

2D Materials



PAPER

Epitaxial growth of single-orientation high-quality MoS₂ monolayers

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Abstract

We present a study on the growth and characterization of high-quality single-layer MoS₂ with a single orientation, i.e. without the presence of mirror domains. This single orientation of the MoS₂ layer is established by means of x-ray photoelectron diffraction. The high quality is evidenced by combining scanning tunneling microscopy with x-ray photoelectron spectroscopy measurements. Spin- and angle-resolved photoemission experiments performed on the sample revealed complete spin-polarization of the valence band states near the K and -K points of the Brillouin zone. These findings open up the possibility to exploit the spin and valley degrees of freedom for encoding and processing information in devices that are based on epitaxially grown materials.

1. Introduction

Novel two-dimensional materials may form the basis of future electronic devices that exploit the valley [1–5] and spin [6, 7] degrees of freedom. Single-layer (SL) transition metal dichalcogenides (TMDCs) are particularly promising for such applications because, unlike graphene, their structure breaks inversion symmetry and integrates atoms with a strong spin-orbit interaction. However, exploiting these new degrees of freedom in an actual electronic device requires a distinction between the K and -K points of the Brillouin zone and thus a single orientation of the layer. Current methods of chemical vapour deposition have not been able to achieve this and have produced mirror twin domains [8–10]. Here we report a protocol for the synthesis of SL MoS₂ of a *single* domain orientation. We demonstrate the structural properties of the MoS₂ layer using photoelectron diffraction and we measure the complete spin polarization of the valence band states near K and -K by spin- and angle-resolved photoemission spectroscopy.

Early successes in fabricating electronic devices based on single-layer transition metal dichalcogenides took advantage of the direct band gap in SL MoS₂ [11] and WS₂, which guarantees large on-off current ratios in field effect transistors [12–17] and permits optical applications not attainable in the indirect band gap parent materials [18–24]. To realize this, large flakes of high quality materials are desirable and the presence of differently oriented domains is not a fundamental limitation, apart from the extended defects induced by the presence of domain boundaries. The exploitation of the valley [1–4] and spin degrees of freedom [6, 7, 25–27], on the other hand, requires a specific orientation of the material's unit cell and thus the absence of mirror domains. This is illustrated in figure 1(a), which shows the unit cell and electronic structure of SL TMDC mirror domains, illustrating the spin-reversal in the valence band maxima near K and -K. For a simultaneous presence of two twin domains, the spin and valley polarization is lost on average and the observation of a valley Hall effect is prevented.

