Martensite is a needle-shaped microstructure formed by a rapid, diffusionless transformation and significantly affects the mechanical properties of materials. Here, in two-dimensional ReS$_2$ we show that martensite-like domain structures can form via a diffusionless transformation, involving small lattice deformations. By analyzing the strain distribution and topology of the as-grown chemical vapor deposition samples, we find that cooling-induced strain at the ReS$_2$/substrate interface is responsible for the mechanical loading and is essential for martensite-like domain formation. Meanwhile, the effect of cooling rate, flake size and substrate on the microstructures revealed the mechanical origin of the transformation. The strain-induced lattice reconstructions are rationalized and possibly lead to ferroelastic effects. In view of the strong anisotropy in electronic and optical properties in two dimensional materials like ReS$_2$, opportunities exist for strain-correlated micro/nanostructure engineering, which has potential use in next-generation strain-tunable devices.
Two-dimensional (2D) materials are attractive to the community for their unprecedented deformability and a variety of novel properties. Similar to the bulk semiconductors, extensive efforts have been devoted to the strain engineering on 2D materials. Including graphene and transition metal dichalcogenides (TMDs), the electronic band structures of 2D materials are prevalently sensitive to the external loading, leading to significant changes in electrical or optical properties. It is more intriguing for the anisotropic 2D materials such as rhenium disulfide (ReS$_2$), as the crystal lattice reconstructions and sub-domain structures can contribute an additional degree of freedom for control.

According to the elasticity and plastic theory, stressing the three-dimensional (3D) bulk materials can result in a certain deformation in crystal lattices. On the contrary, for 2D materials, which can be regarded as nearly ideal plane-stress deformation due to the vanishing thickness, the induced strain could have higher uncertainties. For example, the out-of-plane undulation/rippling in 2D materials is almost unavoidable, especially under compressive or shear stress. Such rippling is also intrinsic to stabilize the suspended 2D crystals. On the other hand, 2D materials can usually be stressed close to the ideal strength due to the exclusion of atomic defects, leaving the elasticity entering the deep nonlinear regime. Indeed, abundant theories and experiments on the 2D materials such as graphene and 2D TMDs have shown their excellent mechanical properties, which also justify their great potentials in future applications.

Compared to the isotropic 2D counterparts, less is known about the anisotropic 2D materials. It has been reported the 2D ReS$_2$ or ReSe$_2$ can undergo atomic lattice reconstruction under electron beam effect. The 2D ReS$_2$ after growth were also observed with stripe-like or needle-like domains and parallel domain boundaries, though the as-grown samples on the growing substrates without transfer have not been systematically investigated. Apart from that, the anisotropic atomic structures (viz. the directional Re atomic chains, see Fig. 1a) in these T'-phases of 2D TMDs have also been manifested to have anisotropic electrical and optical properties. The anisotropic high electron mobility and the hierarchical domain structure in ReS$_2$ may create new possibilities for the property modulation in 2D structures.

It was explained that the sub-domain structures in 2D ReS$_2$ exist with mirror (twin) boundaries. However, the origin of these domains is not clear yet. As we will discuss in the current paper, the rich-domain structures in 2D ReS$_2$ can be attributed to the atomic lattice reorganization under ultra-low strain. Here we will give a complete explanation for the origin of all the needle-like domain patterns with a complete list of domain boundary types. These observed domains are not originated during growth in the chemical vapor deposition (CVD) at high temperature (CVD-grown products are virtually single crystal for each flake from a single nuclei), rather they are attributed to the cool-down stress by sample-substrate thermal expansion differences. The rich-domain structures in 2D ReS$_2$ are formed by lattice reorganization during the cooling process, which resembles the formation of the needle-like microstructure of martensite by iron atomic rearrangement during the cooling process. Here the origin of “martensite-like” structures in anisotropic 2D ReS$_2$ is investigated.

**Results and discussion**

The martensite-like domain structures in 2D ReS$_2$. The monolayer and multilayer 2D ReS$_2$ samples in our experiments were synthesized either by the CVD method or mechanical exfoliation method (see “Methods” section). Similar to previous reports, the CVD-grown monolayer (1L) ReS$_2$ samples, irrespective of monolayer or multilayers, predominantly exhibited sub-domain structures and parallel stripe patterns under polarized optical microscopy (OM) (Fig. 1a). After the samples were wet-transferred (see “Methods” section) onto new substrates, the domain and stripe patterns persisted (Supplementary Fig. 1). Comparatively, the mechanically exfoliated 1L-ReS$_2$ samples did not show any domain patterns above (Fig. 1b). We have verified the above patterns using dynamic force microscopy on the as-grown samples (DFM, see “Methods” section), in the topography

