreover, we performed the angle-resolved polarization Raman measurements of the monolayer ReSe₂. The sample is rotated from 0° to 180°, with a spectrum acquisition every 15° in both parallel and crossed configurations. The evolution of the polarized Raman spectra (XX) of the monolayer ReSe₂ is shown in Fig. 3a, while the corresponding intensity color map is provided in Fig. 3b for better visualization of the angle dependence of each phonon mode. For XX configuration, the frequencies of various Raman modes in monolayer ReSe₂ exhibit little change with angles, whereas their intensities present pronounced angle dependence, demonstrating the high in-plane anisotropy of our as-grown monolayer ReSe₂. To quantitatively characterize the anisotropy characteristic of ReSe₂, we performed detailed lorentzian fitting for the angle-resolved polarized Raman spectra to extract the intensity of each mode. Considering the 17 observed Raman-active modes are all first-order A_g symmetry, we primarily focused on two prominent peaks at 123 cm⁻¹ and 173 cm⁻¹. The normalized

intensities of the two modes as a function of rotation angle are presented in the polar diagrams of Fig. 3c and d, respectively. Clearly, the intensities of the ReSe₂ Raman modes at 123 cm⁻¹ and 173 cm⁻¹ vary periodically with rotation angle, with a period close to 90°. Owing to the different basis functions of the C_i point group, not all Raman modes exhibit a 90° period,³⁹ for example, the 261 cm⁻¹ mode shows a ~180° periodicity (see Fig. 3b). In addition, Fig. S4 presents the angle-resolved Raman responses of the as-grown monolayer ReSe₂ under XY polarization configuration. Similarly, the intensities of the 123 cm⁻¹ and 173 cm⁻¹ Raman modes exhibit significant angle dependence, which further indicates the strong in-plane anisotropy of monolayer ReSe₂.

Fig. 3e displays the evolution of the polarized Raman spectra (XX configuration) of the as-grown monolayer WSe₂, while the corresponding intensity color map is provided in Fig. 3f. The normalized intensities of the E_{2g}^1 and A_{1g} modes *versus* rotation angle are plotted in the polar diagrams of Fig. 3g and h. As expected, no obvious periodic variation is observed, consistent with the isotropic nature of monolayer WSe₂. Based on the preceding evidence for ultra-strong interlayer coupling in our CVD-grown WSe₂/ReSe₂ heterostructures, we anticipate that the pronounced in-plane anisotropy of ReSe₂ may induce a measurable anisotropy in the top-layer WSe₂. To verify the deduction, we further conducted angle-resolved polarization Raman measurements of the as-grown WSe₂/ReSe₂ heterostructures. Fig. 3i presents the angle-resolved polarized Raman spectra of the WSe₂/ReSe₂ heterostructure under parallel configuration, with the corresponding intensity color map shown in Fig. 3j. The WSe₂ E_{2g}^1 mode exhibits a pronounced period intensity modulation, whereas the A_{1g} signal is too weak for its periodicity to be directly observed. Likewise, lorentzian fitting was employed to extract the angle-dependent intensities of E_{2g}^1 and A_{1g} modes, whose normalized results are displayed in the polar diagrams of Fig. 3k and l, respectively. Interestingly, the two Raman modes of WSe₂ exhibit a consistent ~90° period, while the E_{2g}^1 mode demonstrates a stronger angle dependence relative to the A_{1g} mode. Particularly, E_{2g}^1 mode typically involves in-plane vibrations of W and Se atoms, making it more sensitive to in-plane symmetry variations than A_{1g} mode.²⁶ These experimental findings clearly suggest that the Raman modes of top-layer WSe₂ exhibit optical in-plane anisotropy in our CVD-grown WSe₂/ReSe₂ heterostructures, evidencing a lattice-symmetry perturbation induced within WSe₂ layer. We attribute this transformation to the strong interfacial coupling between WSe₂ and low-symmetry ReSe₂ substrate, which induces a pronounced anisotropic phonon-photon coupling behavior in the WSe₂ layer.