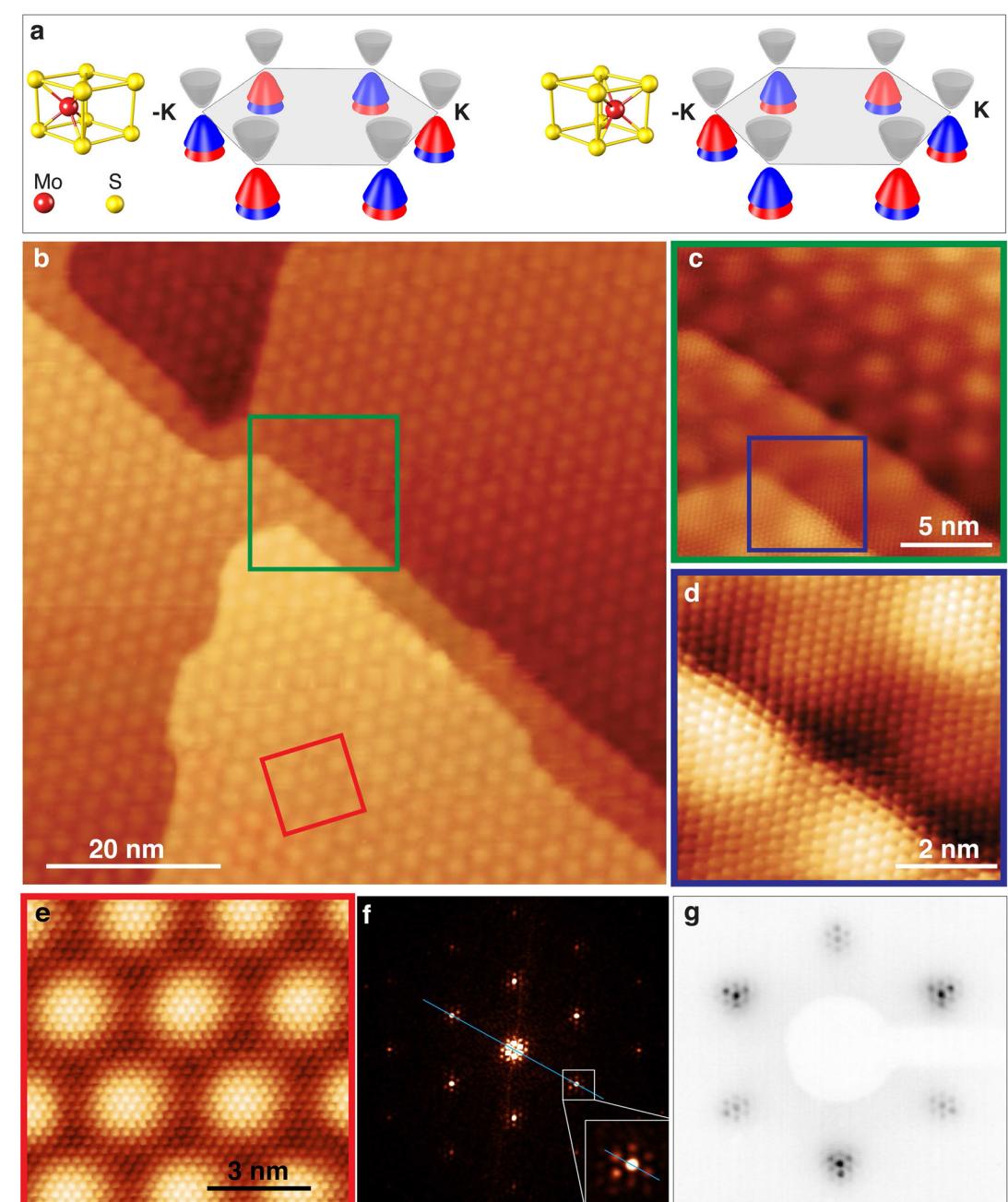


Figure 1. STM and LEED characterization of single-orientation SL MoS₂. (a) Structure, Brillouin zone and schematic band structure for two mirror domain orientations of SL MoS₂. The colors of the split valence band maximum (blue/red) refer to the different spin orientation of these states. (b) STM topography acquired on a large area $V_T = 0.525\text{ V}$, $I_T = 1.04\text{ nA}$. (c) and (d) STM images acquired close to an atomic step of the Au substrate (image c: $V_T = 0.525\text{ V}$, $I_T = 1.05\text{ nA}$, image d: $V_T = 0.525\text{ V}$, $I_T = 1.05\text{ nA}$). The different areas probed are framed in different colors. (e) STM image ($V_T = 0.525\text{ V}$, $I_T = 0.89\text{ nA}$) acquired on a Au terrace (red frame) and (f) corresponding FFT analysis of the image shown in (e). A detail of the framed spot is magnified in the bottom inset. (g) LEED pattern ($E_p = 185\text{ eV}$).

While van der Waals epitaxy of SL TMDCs on weakly interacting substrates such as sapphire [14, 15], silicon oxide [28] and graphene [29] yields an angular distribution of domain orientations, highly crystalline films are achieved on h-BN by using very high growth temperature [30]. On the other hand, growth on a more strongly coupling substrate results in two mirror domains aligned with the substrate lattice. A well-studied case is the epitaxial growth of SL MoS₂, the prototypical TMDC, on a Au(111) sin-

gle crystal surface [31]. The presence of two mirror domains is particularly evident in the initial stages of the growth when two types of triangular MoS₂ nano-islands are found, rotated by 180° with respect to each other [32]. The simultaneous presence of mirror domains is detectable neither in the position of diffraction spots nor in the band structure obtained from angle-resolved photoemission (ARPES), at least in a non-spin resolved experiment. However, for a non-equal distribution of the two mirror domain areas, a

finite average spin polarization or circular dichroism for excitations across the bands might still be detectable [33].

Herein, we report on the growth of singly-oriented SL MoS₂ on Au(111) and measure the complete spin polarization of the states located near *K* and $-K$ by means of spin-resolved ARPES.

2. Results and discussion

The growth procedure used to synthesize SL MoS₂ differs from that reported earlier since here the synthesis takes place at high temperature. Earlier reports [31, 32, 34] obtained MoS₂ SL on Au(111) with a two-step process: first, Mo was evaporated in H₂S atmosphere with the substrate at room temperature and then the sample was annealed in H₂S at high temperature. In our experiment, Mo was instead evaporated in a background pressure of 2×10^{-6} mbar of H₂S onto a clean Au(111) surface while the substrate was kept at a temperature of 823 K. The Mo deposition rate was kept low ($0.005 \text{ ML min}^{-1}$). These optimal conditions were determined by following the behaviour of the Mo 3d and S 2p core levels during the growth measured in real time with fast x-ray photoemission spectroscopy (XPS) as shown in figure S1. The high temperature is pivotal to ensure the mobility of the Mo atoms on the Au surface, which promotes the growth of larger MoS₂ islands. However, one should consider that also Mo–Au alloying or S depletion could occur if the temperature of the substrate is too high. The acquisition of fast-XPS spectra during the growth was thus crucial to monitor the appearance only of the core-level components associated with MoS₂ and therefore fine tune the growth parameters. Furthermore, it was possible to avoid the growth of Mo clusters and partially sulfided species (giving rise to Mo 3d core level components appearing at lower binding energy compared to those of MoS₂ [35]), which did not convert into MoS₂ even after prolonged annealing in H₂S atmosphere without dosing Mo.