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**Fig. 1 The martensite-like domain structures in 2D ReS$_2$.** a, b The color optical images under bright field, polarized 70° and 110° for CVD-grown 1L-ReS$_2$ transferred on 300 nm SiO$_2$/Si substrate, and mechanical exfoliated ReS$_2$ on 300 nm SiO$_2$/Si substrate, respectively. The inset shows the crystal structure model of monolayer (1L) ReS$_2$ with chain-like structures (dashed black lines). c, d The DFM phase and topography images of monolayer ReS$_2$ as grown on mica (ReS$_2$/mica). e The magnified DFM topography image corresponding to the white dashed square in d. f The line profiles corresponding to the black dashed lines in e.
images, the similar patterns can be clearly identified, while the phase images can further enhance the domain contrasts (Fig. 1c, the corresponding topography images show in Supplementary Fig. 2). Notably, the multilayer ReS₂ also possess similar sub-domain structures irrespective of the thickness (Supplementary Fig. 3). These domain structures should be attributed to the different orientated crystal grains of 2D ReS₂. Hence the DFM phase images can be understood—domain contrast came from the different molecular interactions between scanning tips and ReS₂ surfaces along different crystal orientations. However, the brighter contrast half (higher) in the DFM topography image implies there is partial delamination between the 2D flake and substrate (Fig. 1d–f). Particularly, for the triangle flakes, the needle-like or stripe-like patterns are more likely to form in the undelaminated half flakes (the darker halves in the DFM topography images) than the delaminated half (the brighter halves in DFM topography images) (Fig. 1e), which will be explained later.

The strain distribution of the as-grown ReS₂. Here we used the angle-resolved polarized Raman (ARPR) spectroscopy⁴⁹ to examine the 2D monolayer ReS₂ samples without transfer (which means they are measured on the growth substrates directly). The polarized optics in our Raman setup has been verified on 1L-MoS₂ prior to the measurement on ReS₂ (Supplementary Figs. 4 and 5). We have correlated the Raman and DFM results for the same ReS₂ flakes (CVD grown on fluorophlogopite mica) (see “Methods” section). Considering the anisotropic optical property of ReS₂ and the non-polarized Raman measurement is less sensitive to anisotropic structures, the ARPR measurement is imperative for the in-plane anisotropic ReS₂ (Fig. 2). In the DFM phase images, the brighter color contrast indicates the lattice reconstruction area and the darker color contrast indicates the delaminated area (Fig. 2a). The Raman peak at 3, 4, 5, and 6 (Fig. 2b) are the Re atoms in-plane vibration modes⁴⁰. Different vibration modes in 1L-ReS₂ are excited with different polarization directions. Besides, the ARPR intensity of peak 5 indicates the orientation of Re chain³⁴. As shown in Fig. 2c, d, the intensities of Raman peak 5 exhibit stark differences between the delaminated/undelaminated domains. The Raman intensity variance can be explained by the different strain levels. The crystal orientations influence the angle of reaching maximum peak intensity (Fig. 2e, f). Peaks 3, 5, and 6 of sub-domain I reach maximum intensity at polarized angles 138°, 70°, and 65° (Fig. 2e). It shows a decreasing trend. On the contrary, the peaks 3, 5, and 6 of sub-domain II reach the maximum intensity at polarized angles 108°, 135°, and 150° (Fig. 2f). It shows an increasing trend. The inverse trend indicates the different lattice orientations (Fig. 2e, f). However, as for the orientation of each vibration mode, there are still huge intensity differences. As shown in Fig. 2g–i, the maximum peak intensity at peaks 3, 5, and 6 of sub-domain I are ~120, ~250, and ~250% of sub-domain II. The lattice reconstructed area has stronger ARPR intensity than the delaminated area. The reduction of Raman intensity of ReS₂ is owing to the existed strain³⁴. Hence, the residual strain effects are verified by the difference of ARPR intensity at oriented directions. More strain is released in the lattice reconstructed area, leading to a higher ARPR intensity than the delaminated area.

Cooling-induced train at ReS₂/substrate interface. The delamination of CVD as-grown flakes usually occurs during the cooling down process, so that the cooling rate is modulated to investigate the cool-down stress (Fig. 3). The slow-cooling rate produces less domain patterns on the sample (Fig. 3a, d), while the fast-cooling rate leads to more and denser domain patterns on 1L-ReS₂ sample (Fig. 3c, f). Initially, the sub-domain structures observed under polarized 70° OM images (Fig. 3a–c) indicate the distinguishing Re chain orientations. Furthermore, the needle-like patterns are studied using the DFM topography and phase characterizations (Fig. 3d–f). The delaminated areas have brighter contrast (higher) in DFM topography images resulted from the release of the cool-down stress. On the other hand, the lattice reconstruction is the stress release of the undelaminated area. Apparently shown in DFM phase images, the delaminated area shows the darker contrast and the lattice reconstructed area shows the brighter contrast. With a slow-cooling rate, the interfacial stress can be released by a large area, continuous flake-substrate delamination, while with a fast-cooling rate, the interfacial stress tends to be released through local lattice reconstructions. Fast cooling exerts larger cooling stress due to less strain relaxation, providing larger tensile stress on the 1L-ReS₂. As a result, the fast-cooling ReS₂ flakes have more needle-like patterns.