Specifically, monolayer WSe₂ is intrinsically in-plane isotropic, whereas its phonon scattering exhibits pronounced angle dependence (XX) as bonding with ReSe₂ *via* van der Waals forces. Such polarization-sensitive Raman scattering usually signals a modification of in-plane rotational symmetry.^{14,15} In other words, the ultra-strong interlayer coupling between monolayer WSe₂ and low-symmetry ReSe₂ breaks the intrinsic symmetry of WSe₂. Meanwhile, such strong interlayer coupling allows ReSe₂ to impose an anisotropic potential on WSe₂,



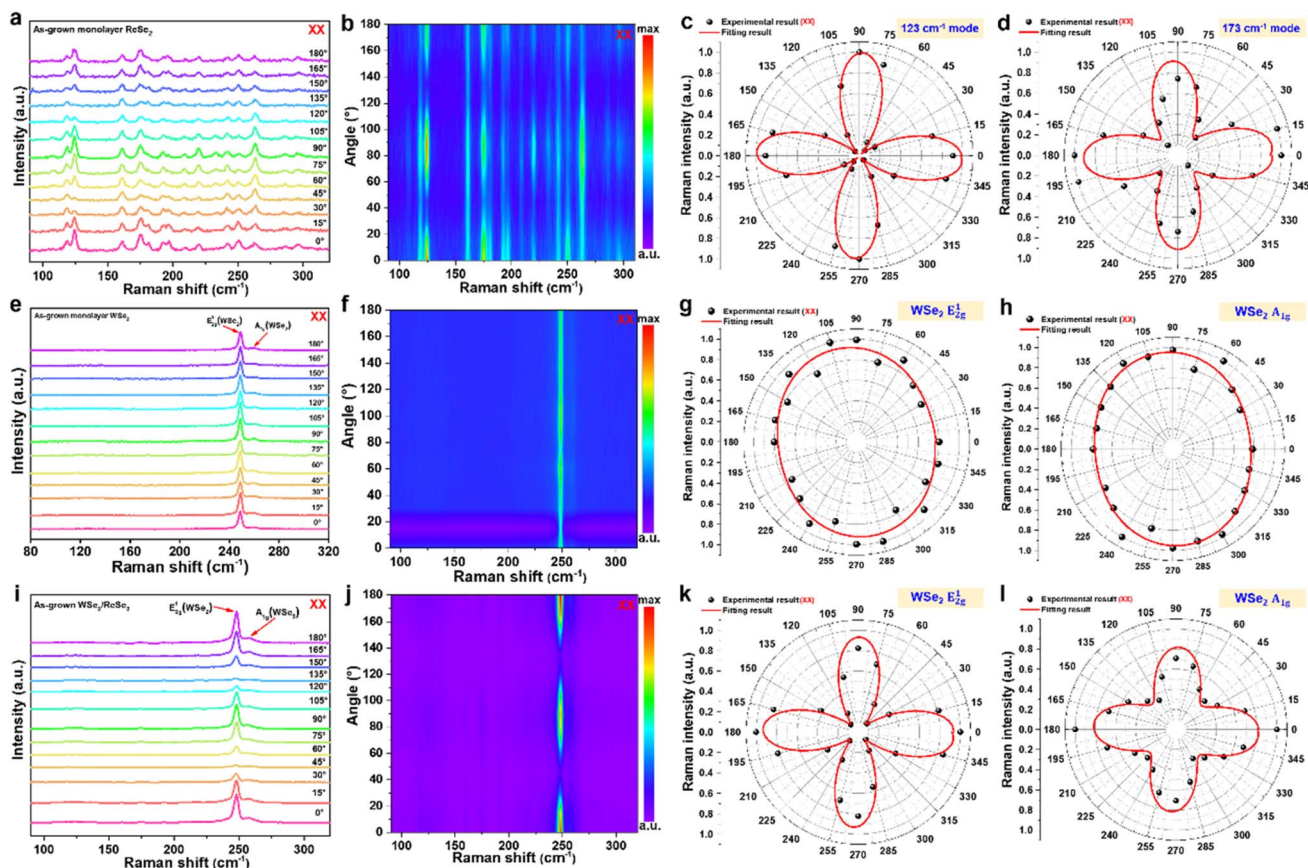


Fig. 3 Polarization Raman measurements of the CVD-grown monolayer ReSe₂, monolayer WSe₂, and WSe₂/ReSe₂ vertical heterostructure. (a) Angle-resolved polarized Raman spectra of the as-grown monolayer ReSe₂ under parallel configuration. (b) The Raman intensity color map corresponding to (a). (c) and (d) The normalized Raman intensities of ReSe₂ 123 cm⁻¹ and 173 cm⁻¹ modes as a function of rotation angle. (e) Angle-resolved polarized Raman spectra of the as-grown monolayer WSe₂ under parallel configuration. (f) The Raman intensity color map corresponding to (e). (g) and (h) The normalized Raman intensities of WSe₂ E_{2g}¹ and A_{1g} modes as a function of rotation angle. (i) Angle-resolved polarized Raman spectra of the as-grown WSe₂/ReSe₂ vertical heterostructure under parallel configuration. (j) The Raman intensity color map corresponding to (i). (k) and (l) In WSe₂/ReSe₂ heterostructure, the normalized Raman intensities of WSe₂ E_{2g}¹ and A_{1g} modes as a function of rotation angle.

driving its phonon scattering to become in-plane anisotropic. Moreover, the charge redistribution and band renormalization induced by efficient interlayer charge transfer may further produce a contribution to the anisotropy of phonon-phonon coupling behavior.⁴⁰ Particularly, Fig. S5 presents the angle-resolved polarized Raman spectra of the WSe₂/ReSe₂ heterostructure under XY configuration. In contrast to XX configuration, both Raman modes of WSe₂ exhibit weak angle dependence, with a period intensity modulation of ~180°, which implies that the anisotropy induced by strong interlayer coupling also follows the Raman tensor selection rule.