The overall structural properties of the resulting layer were characterized by scanning tunneling microscopy (STM) and low-energy electron diffraction (LEED). The STM images (figures 1(b)–(d)) show a small-scale hexagonal atomic structure due to the top sulfur layer of the S-Mo-S structure of MoS₂ together with a large scale moiré pattern due to the lattice mismatch between MoS₂ and Au(111). The moiré superlattice is well visible on the Au terraces and maintains its orientation across the Au atomic steps. The atomically-resolved images acquired in the green- and blue-framed regions (figures 1(c) and (d), respectively) show that the MoS₂ layer extends over the Au atomic steps with carpeting effect. The large area image (figure 1(b)) evidences the lack of any domain boundaries or dislocations on the entire probed area ($\sim 80 \times 80 \text{ nm}^2$).

Fast-Fourier transform (FFT) analysis (figure 1(f)) was carried out on the representative atomically-resolved image measured on a Au terrace (figure 1(e)). The alignment of the FFT spots as indicated by the blue line in the figure shows that the MoS₂ layer is aligned along the direction of the moiré superstructure, and thus aligned along the crystallographic axes of the Au(111) substrate. This implies that only two orientations of the MoS₂ layer are possible, rotated by 180° with respect to each other. This finding is at variance with respect to the results reported in literature for SL MoS₂ grown with the earlier synthesis method [31, 32, 34], for which a misalignment angle of 0.45° between the MoS₂ and the Au substrate was observed [34]. Moreover, by comparing this outcome with the LEED (figure 1(g)) and SPA-LEED patterns (see below) we can deduce that the moiré superstructure is due to the 10×10 surface unit cell of MoS₂ over the 11×11 unit cell of Au(111).

To measure the average domain size of the MoS₂ layer we performed high-resolution *k*-space measurements by means of spot-profile analysis (SPA)-LEED. The two-dimensional pattern measured at a kinetic energy of 120 eV is shown in figure 2(a). Besides the zeroth- and first-order spots, the image shows the appearance of additional diffraction spots due to the moiré superlattice. This can be appreciated in figure 2(b), where the *k*-space has been probed around the (0,0) diffraction beam. The spot profile along the [10] direction is presented in figure 2(c). The panel in figure 2(d) shows a detail of the (1,0) MoS₂ spot together with the best fit analysis obtained using a Voigt function. The Gaussian width is 0.0210 \AA^{-1} , while the Lorentzian full width at half maximum (L_w) is 0.007 \AA^{-1} , corresponding to an average domain size of $1040 \pm 50 \text{ \AA}$. Since this is comparable with the transfer width of the instrument, the average domain size could largely exceed this value.

The high structural quality is also reflected in core level spectra obtained by XPS and shown in figures 3(a) and (b) for Mo 3d and S 2p, respectively. The Mo 3d spectrum can be fitted with a doublet (red) with the Mo 3d_{5/2} centered at 229.19 eV (spin-orbit splitting of 3.15 eV). The broad peak at 226.43 eV (yellow) is the S 2s core level. The S 2p core level shows two strong spin-orbit doublets [34]. The more intense peak at 162.15 eV (light orange) is assigned to the S 2p_{3/2} core level of the upper sulfur layer (i.e. the layer towards vacuum) and the weaker at 162.44 eV (dark orange) to the layer towards the Au surface. The width of these components (table S1) and the absence of additional peaks related to sulfided species not converted into MoS₂ or to the atoms at the edges of the MoS₂ islands [36] are indicative of the high quality of the layer, in accordance with the STM and LEED results.

The single orientation of the MoS₂ layer can be ascertained in an x-ray photoelectron diffraction (XPD) experiment, as already successfully demon-

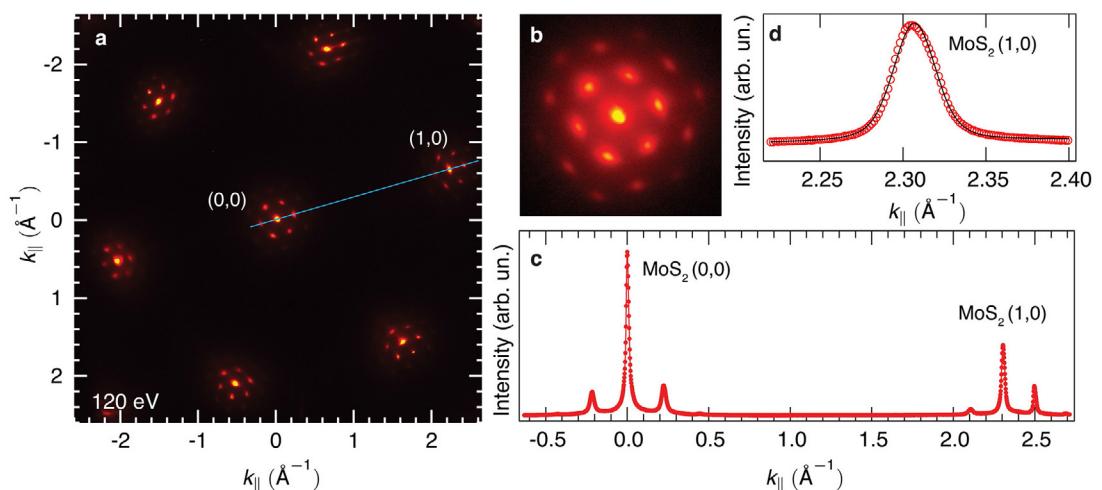


Figure 2. SPA-LEED measurements on SL MoS₂. (a) Two-dimensional SPA-LEED pattern measured at a kinetic energy of 120 eV. (b) Detail acquired around the (0, 0) diffraction spot. (c) Spot profile along the [10] direction (blue line shown in figure (a)). (d) Detail of the (1, 0) diffraction spot with profile analysis.