The residual stress/strain is distinctly different in the original lattice (delaminated) and lattice reconstructed domains. More strain is released in the lattice reconstructed areas (Fig. 2), leading to lowered remaining stress in those areas. According to the DFM topography results, the magnitude of the partial delaminations in lattice reconstructed domains is within 0.1–0.2 nm. With delaminations, the lattice mismatch strains between ReS₂ and substrate upon cooling are released. According to the Poisson ratio effect (elasticity of membranes), if the tensile strain in the membrane is relaxed, the thickness of the membrane will increase, in agreement with our DFM results as well. On the other hand, it is intrinsic behavior for the different optical responses under polarized light along with the different lattice directions (or domains) in 2D ReS₂.⁷⁷ However, in our experiments, the contrast of ARPR spectroscopy and DFM topography results between different domains should be attributed to the contrasted residual strain levels in the pristine and reconstructed domains.

The strain-originated domain structure can also explain the absence of domain patterns in the mechanical exfoliated ReS₂ flakes. The mechanical exfoliated flakes mainly experienced out-of-plane stressing instead of in-plane stress which might induce domain structures. Except for the out-of-plane bending, no apparent in-plane straining that is required for mechanical twinning is applied during the mechanical exfoliation process⁴². In CVD-grown flakes, owing to the easier relief of interfacial stress in smaller flakes and larger residual stress remained in larger flakes, the smaller 1L-ReS₂ flakes usually possess remarkably less domain structures (Fig. 4a–f). It is also found enhanced/increased domain structures in 1L-ReS₂ grown on SiO₂/Si substrate, compared with fluorophlogopite mica and c-face sapphire. It is attributed to the larger thermal expansion coefficient (TEC) misfits of SiO₂/Si with 2D anisotropic materials (Fig. 4g–o and Supplementary Table 1). The substrates with lower TEC such as Si/SiO₂ will shrink much less than 1L-ReS₂ during cooling, hence induce a larger traction force. The CVD-grown ReS₂ samples after transfer onto another substrate (Supplementary Fig. 1) kept the original patterns after synthesis due to no additional loading is applied in the processing. For the multilayer specimens, the cooling down strain can be actually transferred by interlayer interactions (signature of T′ phases of TMD materials)⁴₃, inducing similar lattice switching behavior throughout all the thicknesses.

The sub-domains boundaries trigger by the above lattice reconstructions have been fully corroborated by our high-resolution scanning transition electron microscopy (STEM) observations (Fig. 5a–c and Supplementary Fig. 6), which should be nucleated from the free edges (lowest strain energy cost at the edges). In the case of CVD cooling down the process for 2D ReS₂ in our experiments, the biaxial thermal expansion difference
between ReS$_2$ and underlying substrates (mica or sapphire) created a tensile strain field in ReS$_2$. There are no clear boundary conditions for strain/stress, the ReS$_2$ flakes and substrates are engaged by the vdW interactions (can be rationalized as static frictions) at the interface. In contrast to the out-of-plane buckling under compressive cooling down strains in 2D WS$_2$,[44] the interactions between ReS$_2$ and substrates here can create the homogeneous in-plane tensile strain in 2D ReS$_2$, while the relative sliding and delamination between 2D flakes and the substrates prefer to start at the free edges (Fig. 5d, e). Assuming the frictional force (between ReS$_2$ and substrate) is constant (maintained at maximum) in final states near edges at room temperature, the tensile stress/strain can be assigned according to the location in the 2D flakes (Fig. 5f, g). Apparently, the normal stress pointing to the edges will decrease from the corners to the interiors along edges, causing the shear strain in the “wing” like structures. Therefore, the energy favorable twinning domains will be triggered from the corners, sweep over half of the edges, and be ceased at the middle points of edges, as we observed above. The delaminated half (brighter half in the DFM topography images, e.g., Fig. 1d–f) thus have less or even no sub-domain structures due to the relief of interfacial stress upon this first-step delamination, while the undelaminated half (darker half in the DFM topography images) will undergo further transformations by the tensile stress starting from the edges, forming the needle-like sub-domains in the darker half.

The lattice reconstructed domains thus can have stark differences in residual stress, especially between sample and substrates. Compared to the original lattice domains (delaminated sub-domains), the interfacial strains have been largely released during lattice reconstructions in reconstructed domains. The contrasts between original and reconstructed domains seen by the DFM topography images (Fig. 1e) and the polarized optical characterizations (Fig. 2b–i) can be attributed to the strain differences between reconstructed and pristine domains. Particularly, the small shear angle changes in lattice directions, for pristine and reconstructed domains, can also be clearly visualized on free edges in the DFM/optical images (Fig. 1a, c, d and 3). The tensile traction exerted on the monolayer ReS$_2$ samples will cause the stripe patterns along certain crystal directions (Figs. 1e and 4).

Since all of the twin boundaries shown in Fig. 6 have very low lattice mismatches thus low formation energies, in real samples we can found quite complicated domain structures consisting of all the possible domain boundaries, crossing with angles close to 30°, 60°, 90°, 120°, and 150° (Supplementary Fig. 7).