To further verify this conclusion, we fabricated an additional vertical WSe₂/ReSe₂ heterostructure through transferring a mechanically exfoliated (ME) monolayer WSe₂ onto the as-grown monolayer ReSe₂. Fig. 4a presents the optical image of the sample, where the yellow frame represents the heterostructure region. We respectively acquired the Raman spectra in the individual ReSe₂, WSe₂ and WSe₂/ReSe₂ heterostructure regions, as shown in Fig. 4b. In the heterostructure, both Raman peaks of WSe₂ are clearly visible, whereas the ReSe₂

peaks are too weak to be observed, consistent with the CVD-grown samples (see Fig. 2b). In contrast to the pronounced redshift observed in CVD-grown heterostructures, the unchanged peak position of WSe₂ indicates a relatively weak interfacial coupling in the ME WSe₂/ReSe₂ heterostructure. Fig. 4c directly compares the PL spectrum of monolayer WSe₂ on SiO₂/Si substrate with that of the ME WSe₂/ReSe₂ heterostructure at room temperature. Note that, the mechanically-assembled heterostructure sample had been annealed at 300 °C for 4 hours to strengthen interfacial coupling before optical measurements. As expected, the WSe₂ PL emission is incompletely quenched in the heterostructure, with a weak emission peak still visible (red curve). This incomplete PL quenching at room temperature suggests a relative weak interlayer coupling in the mechanically-assembled WSe₂/ReSe₂ heterostructure compared to that of the CVD-grown sample. Therefore, we further conducted angle-resolved polarization Raman measurements for the individual WSe₂ and ME WSe₂/ReSe₂ heterostructure, respectively. Fig. 4d-g display the intensity color maps of the angle-resolved Raman spectra for the