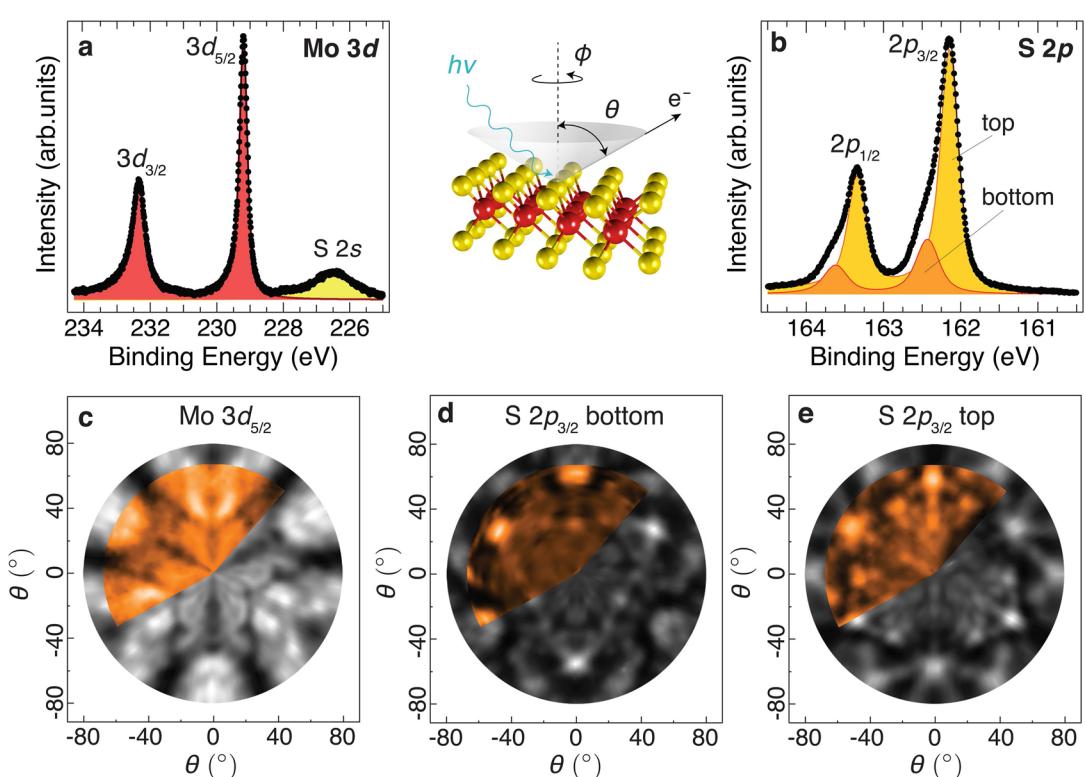
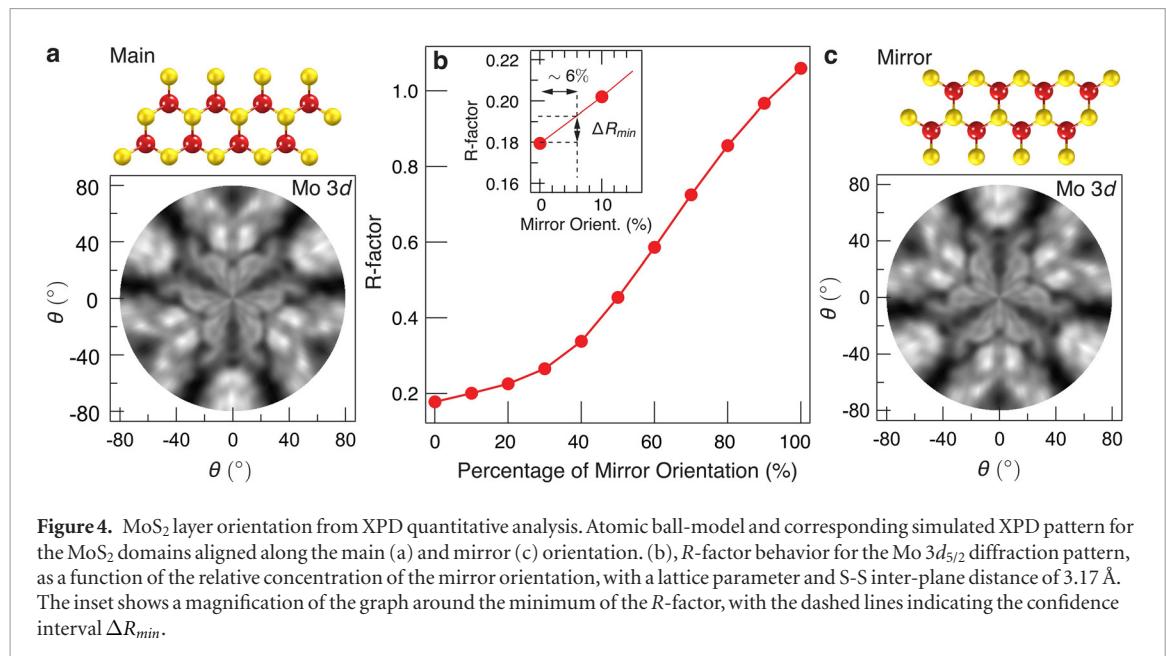