The structural and strain analysis of twin structures by lattice reconstructions in 1L-ReS$_2$. 2D ReS$_2$ has a monoclinic crystal

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**Fig. 2** The ARPR characterizations on CVD-grown 2D 1L-ReS$_2$. a The DFM phase image. b The single Raman spectra (polarized 0°) of 1L-ReS$_2$ at different sub-domains corresponding to the blue and red spots in c. c, d The polarized 0° and polarized 90° Raman peak 5 intensity mapping, respectively. e, f Polarization-dependent Raman intensity mappings corresponding to the blue and red spots in c, respectively. g–i Angle-dependent polar plots for sub-domain I and sub-domain II at peaks 3, 5, 6 (scatter) and fitted by Sine function (line), respectively.
Fig. 3 DFM topography and phase images of CVD-grown 2D ReS₂ with different cooling rates. a–c Polarized OM images of 1L-ReS₂/mica synthesized with the slow-cooling rate (5 °C min⁻¹), normal slow-cooling rate (25 °C min⁻¹), and fast-cooling rate (120 °C min⁻¹), respectively. d–f DFM topography image and phase image corresponding to samples in a–c, respectively. g–i Line profiles corresponding to dash black lines in d–f.

Fig. 4 Size dependence and substrate effects. a, b The DFM topography image and phase image of 20 µm 1L-ReS₂ grown on fluorophlogopite mica. c, d The DFM topography image and phase image of 10 µm 1L-ReS₂ grown on fluorophlogopite mica. e, f The DFM topography image and phase image of 5 µm 1L-ReS₂ grown on fluorophlogopite mica. g–i, j–l, m–o The optical micrographs of CVD-grown ReS₂ on 300 nm SiO₂/Si, c-face sapphire, and fluorophlogopite mica substrates, respectively.
structure or called $T'$ (tetrahedral) phases.$^{39,45}$ This structure is flexible to switch either two of the base vectors ($a, b, i$) under ultra-
low straining (see Fig. 6a, b, and Supplementary Fig. 8), the 1L-
ReS$_2$ structures are relaxed by density functional theory (DFT)
calculations, see "Methods" section. We have listed all the possible normal strains and shear strains that can result in the lattice
switching (viz. lattice reconstruction) in monolayer ReS$_2$
(Fig. 6c–n). For the three types of twin boundaries (Fig. 6c, d, e),
the twin boundary directions are always in 0° or 60° with respect
to the zigzag edges of 2D ReS$_2$, while for the other three types of
twin boundaries (Fig. 6i, j, k), the twin boundary directions are in
30° or 90° with respect to the zigzag edges of 2D ReS$_2$. Therefore,
we can readily distinguish these two groups of twin boundaries
between the DFM or polarized OM images. It should be noted here
that the strain condition needs to be comprehensively understood
with the domain structures resulted from lattice reconstructions,
and the strain effect for these lattice reconstructions are not
relevant to the exact atomic structures of the domain boundaries,
but only determined by the original (pristine) and the recon-
structed domains. In principle, there is no strain along all the
twinning domain boundaries. However, in other directions, the
lattice reconstructions can accommodate the in-plane normal strain (either compressive or tensile) from $-1.5\%$ to $2.5\%$, and the in-plane shear strain (in terms of shear angle for certain directions) from $-1.6^\circ$ to $3.4^\circ$ (Fig. 6c–n), depending on the
orientation of twin boundaries (along $a, b, i, u, v, w$, respectively, see Fig. 6a for the definition).

Apparently, the domain boundaries prefer to be twinning
boundaries for the lowest energy cost (least lattice mismatch and
zero strain along twin boundary directions). The remarkable low shear or normal strain (shear angle $<1^\circ$) that absorbed by twinning are associated with the lattice reconstruction in such
anisotropic 2D materials. The parallel needle-like or stripe-like
domain structures well resemble the famous martensite phases$^{46}$
widely existed, however, the energy required for twinning here are
much lower than the normal mechanical twinning. Therefore,
such lattice reconstruction can act as an efficient deformation
mechanism for low straining level in 2D ReS$_2$. Classified by
boundary angles with respective to the ReS$_2$ edges, the triangular
domain structures (twin boundary in 90° or 30° with edges) and
the parallel stripe domains (twin boundary in 60° with edges) can
be correlated with the deformation under shear and normal stress
in samples, respectively (Fig. 6o, p). In addition, it should be
noted that if the sulfur (S) atoms are considered, the lattice
symmetry normal to the basal plane needs to be counted. That is,
the S atoms cannot penetrate the basal plane during lattice
reconstructions, so only the cases shown in Fig. 6c–e are entirely
twin boundaries (for both Re and S atoms), whereas the cases in
Fig. 6i–k are twin boundaries for Re atoms, but inverse twin
boundaries for S atoms.