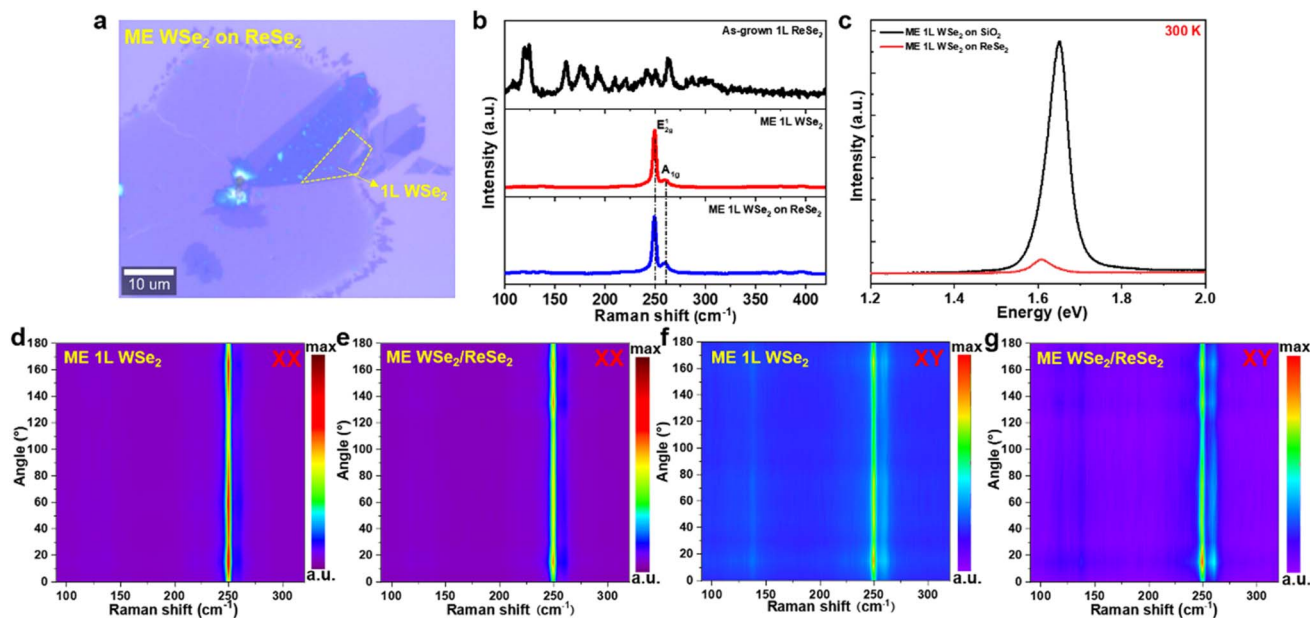


Fig. 4 Optical anisotropy characterization of the mechanically-assembled $\text{WSe}_2/\text{ReSe}_2$ heterostructure. (a) Optical image of the mechanically-assembled $\text{WSe}_2/\text{ReSe}_2$ heterostructure. (b) Raman spectra measured in the individual ReSe_2 , WSe_2 and $\text{WSe}_2/\text{ReSe}_2$ heterostructure regions. (c) Comparison of the PL spectra between the monolayer WSe_2 and ME $\text{WSe}_2/\text{ReSe}_2$ heterostructure. (d) and (e) Raman intensity color maps of the monolayer WSe_2 and ME $\text{WSe}_2/\text{ReSe}_2$ heterostructure under parallel configuration. (f) and (g) Raman intensity color maps of the monolayer WSe_2 and ME $\text{WSe}_2/\text{ReSe}_2$ heterostructure under crossed configuration.

monolayer WSe_2 and ME $\text{WSe}_2/\text{ReSe}_2$ heterostructure, recorded in both XX and XY polarization configurations. Intuitively, the WSe_2 Raman modes in the ME $\text{WSe}_2/\text{ReSe}_2$ heterostructure remain angle-independent, and their intensity evolution almost keep completely consistent with that of monolayer WSe_2 on SiO_2/Si . These experimental results indicate that there is no in-plane anisotropy is induced in the WSe_2 layer owing to the weak interlayer coupling in the mechanically-assembled $\text{WSe}_2/\text{ReSe}_2$ heterostructure. This finding further corroborates that the pronounced in-plane anisotropy of WSe_2 in CVD-grown $\text{WSe}_2/\text{ReSe}_2$ heterostructures stems from robust interlayer coupling.

4 Conclusions

In summary, high-quality $\text{WSe}_2/\text{ReSe}_2$ vertical heterostructures are grown by one-step CVD technique and exhibit ultra-strong interfacial coupling, evidenced by complete WSe_2 PL quenching. Angle-resolved polarized Raman spectroscopy reveals that the Raman modes of WSe_2 exhibit obvious in-plane anisotropy in CVD-grown $\text{WSe}_2/\text{ReSe}_2$ vertical heterostructures. This is attributed to the intrinsic symmetry breaking of WSe_2 induced by strong interlayer coupling between WSe_2 and low-symmetry ReSe_2 . Our results clarify how interfacial coupling affects anisotropy across van der Waals interfaces and provide a foundation for designing polarization-sensitive optoelectronic devices based on anisotropic TMDs heterostructures.