Figure 3. XPS and XPD characterization of SL MoS₂. (a) Mo 3d core level spectrum taken at $h\nu = 360$ eV (data points) with the resulting fit (line) and the fitted components (solid areas). (b) S 2p core level spectrum taken at $h\nu = 260$ eV (data points) with the resulting fit (line) and the fitted components (solid areas). The light and dark orange peaks correspond to the upper and lower sulfur layers, respectively. (c)–(e) Stereographic projections of the modulation function χ for Mo \$3d_{5/2}\$, S \$2p_{3/2}\$ bottom and S \$2p_{3/2}\$ top taken at a photon energy of 360 eV, 560 eV and 270 eV, corresponding to a kinetic energy of 130 eV, 397 eV and 108 eV, respectively. The colored sectors are the experimental data and the greyscale disks are the XPD patterns simulated for a layer with a single orientation and with the structure determined as specified in the following. The sketch in the centre depicts the XPD experimental geometry, where θ and ϕ are the polar and azimuthal angle, respectively.

strated for h-BN [37]. This technique is based on emission-angle-dependent modulations of the core level photoemission intensity from the different atoms in the layer [38]. The intensity modulations arise from the length difference between individual scattering pathways from the emitting atom to the detector and the coherent interference of the scattered waves. The

XPD modulations are thus directly reflecting the local structural environment of the emitting atom.

Figures 3(c)–(e) show stereographic projections of the modulation function χ (see Methods) for Mo and the lower and upper S atoms, respectively. The colored part is the data and the greyscale part is a simulation for a layer with a single orientation (see below). The



kinetic energy of the photoelectrons for the main XPS line in these experiments can be chosen by tuning the incoming photon energy; it was set high for the lower S atoms, favoring forward scattering processes from the Mo and S above, and low for the Mo and upper S atoms, favoring backscattering processes. All three diffraction patterns show a clear three-fold symmetry. Assuming a negligible influence of the underlying Au surface on the symmetry of the pattern, this already excludes the presence of equal areas of mirror domains, since these would give rise to a six-fold pattern.

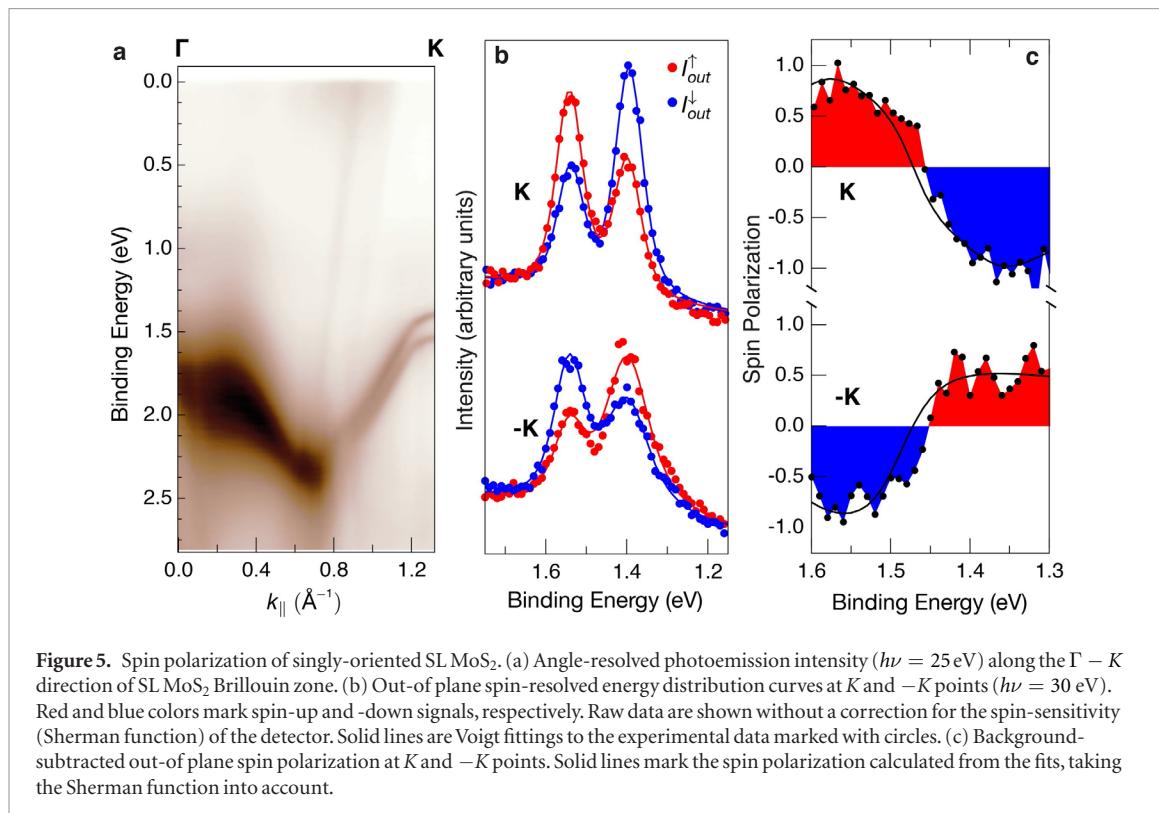
For further analysis, simulated diffraction patterns were calculated for the Mo 3d_{5/2} of trial MoS₂ structures using the software package for Electron Diffraction in Atomic Clusters (EDAC) [39]. In these simulations the underlying Au surface was totally neglected, which is appropriate because of the lack of a specific local adsorption configuration of MoS₂ on the substrate due to the lattice mismatch with Au(1 1 1). The orientation and the structure of SL MoS₂ were determined by minimizing the reliability (R) factor as a function of the abundance of the two possible orientations observed with STM, of the lattice parameter and of the S–S inter-plane separation. (see supplementary information for details (stacks.iop.org/TDM/5/035012/mmedia)). The total intensity I_{tot} in the simulated diffraction patterns for different admixtures of the two mirror orientations can be expressed as