Fig. 5 The STEM images and illustration of strain field in ReS$_2$. a–c The STEM annular dark field (ADF) images showing the typical twinning domain
boundary structures stemming from free edges, the diamond chain directions (b direction in Fig. 6) are highlighted, scale bars =1 nm. d The tensile
tractions on ReS$_2$ flake exerted by the engaged substrate. e The sketch map of the tensile strain level, decreasing from the center to flake edges. f The
mechanical schematic of tensile strain resulted interfacial sliding and traction forces in equilibrium. g The diagram illustration traction and strain level
associated with the location in the sample flakes.
Next, opposite to the tensile strain discussed above, we will focus on the effects of compressive stress on the 2D anisotropic materials. The biaxial compressive strain in 2D materials can be generated by flexible substrates in the previous works\textsuperscript{47}. Albeit the wrinkle patterns can be frequently seen in 2D materials strained by the underlying flexible substrates\textsuperscript{48,49}, mechanical approach is less controllable, particularly for biaxial straining. Here we applied ultraviolet (UV) exposure with moisture condition on the 2D ReS\textsubscript{2} samples (see “Methods” section for details). Unlike the previous graphene or isotropic TMD materials\textsuperscript{50,51}, the regular wrinkle patterns emerged in 2D ReS\textsubscript{2}. As seen from the DFM characterization results, the wrinkles preferentially follow the low-index directions (Fig. 7 and Supplementary Figs. 9, 10). The wavelengths, wrinkle heights for these three main directions are maintained almost identical throughout all the subdomains in the flake (originated from CVD cooling) (Fig. 7e, f). Therefore, the wrinkle patterns also exhibit the domain patterns following the crystals. Apart from the preferences in crystal orientations, the wrinkles are prone to be formed along the domain boundaries. More wrinkle patterns generated can be found in Supplementary Fig. 10.

The compressive strain above is supplied by the photochemistry relevant to the surface of ReS\textsubscript{2} (see “Methods” section). The oxygen and hydroxyl radicals introduced by UV treatment are able to anchor in the lattice space lead to the compressive strain in the lattice. Supplementary Fig. 11a–c shows the TEM

![Figure 6: The structural and strain analysis of twin structures by lattice reconstructions in 1L-ReS\textsubscript{2}](image-url)
characterization result of the spontaneous 2D wrinkling in monolayer ReS$_2$ under the biaxial compressive field. The wrinkles formation during UV treatment tends to follow the energy favorable directions—along basis vectors $a$ or $b$. As shown in Supplementary Fig. 11d, the expansion of lattices after UV oxidation is confirmed by our DFT simulations. The basal plane is expanded during UV treatment meanwhile the ReS$_2$ are constrained by the underlying substrates, subsequently, the wrinkles can be developed to release the stress. The atomic structures of 1L-ReS$_2$ can be slightly distorted with the exotic covalently-bonded surface oxygen atoms$^{41}$. The ReS$_2$ fold (Supplementary Fig. 11e) is also in part owing to the stacking of two counter layers after folding (the stable wrinkles can be reckoned as vertical folds$^{52,53}$, along these directions has the lowest interlayer energies. It should also be noted that due to the monoclinic structure of 1L-ReS$_2$, the folded two layers (in mirror symmetry) cannot match each other (rotate by 1–2 degrees) with the original lattice. Therefore, to minimize the stacking energy of the wrinkles, lattice switching can take place and accommodate the strains (Supplementary Fig. 11f).

The wrinkles along the grain boundaries (by merging/stitching of grains during growth from different nuclei) (Fig. 7e, f) are also clearly observed. The wrinkle structures close to the edges are related to the special strain distribution along the edges (almost zero strain parallel to edges), hence the remaining compressive strain perpendicular to edges will generate the perfect periodic one-dimensional wrinkle patterns. Close to the edges, there are always less interactions from the substrate effects, rendering the wavelength/period of the wrinkles also become larger at edges (Supplementary Fig. 10).

According to the above experiments and explicit analysis, it has been clarified that the strain effects play essential roles in the formation of martensite-like structures and the twinning domain boundaries in anisotropic 2D materials such as ReS$_2$ and ReSe$_2$. These domain structures can conversely have a significant impact on the mechanical behavior as well as other physical properties.

The rich atomic structural reconstructions under strain loading, and the associated strong anisotropy in physical properties (electrical, electronic, optical, etc.), could open great opportunities for future strain engineering on these anisotropic 2D materials.

Methods

Synthesis of ReS$_2$ on fluorophlogopite mica, c-face sapphire and 300 nm SiO$_2$/Si, MoS$_2$ on 300 nm SiO$_2$/Si substrates. 1L-ReS$_2$ and MoS$_2$ flakes are synthesized in a two-splitting heating center tubular furnace. Firstly, 2 mg ammonium persulfate (NH$_4$ReO$_4$) (Aldrich, 99.999%) powder for ReS$_2$ growth, sodium molybdate dihydrate (Na$_2$MoO$_4$·2H$_2$O) (Aldrich, 99%) for MoS$_2$ growth is placed in a quartz boat and covered by a piece of 1 cm * 1 cm fluorophlogopite mica (K$_2$Mg$_3$AlSi$_3$O$_10$F$_2$), c-face sapphire or 300 nm SiO$_2$/Si substrate then located on the downstream heating center. Next, 100 mg sulfur (Aldrich, 99.998%) plates for ReS$_2$ and MoS$_2$ growth are placed in a separated quartz boat located on the upstream heating center. As the downstream heating center ramped up to 850 and 800 °C, for ReS$_2$ and MoS$_2$, growth, respectively. Meanwhile, the upstream heating center climbed to 200 °C. The 80 sccm Argon carrier gas is introduced during synthesis. After maintaining the target temperature for 10 min, the furnace is cooled down naturally.