Author contributions

Chunxiao Cong, Min Jin and Xiaofei Yue directed the research work. Yong Guo, Xiaofei Yue and Jiajun Chen conceived and designed the experiments. Jiajun Chen fabricated the samples. Yong Guo and Xiaofei Yue conducted the experiments and analyzed the data. Yong Guo and Xiaofei Yue co-wrote the manuscript. All authors discussed the results and commented on the manuscript.

Conflicts of interest

There are no conflicts to declare.

Data availability

The data supporting this article have been included as part of the supplementary information (SI). Supplementary information: growth situation with/without NaCl, AFM height profiles, and XY -polarization Raman results. See DOI: <https://doi.org/10.1039/d5ra08161k>.

Acknowledgements

This work is supported by the National Natural Science Foundation of China (62374037, 52371193, 52272006), the Chengguang Program supported by the Shanghai Education Development Foundation & Shanghai Municipal Education Commission, the Shanghai Shuguang Program, the Shanghai Academic Research Leader (23XD1421200), the Shanghai



Rising-Star Program (23QA1403900), the Program for Professor of Special Appointment (Eastern Scholar) at Shanghai Institutions (TP2022122), the Shanghai Oriental Talented Youth Project, Space Application System of China Manned Space Program (KJZYY-NCL-0405), Zhejiang Provincial Natural Science Foundation of China (No. LD25E020001), Opening Project of State Key Laboratory of Functional Crystals and Devices (SKLFCD202505SIC).