$$I_{tot} = aI_0 + bI_{mir} \quad (b = 1 - a) \quad (1)$$

being I_0 the contribution to the XPD pattern sourcing from the main orientation (figure 4(a)) and I_{mir} the contribution from the mirror orientation (figure 4(c)). As displayed in figure 4(b), the R-factor shows a minimum when only the main orientation is present, while the agreement is worsened for any admixture of the two orientations. We estimated the maximum amount of the mirror orientation consistent with our data by calculating the confidence interval

displayed in the inset of figure 4(b) (see supplementary information for details). Based on this analysis, we can infer that the fraction of the mirror orientation in the MoS₂ layer does not exceed ~6%. These outcomes are consistent with the results stemming from the LEED pattern in figure 1(g) showing a clear three-fold symmetry, although these observations alone would not be sufficient to establish the domain orientation, which can be conclusively determined by XPD. The single orientation growth on Au(1 1 1) is likely due to the symmetry breaking originating from the substrate. While the first atomic layer of Au(1 1 1) has a six-fold symmetry and would permit both MoS₂ orientations, when the deeper Au layers are considered the crystal symmetry results to be three-fold. This apparently tips the balance between the two possible aligned orientations towards a single one (See supplementary information for the details on the stacking registry between MoS₂ and the Au(1 1 1) substrate).

Having established the presence of a single domain orientation by analyzing the geometric structure of the layer, we proceed by demonstrating its effect on the electronic structure. Figure 5(a) shows the dispersion of the MoS₂ bands measured by ARPES. The valence band maximum at K is clearly visible, including the spin-orbit splitting of the state. As pointed out above, the band structure observed in an ARPES measurement without spin resolution is not substantially affected by the presence of mirror domains. Many small domains would merely lead to a spectral broadening due to defect scattering. While the data in figure 5(a) are similar to previous findings for this system [31, 40], the linewidth of the states near K is substantially smaller (51 and 70 meV for the upper and lower band, respectively) than what was reported earlier [31], indicating a higher quality of the layer. The ARPES results show that the system has no detectable contributions from a second layer, as this would be observed as a second band near Γ (see



supplementary for a detailed discussion about this issue) [41, 42].

The single orientation character of the layer can be expected to result in a complete spin polarization of the bands near K and $-K$ and figures 5(b) and (c) show an experimental demonstration of this using spin-resolved photoemission spectroscopy. We find an out-of plane spin polarization of $86 \pm 14\%$, which is opposite for K and $-K$. In case of multiple-domain crystals, we expect the measured signal to be a mixture of contributions from oppositely spin-polarized K and $-K$ valleys, leading to a decreased value of spin polarization. Here, on the contrary, we measure a high magnitude of spin polarization, which further confirms the single domain orientation of the MoS₂ monolayer.