Transfer of ReS$_2$ sample on SiO$_2$/Si, TEM grid. The as-grown ReS$_2$ was spin-coated with polymethyl methacrylate (PMMA) (A4) to form PMMA/ReS$_2$/substrate structure. Next, the PMMA/ReS$_2$/substrate is floated on the 75 °C ultrapure water for one hour to detach PMMA/ReS$_2$ from the substrate. Following, a 300 nm SiO$_2$/Si substrate or Quantifoil™ TEM grid is applied to scoop out the PMMA/ReS$_2$ film. Subsequently, the PMMA/ReS$_2$ on the target substrate was dried under ambient temperature to increase the adhesion between ReS$_2$ and the target substrate. Finally, the acetone or acetone vapor is introduced to remove the PMMA layer on 300 nm SiO$_2$/Si substrate or on the TEM grid, respectively.

Prepare of ReS$_2$ on SiO$_2$/Si from a bulk sample. With the assistance of the semiconductor transfer tape (USI, Blue Adhesive Plastic Film), bulk ReS$_2$ (2D Semiconductors Inc., USA) was exfoliated into pieces and attached to a 300 nm SiO$_2$/Si substrate.

UV treatment method. 1L-ReS$_2$ on the substrate is placed into a chamber fitted out a mercury lamp (LH-arc, Lichtten Co. Ltd, Korea) with ~90% of emitted light at a wavelength of 254 nm and the minority of light at a wavelength of 185 nm (20 mW cm$^{-2}$). The chamber is connected with two injection pipes and one output

![Fig. 7 The results of DFM characterizations and wrinkling behavior of 1L-ReS$_2$ as grown on fluorophlogopite mica substrate after 180 s UV treatment.](image-url)
one injection pipe is for a humidifier to introduce moisture, while another one is for N₂ gas to extrude the humid and the output valve is used to balance the pressure at ambient pressure. The humid level was monitored using a hygro-thermometer (accuracy ± 3%). The 1L-RE₆ is treated with UV light for 180 s. The radicals were generated by UV light treated moisture air as the following reactions:

\[
\begin{align*}
\text{O}_3 & \rightarrow \text{O}_2 + \text{O} + \text{O}_2, \\
\text{O}_2 + \text{H}_2\text{O} & \rightarrow 2\text{OH}^+ + 3\text{O}_2, \text{H}_2\text{O} \rightarrow \text{OH}^- + \text{H}^+.
\end{align*}
\]

Angle-resolved polarized Raman (ARPR) measurement. The ARPR measurement is performed using a commercial inVia confocal Raman microscope (Renishaw, UK) with a ≤0.75/N.A. objective (LEICA, German) with a spatial resolution ~1 µm. The ARPR of ReS₂ and MoS₂ are measured at excitation wavelengths of 785 nm (~1.6 mW power) and 514 nm laser (~0.54 mW power) with 12001 mm⁻² grating and 18001 mm⁻² grating with a ≤0.50/0.75/N.A. objective (LEICA, German), respectively. The laser power is calibrated using standard photodiode power sensor Si121C (Thorlabs, USA). The ARPR measurement setup is equipped with a rotatable Polarizer I, one half-wave plate, and a linear Polarizer II, the setup shown in Supplementary Fig. 4. The ARPR spectra are recorded every 10° with the incident laser being rotated by Polarizer I from 0° to 360° while the sample is fixed. As analyzer, the half-wave plate and the linear Polarizer II are applied to collect the scattered light. The ARPR spectra were fitted using Gaussian contributions to extract the peak intensity value.

Data availability

All data sets needed to evaluate the conclusions are presented in the paper. Additional data related to this paper may be requested from the authors for educational and academic purposes.