Notes and references

- 1 X. Zhang, Z. C. Lai, Q. L. Ma and H. Zhang, *Chem. Soc. Rev.*, 2018, **47**, 3301–3338.
- 2 Y. Wang, S. Sarkar, H. Yan and M. Chhowalla, *Nat. Electron.*, 2024, **7**, 638–645.
- 3 J. H. Wang, X. Z. Xu, T. Cheng, L. H. Gu, R. X. Qiao, Z. H. Liang, D. D. Ding, H. Hong, P. M. Zheng, Z. B. Zhang, Z. H. Zhang, S. Zhang, G. L. Cui, C. Chang, C. Huang, J. J. Qi, J. Liang, C. Liu, Y. G. Zuo, G. D. Xue, X. J. Fang, J. P. Tian, M. H. Wu, Y. Guo, Z. X. Yao, Q. Z. Jiao, L. Liu, P. Gao, Q. Y. Li, R. Yang, G. Y. Zhang, Z. L. Tang, D. P. Yu, E. G. Wang, J. M. Lu, Y. Zhao, S. W. Wu, F. Ding and K. H. Liu, *Nat. Nanotech.*, 2022, **17**, 33–38.
- 4 S. Ghosh, A. Varghese, K. Thakar, S. Dhara and S. Lodha, *Nat. Commun.*, 2021, **12**, 3336.
- 5 M. Koperski, M. R. Molas, A. Arora, K. Nogajewski, A. O. Slobodeniuk, C. Faugeras and M. Potemski, *Nanophotonics*, 2017, **6**, 1289–1308.
- 6 X. Hu, L. Yan, L. Y. Ding, N. N. Zheng, D. Li, T. Ji, N. D. Chen and J. Q. Hu, *Coordin. Chem. Rev.*, 2024, **499**, 215504.
- 7 S. Joseph, J. Mohan, S. Lakshmy, S. Thomas, B. Chakraborty, S. Thomas and N. Kalarikkal, *Mater. Chem. Phys.*, 2023, **297**, 127332.
- 8 Y. C. Lin, H. P. Komsa, C. H. Yeh, T. Björkman, Z. Y. Liang, C. H. Ho, Y. S. Huang, P. W. Chiu, A. V. Krashenninnikov and K. Suenaga, *ACS Nano*, 2015, **9**, 11249–11257.
- 9 S. X. Yang, C. Wang, H. Sahin, H. Chen, Y. Li, S. S. Li, A. Suslu, F. M. Peeters, Q. Liu, J. B. Li and S. Tongay, *Nano Lett.*, 2015, **15**, 1660–1666.
- 10 E. F. Liu, Y. J. Fu, Y. J. Wang, Y. Q. Feng, H. M. Liu, X. G. Wan, W. Zhou, B. G. Wang, L. B. Shao, C. H. Ho, Y. S. Huang, Z. Y. Cao, L. G. Wang, A. D. Li, J. W. Zeng, F. Q. Song, X. R. Wang, Y. Shi, H. T. Yuan, H. Y. Hwang, Y. Cui, F. Miao and D. Y. Xing, *Nat. Commun.*, 2015, **6**, 6991.
- 11 F. F. Cui, X. B. Li, Q. L. Feng, J. B. Yin, L. Zhou, D. Y. Liu, K. Q. Liu, X. X. He, X. Liang, S. Z. Liu, Z. B. Lei, Z. H. Liu, H. L. Peng, J. Zhang, J. Kong and H. Xu, *Nano Res.*, 2017, **10**, 2732–2742.
- 12 R. X. Wu, H. M. Zhang, H. F. Ma, B. Zhao, W. Li, Y. Chen, J. T. Liu, J. Y. Liang, Q. Y. Qin, W. X. Qi, L. Chen, J. Li, B. Li and X. D. Duan, *Chem. Rev.*, 2024, **124**, 10112–10191.
- 13 M. M. Liu, S. H. Li, Y. K. Guo, L. Q. Zhang, D. Y. Shen, Q. Ye, Z. F. Peng, W. X. Qi, R. X. Wu, J. Li and X. D. Duan, *Adv. Funct. Mater.*, 2025, **35**, 2500876.
- 14 Z. Y. Li, J. W. Huang, L. Zhou, Z. Xu, F. Qin, P. Chen, X. J. Sun, G. Liu, C. Q. Sui, C. Y. Qiu, Y. F. Lu, H. Y. Gou, X. X. Xi, T. Ideue, P. Z. Tang, Y. Iwasa and H. T. Yuan, *Nat. Commun.*, 2023, **14**, 5568.
- 15 X. Xie, J. N. Ding, B. Wu, H. H. Zheng, S. F. Li, J. He, Z. W. Liu, J. T. Wang and Y. P. Liu, *Appl. Phys. Lett.*, 2023, **123**, 222101.
- 16 L. Ye, P. Wang, W. J. Luo, F. Gong, L. Liao, T. D. Liu, L. Tong, J. F. Zang, J. B. Xu and W. D. Hu, *Nano Energy*, 2017, **37**, 53–60.
- 17 S. W. Zhao, J. C. Wu, K. Jin, H. Y. Ding, T. S. Li, C. Z. Wu, N. Pan and X. P. Wang, *Adv. Funct. Mater.*, 2018, **28**, 1802011.