Previous spin-resolved ARPES experiments on bulk TMDCs have revealed a surprising complexity in the observed spin texture. Naively, inversion symmetry should lead to no observable spin polarization from the 2H structural polymorph while the 3R polymorph could give rise to a spin-polarized signal. The latter was indeed found for 3R MoS₂ [43] while, surprisingly, also 2H WSe₂ revealed strongly spin-polarized bands, essentially due to a combination of local symmetry breaking and surface sensitivity in photoemission [44]. Later, it was shown that the observed spin polarization from 2H MoS₂ could even be switched by excitation with circularly polarized light of different handedness [45]. In simple terms, this effect is based on the coupling of the light to different valleys in the band structure and the locking of valley and layer degrees of freedom in the 2H structure. Such a switching effect, and more in general a strong

dependence on photon energy and light polarization of the spin-polarization, should not be observable in the case of a single-orientation single layer [45]. This is confirmed by results for SL MoSe₂ grown on bilayer graphene [46]: the spin polarization was not found to change significantly with photon energy, even if it is much smaller than the value reported here, due to the presence of mirror domains. However, due to the complexity of the spin-polarized photoemission process, changes of the observed polarization due to e.g. final state effects cannot be excluded. We thus emphasize that the observed spin polarization in our case is consistent with the presence of a single domain, but that our analysis of the domain distribution rests not on this but on the structural analysis based on XPD.

3. Conclusions

In summary, we have presented a synthesis method to produce high quality, singly-oriented SL MoS₂. We used a multi-method approach to determine the structural properties of the layer and we measured for the first time the complete spin polarization of the states near K and $-K$ of SL MoS₂ by spin-resolved ARPES.

The synthesis method outlined here may represent a breakthrough for the large scale production of high-quality MoS₂ monolayers with a low number of dislocation defects. The availability of the singly-oriented MoS₂ monolayers obtained with this protocol may boost the research on the spin-valley degree of freedom in two-dimensional materials and could be the key to realize mass-produced devices based on the valleytronics concept. This growth protocol could

potentially be applied also for the synthesis of high-quality singly-oriented WS₂ or MoSe₂ monolayers on Au(111). Guided by the developments in graphene synthesis, one could expect that this method can be employed on other substrates or that destruction-free transfer mechanisms for large areas can be devised.

4. Experimental methods

The growth of MoS₂ samples, the LEED, the high-resolution XPS and the XPD experiments were carried out at the SuperESCA beamline of the Elettra synchrotron radiation facility in Trieste (Italy) [47]. The UHV experimental chamber is equipped with a Phoibos hemispherical electron energy analyzer (150 mm mean radius), implemented with a home-made delay line detector. The experimental chamber is equipped also with a rear-view LEED system. The Au single crystal was fixed on a Ta plate and the sample temperature was measured by two thermocouples spot-welded very close to it. The sample holder was mounted on a 5 degrees of freedom (x, y, z, θ, ϕ) manipulator.

The Au(111) surface was prepared by repeated cycles of Ar⁺ sputtering followed by annealing up to 920 K for 10 min. The heating and cooling rate was 1 K s⁻¹. After the cleaning procedure, the sample cleanliness was checked with XPS, which did not detect any trace of contaminants within the detection limit of 0.1% of a monolayer (ML) where 1 ML corresponds to the surface atomic density of the Au(111) surface. The long-range order was verified by acquiring the LEED pattern on the freshly prepared sample, which showed the extra spots of the herringbone reconstruction.

MoS₂ monolayers were grown by dosing molybdenum from a home-built evaporator, consisting of a Mo filament annealed through direct current heating, while keeping the Au substrate at 823 K and dosing H₂S (nominal purity 99.8%) through a leak valve at background pressure of 2×10^{-6} mbar. The Mo deposition rate was estimated by means of a quartz microbalance and amounted to $\sim 5 \times 10^{-3}$ ML min⁻¹. Therefore, the total amount of Mo deposited in 8000 s was ~ 0.67 ML. From the attenuation of the surface component of the Au 4f core level due to the presence of the MoS₂ layer, we estimated a MoS₂ coverage of 0.65 ML (1 ML corresponds here to one layer of MoS₂ covering the whole surface). High-resolution S 2p and Mo 3d core level spectra were measured at room temperature on the as-grown MoS₂ monolayer, using photon energies of 260 eV and 360 eV, respectively. The overall energy resolution was better than 50 meV. The surface normal, the incident beam, and the electron emission direction were all in the horizontal plane, with the angle between the photon beam and the electron energy analyser fixed at 70°. The high resolution core level spectra were acquired at normal electron emission.

SPA-LEED measurements were carried out at the Surface Science Laboratory of Elettra Sincrotrone Tri-

este using a commercial Omicron SPA-LEED [48]. The transfer width was better than 1000 Å. The instrument was used to acquire two-dimensional diffraction patterns at fixed energy as well as to measure one-dimensional high-resolution profiles along specific reciprocal space directions. The line-profile of the diffraction spots were modeled with a Voigt function [49]. The Gaussian broadening accounts for the finite coherence length of the primary electron beam and for the corrugation of the substrate. The Lorentzian contribution is connected with the size of the MoS₂ domains on the surface through the formula:

$$\frac{L_w}{a^*} = \frac{a}{D} \quad (2)$$

where L_w is the full-width at half maximum of the Lorentzian component, a^* and a are the reciprocal and real lattice vectors of MoS₂, respectively, and D is the average width of the crystalline domains.