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References

1. Lee, C., Wei, X. D., Kysar, J. W. & Hone, J. Measurement of the elastic properties and intrinsic stress of monolayer graphene. Science 321, 385–388 (2008).
2. Androulidakis, C., Koukaras, E. N., Paterekis, G., Trakakis, G. & Galiotis, C. Tunable macrosscale structural superlubricity in two-layer graphene via strain engineering. Nat. Commun. 11, 1–11 (2020).
3. Qi, Y. et al. Superconductivity in Weyl semimetal candidate MoTe₂. Nat. Commun. 7, 1–7 (2016).
4. Zhang, X. et al. Two-dimensional MoS₂-enabled flexible rectenna for Wi-Fi band wireless energy harvesting. Nature 566, 368–372 (2019).
5. Lee, G. H. et al. High-strength chemical-vapor-deposited graphene and grain boundaries. Science 340, 1073–1076 (2013).
6. Liu, Z. et al. Strain and structure heterogeneity in MoS₂ atomic layers grown by chemical vapour deposition. Nat. Commun. 5, 1–9 (2014).
7. Ni, Z. H. et al. Uniaxial strain on graphene: Raman spectroscopy study and band-gap opening. ACS Nano 2, 2301–2305 (2008).
8. Lloyd, D. et al. Band gap engineering with ultralarge breathing radii in suspended monolayer MoS₂. Nano Lett. 16, 5836–5841 (2016).
9. Shi, H. L., Pan, H., Zhang, Y. W. & Yakobson, B. I. Quasiparticle band structures and optical properties of strained monolayer MoS₂ and WS₂. Phys. Rev. B 87, 155304 (2013).
10. Ho, C. H. Optical study of the structural change in ReS₂, single crystals using polarized thermoreflectance spectroscopy. Opt. Express 13, 8–19 (2005).
11. Yu, S. et al. Strain-engineering the anisotropic electrical conductance in ReS₂ monolayer. Appl. Phys. Lett. 108, 191901 (2016).
12. Zhou, Z. et al. Anisotropic Raman scattering and mobility in monolayer 1T-ReS₂ controlled by strain engineering. Appl. Surf. Sci. 404, 276–281 (2017).
13. Li, J., Shan, Z. W. & Ma, E. Elastic strain engineering for unprecedented materials properties. Mrs Bull. 39, 108–117 (2014).
14. Novoselov, K. S. et al. Two-dimensional atomic crystals. Proc. Natl. Acad. Sci. USA 102, 10451–10453 (2005).
15. Shin, B. G. et al. Indirect band-gap puddles in monolayer MoS₂ by substrate-induced local strain. Adv. Mater. 28, 9378–9384 (2016).
16. Bao, W. Z. et al. Controlled ripple texturing of suspended graphene and ultrathin graphite membranes. Nat. Nanotechnol. 4, 562–566 (2009).
17. Luo, S. et al. Formation of ripples in atomically thin MoS₂ and local strain engineering of electrostatic properties. Nanotechnology 26, 105705 (2015).
18. Cadelano, E., Pallà, P. L., Giordano, S. & Colombo, L. Nonlinear elasticity of monolayer graphene. Phys. Rev. Lett. 102, 235502 (2009).
19. López-Polín, G. et al. Increasing the elastic modulus of graphene by controlled defect creation. Nat. Phys. 11, 26–31 (2015).
20. Li, T. S. Ideal strength and phonon instability in single-layer MoS₂. Phys Rev B 85, 235407 (2012).
21. Guinea, F., Katsnelson, M. I. & Vozmediano, M. A. H. Midgap states and charge inhomogeneities in corrugated graphene. Phys. Rev. B 77, 075422 (2008).
22. Yu, S. et al. Strain-engineering the anisotropic electrical conductance in ReS₂ monolayer. Appl. Phys. Lett. 108, 191901 (2016).
23. Singh, A. & Waghamare, U. V. Structural instabilities and wrinkles at the grain boundaries in 2-D h-BN: a first-principles analysis. Phys. Chem. Chem. Phys. 16, 21664–21672 (2014).
24. Li, Z. et al. Efficient strain modulation of 2D materials via polymer encapsulation. Nat. Commun. 11, 1–8 (2020).
25. Lee, J. E., Ahn, G., Shim, J., Lee, Y. S. & Ryu, S. Optical separation of mechanical strain from charge doping in graphene. Nat. Commun. 3, 1–8 (2012).
26. Lin, Y. C. et al. Single-layer ReS₂: two-dimensional semiconductor with tunable in-plane anisotropy. ACS Nano 9, 11249–11257 (2015).
27. Li, X. et al. Nanoassembly growth model for subdomain and grain boundary formation in 1T’-layered ReS₂. Adv. Funct. Mater. 29, 1906385 (2019).
28. He, X. et al. Chemical vapor deposition of high-quality and atomically layered ReS₂, SmI₂, SnI₂, and SnI₃. Nat. Mater. 11, 5423–5428 (2015).
29. Ho, C. H., Huang, Y. S., Tiong, K. K. & Liao, P. C. In-plane anisotropy of the optical and electrical properties of layered ReS₂ crystals. J. Phys. Condens. Matter. 11, 5367–5375 (1999).
30. Zhong, H. X., Gao, S. Y., Shi, J. J. & Yang, L. Quasiparticle band gaps, excitonic effects, and anisotropic optical properties of the monolayer distorted 1T’-graphene and ReS₂. Phys. Rev. B 92, 115438 (2015).
31. Sim, S. et al. Selectively tunable optical stark effect of anisotropic excitons in atomically thin ReS₂. Nat. Commun. 7, 1–6 (2016).