- 18 A. Usman, M. A. Aly, H. Masenda, J. J. P. Thompson, S. M. Gunasekera, M. Mucha-Kruczynski, S. Brem, E. Malic and M. Koch, *Nanoscale*, 2022, **14**, 10851–10861.
- 19 M. Zhao, W. T. Zhang, M. M. Liu, C. Zou, K. Q. Yang, Y. Yang, Y. Q. Dong, L. J. Zhang and S. M. Huang, *Nano Res.*, 2016, **9**, 3772–3780.
- 20 B. Wu, X. Xie, H. H. Zheng, S. F. Li, J. N. Ding, J. He, Z. W. Liu and Y. P. Liu, *Opt. Lett.*, 2023, **48**, 5867–5870.
- 21 X. F. Yue, J. J. Chen, J. K. Han, Y. B. Shan, S. W. Shen, W. X. Wu, B. J. Liu, L. J. Li, Y. Chen, R. J. Zhang, L. G. Hu, R. Liu, Z. J. Qiu and C. X. Cong, *Sci. China Mater.*, 2024, **67**, 3012–3020.
- 22 Y. B. Shan, X. F. Yue, J. J. Chen, B. G. M. Ekoya, J. K. Han, L. G. Hu, R. Liu, Z. J. Qiu and C. X. Cong, *ACS Appl. Nano Mater.*, 2022, **5**, 17986–17994.
- 23 M. Suleman, S. H. Lee, M. Kim, V. Nguyen, M. Riaz, N. Nasir, S. Kumar, H. M. Park, J. W. Jung and Y. Seo, *ACS Omega*, 2022, **7**, 30074–30086.
- 24 A. Singh, M. Moun, M. Sharma, A. Barman, A. K. Kapoor and R. Singh, *Appl. Surf. Sci.*, 2021, **538**, 148201.
- 25 D. Wolverson, S. Crampin, A. S. Kazemi, A. Ilie and S. J. Bending, *ACS Nano*, 2014, **8**, 11154–11164.
- 26 X. Zhang, X. F. Qiao, W. Shi, J. B. Wu, D. S. Jiang and P. H. Tan, *Chem. Soc. Rev.*, 2015, **44**, 2757–2785.
- 27 W. W. Li, Z. Q. Yang, M. T. Sun and J. Dong, *Rev. Phys.*, 2022, **9**, 100077.
- 28 T. F. Yan, X. F. Qiao, X. N. Liu, P. H. Tan and X. H. Zhang, *Appl. Phys. Lett.*, 2014, **105**, 101901.
- 29 X. X. Han, D. W. He, L. Zhang, S. C. Hao, S. Y. Liu, J. L. Fu, Q. Miao, J. Q. He, Y. S. Wang and H. Zhao, *J. Mater. Res.*, 2020, **35**, 1417–1423.
- 30 L. Zhang, D. W. He, J. Q. He, Y. Fu, A. Bian, X. X. Han, S. Y. Liu, Y. S. Wang and H. Zhao, *Opt. Express*, 2019, **27**, 17851–17858.
- 31 P. Zereshki, P. Yao, D. W. He, Y. S. Wang and H. Zhao, *Phys. Rev. B*, 2019, **99**, 195438.
- 32 H. Liu, Y. X. Zong, P. Wang, H. Y. Wen, H. B. Wu, J. B. Xia and Z. M. Wei, *J. Phys. D: Appl. Phys.*, 2021, **54**, 053001.
- 33 S. H. Chen, Z. J. Sun, H. Liu, H. W. Xu, C. Wang, R. Han, Z. H. Wang, S. C. Huang, X. L. Zhao, Z. K. Chen, W. Z. Li and D. M. Liu, *Nanoscale*, 2024, **16**, 14089–14095.
- 34 J. J. Chen, X. F. Yue, Y. B. Shan, H. S. Wang, J. K. Han, H. M. Wang, C. X. Sheng, L. G. Hu, R. Liu, W. H. Yang, Z. J. Qiu and C. X. Cong, *RSC Adv.*, 2023, **13**, 18099–18107.
- 35 J. Q. Zong, W. X. Ji, P. Li and P. J. Wang, *Mater. Res. Express*, 2020, **7**, 055018.



- 36 Q. H. Wang, K. Kalantar-Zadeh, A. Kis, J. N. Coleman and M. S. Strano, *Nat. Nanotech.*, 2012, 7, 699–712.
- 37 M. A. Pimenta, G. C. Resende, H. B. Ribeiro and B. R. Carvalho, *Phys. Chem. Chem. Phys.*, 2021, 23, 27103–27123.
- 38 G. C. Resende, G. A. S. Ribeiro, O. J. Silveira, J. S. Lemos, J. C. Brant, D. Rhodes, L. Balicas, M. Terrones, M. S. C. Mazzoni, C. Fantini, B. R. Carvalho and M. A. Pimenta, *2D Mater.*, 2021, 8, 025002.
- 39 Y. Choi, K. Kim, S. Y. Lim, J. Kim, J. M. Park, J. H. Kim, Z. Lee and H. Cheong, *Nanoscale Horiz.*, 2020, 5, 308–315.
- 40 I. Sánchez-Ramírez, M. G. Vergniory and F. de Juan, *Phys. Rev. B*, 2024, 110, 195138.