XPD patterns for Mo 3d and S 2p core levels were acquired with different photon energies ($h\nu$) in order to change the corresponding electron kinetic energy (KE) to enhance forward and backscattering conditions. Specifically, Mo 3d was acquired with $h\nu = 360$ eV corresponding to electron KE of ~ 130 eV. The S 2p XPD pattern from the top S layer was acquired at $h\nu = 270$ eV (electron KE of ~ 108 eV) to enhance backscattering conditions, while the pattern from the bottom S layer was measured with $h\nu = 560$ eV (electron KE of ~ 397 eV) to enhance forward scattering conditions. At each energy more than 1000 spectra were measured for different polar (θ) and azimuthal (ϕ) angles. For each of these spectra, the peak fit analysis was performed and the intensity $I(\theta, \phi)$ of each component resulting from the fit, i.e. the area under the photoemission line, was extracted. Each XPD pattern was measured over an azimuthal sector of 160°, from normal ($\theta = 0^\circ$) to grazing emission ($\theta = 70^\circ$), as shown in the sketch in the centre of figure 3 of the main text. The resulting XPD patterns are the stereographic projection of the modulation function χ , which was obtained from the peak intensity for each polar emission angle as

$$\chi = \frac{I(\theta, \phi) - I_0(\theta)}{I_0(\theta)} \quad (3)$$

where $I_0(\theta)$ is the average intensity for each azimuthal scan. The agreement between the simulations and the experimental results was quantified by computing the reliability factor (R),

$$R = \frac{\sum_i (\chi_{exp,i} - \chi_{sim,i})^2}{\sum_i (\chi_{exp,i}^2 + \chi_{sim,i}^2)} \quad (4)$$

where $\chi_{sim,i}$ and $\chi_{exp,i}$ are the calculated and the experimental modulation functions for each data point i . The conformation of the MoS₂ layer was determined by minimizing the R-factor upon variation of the structural parameters employed in the simulations [38]. Following the determination of

the minimum R – factor, a confidence interval for the result was estimated by using the approach inspired by the common practice in LEED [50]. The variance of the minimum R -factor R_{min} is calculated by

$$\Delta R_{min} = \sqrt{2/NR_{min}}, \quad (5)$$

where N is the number of well-resolved peaks in a LEED I/V curve. Here we take $N = 350$, which is the approximate number of peaks in the 50 azimuthal scans acquired at different polar emission angles. Consequently, having $R_{min} = 0.18$ we find $\Delta R_{min} = 0.0136$.

ARPES experiments were carried out at the SGM-3 beamline of the synchrotron radiation facility ASTRID2 in Aarhus [51]. The energy and angular resolution were better than 30 meV and 0.2°, respectively. The sample temperature was ~ 30 K. The sample was transferred to Aarhus in air. After inserting it into the ultrahigh vacuum system, it was annealed to 770 K to remove adsorbed impurities.

Spin-resolved ARPES measurements were taken at the APE beamline of Elettra Sincrotrone Trieste, Italy [52]. The experimental chamber is equipped with a VG-Scienta DA30 analyzer and two very low energy electron diffraction (VLEED) spin polarimeters. Measurements were taken with a photon energy of 30 eV and p-polarized light, with the light incidence direction kept fixed at 45° with respect to the electron energy analyzer normal detection direction. The energy and angular resolution were better than 50 meV and 0.75°, respectively. Samples were transferred into the chamber in air and subsequently annealed up to ~ 800 K. Measurements were taken at about 80 K. Spin polarization P_i was determined from spin-resolved energy dispersion curves (EDCs) $I_i^{\uparrow\downarrow}$:

$$P_i = \frac{I_i^{\uparrow} - I_i^{\downarrow}}{S(I_i^{\uparrow} + I_i^{\downarrow})}, \quad (6)$$

where $i = x, y, z$ denotes the spin quantization axis in the reference frame of the detection and $S = 0.3$ is the Sherman function of the detector. $I_i^{\uparrow\downarrow}$ were corrected by a relative efficiency calibration and fitted with Gaussian-broadened Lorentzian (Voigt) peaks. The background contribution consisting of spin-unpolarized tails of lower-lying Au states was taken into account for the calculated P_i spectra. Quantitative spin polarization magnitudes were determined from the area ratio of the fitted peaks. P_i were transformed into the sample's reference frame by applying an Euler's rotation matrix.

STM measurements were carried out at the CoS-MoS facility at Elettra Sincrotrone Trieste. The images were acquired at room temperature with a SPECS STM 150 Aarhus instrument equipped with a W tip. The samples were transferred through air from the growth chamber to the STM chamber, where they were subsequently annealed up to ca. 800 K for 30 min.

The size of the surface areas probed varied between the different experimental techniques. Light spots sizes for the synchrotron radiation experiments were typically in the order of more than 100 μm , the electron beam size in SPA-LEED was about 100 μm while for LEED was several hundreds micrometers. Most importantly, no spatial inhomogeneity or presence of a mirror domain was noted when probing different areas of the Au(111) crystal surface (several mm in diameter) with different techniques.

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