32. Liu, E. et al. Integrated digital inverters based on two-dimensional anisotropic ReS₂ field-effect transistors. Nat. Commun. 6, 1–7 (2015).
33. Ovchinnikov, D. et al. Disorder engineering and conductivity dome in ReS₂ with electrolyte gating. Nat. Commun. 7, 1–7 (2016).
34. Wu, K. et al. Domain architectures and grain boundaries in chemical vapor deposited highly anisotropic ReS₂ monolayer films. Nano Lett. 16, 5888–5894 (2016).
35. Speer, J., Matlock, D. K., De Cooman, B. C. & Schroth, J. G. Partitioning into anisotropic surface martensite transformation. Acta Mater. 51, 2611–2622 (2003).
36. Kitahara, H., Ueji, R., Tsuji, N. & Minamino, Y. Crystallographic features of lath martensite in low-carbon steel. Acta Mater. 54, 1274–1288 (2006).
37. Li, X. et al. Controlled growth of large-area anisotropic ReS₂ atomic layer and its photodetector application. Nanoscale 8, 18956–18962 (2016).
38. Ottaviano, L. et al. Mechanical exfoliation and layer number identification of MoS₂ revisited. 2D Mater. 4, 045013 (2017).
39. Hafeez, M., Gan, L., Li, H. Q., Ma, Y. & Zhai, T. Y. Chemical vapor deposition synthesis of ultrathin hexagonal ReS₂ flakes for anisotropic Raman property and optoelectronic application. Adv. Mat. 28, 8296–8301 (2016).
40. Tongay, S. et al. Monolayer behaviour in bulk ReS₂ due to electronic and vibrational decoupling. Nat. Commun. 5, 1–6 (2014).
41. Huang, L. et al. Redox photochemistry on van der Waals surfaces for reversible doping in 2D materials. Adv. Funct. Mater. 14, 2009166 (2021).
42. Yi, M. & Shen, Z. A review on mechanical exfoliation for the scalable production of graphene. J. Mater. Chem. A 3, 11700–11715 (2015).
43. Yang, J. et al. Elastic and electronic tuning of magnetoresistance in MoTe2. Sci. Adv. 3, eaa04949 (2017).
44. Ly, T. H., Yun, S. J., Thi, Q. H. & Zhao, J. Edge delamination of monolayer transition metal dichalcogenides. ACS Nano 11, 7534–7541 (2017).
45. Kertész, M. & Hoffmann, R. Octahedral vs. trigonal-prismatic coordination and clustering in transition-metal dichalcogenides. J. Am. Chem. Soc. 106, 3453–3460 (1984).
46. Calcagnotto, M., Ponge, D., Demir, E. & Raabe, D. Orientation gradients and geometrically necessary dislocations in ultrafine grained dual-phase steels studied by 2D and 3D EBSD. Mater. Sci. Eng. A 527, 2738–2746 (2010).
47. Zhu, C. R. et al. Strain tuning of optical emission energy and polarization in monolayer and bilayer MoS2. Phys. Rev. B 88, 121501 (2013).
48. Tweedie, M. E. P. et al. Inhomogeneous strain release during bending of WS2 on flexible substrates. ACS Appl. Mater. Inter. 10, 39177–39186 (2018).
49. Brennan, C. J., Nguyen, J., Yu, E. T. & Lu, N. Interface adhesion between 2D Materials and elastomers measured by buckle delaminations. Adv. Mater. Interfaces 2, 1500176 (2015).
50. Zhu, W. et al. Structure and electronic transport in graphene wrinkles. Nano Lett. 12, 3431–3436 (2012).
51. Zhang, Q. et al. Strain relaxation of monolayer WS2 on plastic substrate. Adv. Funct. Mater. 26, 8707–8714 (2016).
52. Zheng, F. et al. The critical stable length in wrinkles of two-dimensional materials. ACS Nano 14, 2137–2144 (2020).
53. Zhao, J. et al. Two-dimensional membrane as elastic shell with proof on the folds revealed by three-dimensional atomic mapping. Nat. Commun. 6, 1–6 (2015).
54. Kresse, G. & Furthmüller, J. Efficient iterative schemes for ab initio total-energy calculations using a plane-wave basis set. Phys. Rev. B 54, 11169 (1996).
55. Kresse, G. & Furthmüller, J. Efficiency of ab-initio total energy calculations for metals and semiconductors using a plane-wave basis set. Comput. Mater. Sci. 6, 15–50 (1996).
56. Blöchl, P. E. Projector augmented-wave method. Phys. Rev. B 50, 17953 (1994).
57. Perdew, J. P., Ernzerhof, M. & Burke, K. Rational for mixing exact exchange with density functional approximations. J. Chem. Phys. 105, 9982–9985 (1996).
58. Perdew, J. P., Burke, K. & Ernzerhof, M. Generalized gradient approximation made simple. Phys. Rev. Lett. 77, 3865–3868 (1996).

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Author contributions
All authors were involved in data interpretation and manuscript preparation. L.H., F.Z., and H.C. conducted the experiments, performed the formal analysis, and original manuscript writing. Q.H.T., X.C., and H.L. performed the materials synthesis and characterizations. C.-S.L. consultation. T.H.L., J.Z., and Q.D. supervised the project, manuscript edit, and funding acquisition.

Competing interests
The authors declare no competing interests.

Additional information

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