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Electronic properties of rhombohedrally stacked bilayer WSe₂ obtained by chemical vapor deposition

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Twisted layers of atomically thin two-dimensional materials support a broad range of quantum materials with engineered optical and transport properties. Transition metal dichalcogenides (TMDs) in the rhombohedral ($3R$, i.e., 0° twist) crystal phase have been the focus of significant research interest in optical applications due to their particular broken inversion symmetry. Here, we report experimental and theoretical study of WSe₂ homobilayers obtained in stable $3R$ configuration by chemical vapor synthesis. We investigate the electronic and structural properties of these $3R$ WSe₂ bilayers with $3R$ stacking using micro-Raman spectroscopy, angle-resolved photoemission nanospectroscopy measurements, and density functional theory calculations. Our results demonstrate that WSe₂ bilayers with $3R$ crystal phase (AB stacking) show a significant valence-band splitting at the K point estimated at 550 ± 20 meV. We derived experimentally effective hole masses of $0.48m_e$ and $0.73m_e$ at the K point for upper and lower bands, respectively. Our work opens up perspectives for the development of optoelectronic and spintronic devices based on $3R$ TMD homobilayers.

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I. INTRODUCTION

Two-dimensional (2D) materials, such as graphene [1], h -BN [2], phosphorene [3], or transition metal dichalcogenides (TMDs) [4], are an ideal platform for studying exciting physical properties not attainable in their bulk counterparts. They exhibit a versatile electronic structure controllable by thickness, surface chemical adsorption, and strain [5,6]. The TMDs family with the MX_2 formula is of particular interest (metal $M = W, Mo$, or any other transition metal, and chalcogen $X = Se, S$). They are semiconductors with unique properties in the 2D limit, i.e., when thinned down to single monolayer (ML): direct band gap, large exciton binding energy, well-defined valley degrees of freedom, and wide spin splitting of the valence band [7]. The iconic feature of the TMDs materials family is most certainly the transition from direct to indirect band gap upon increasing thickness, from one to two MLs [8].

Controlling the in-plane crystal arrangement of layered TMD materials is an effective technique to attain exotic physical states [9,10], as observed in the drastically different physical properties between the $1T$ and $2H$ phases of TMDs [11,12]. Out of plane, the various stacking possibilities between few layer polytypes are also particularly attractive for piezoelectric applications, valleytronics, photodetectors, or transistors [13–15]. In any bilayer TMD configuration, the weak out-of-plane van der Waals coupling between the

adjacent layers offers a controllable degree of freedom as the (in-plane) twist angle between the top and bottom ML, which is associated with a rich solid state physics [16,17]. Along the continuum, selected twist angles are highly favored energetically and depict the most stable polytypes: AA' and AB stacking sequences, also referred to as $2H$ and $3R$ bilayer, respectively [18]. For WSe₂, it has been theoretically predicted that AB bilayer WSe₂ exhibits different electronic properties compared to the AA' stacking [18]. More specifically, the rhombohedral-stacked $3R$ bilayer WSe₂ polytype (AB stacking) hosts a broken inversion symmetry and may offer perspectives in high efficiency second harmonic generation [15]. Most of the previous studies focus on the “natural” $2H$ stacking order (AA'), i.e., identical to bulk hexagonal WSe₂, due to the reliance on mechanical exfoliation, whereas the recent advances in synthesis and sequential stacking processes are now allowing the investigation of the more elusive $3R$ bilayers (AB). In particular, these rhombohedral-stacked TMD $3R$ bilayers have attracted significant interest because of the emergence of an out-of-plane electric polarization from non-ferroelectric monolayer constituents. The bilayer $3R$ TMD has the polar point group symmetry of C_{3v} with the polar axis along the z direction [001]. The $3R$ homobilayer can be constructed by AB or BA stacking sequences, each of which develops a spontaneous electric polarization in the opposite direction of the other [14], leading to ferroelectricity [19] and strong photovoltaic effect [20]. Recent optical studies have shown differences between $2H$ and $3R$ stacked WSe₂ bilayers in their photoluminescence (PL) emission, which suggests that the different vertical stacking orders could provide a

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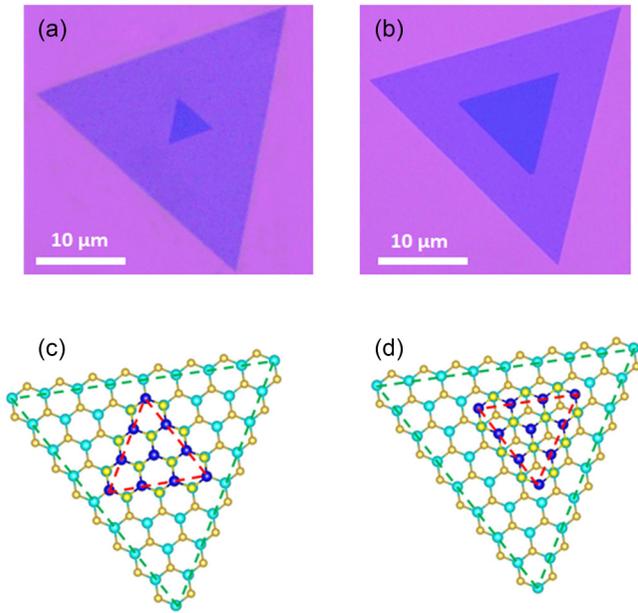


FIG. 1. Optical image of (a) hexagonal AA' ($2H$) and (b) rhombohedral AB ($3R$) WSe_2 homobilayers, as grown on the SiO_2/Si substrate. (c),(d) Corresponding atomic arrangements (Se yellow, bottom W cyan, top W blue). The dashed cyan and red lines mark the zigzag Se edges. The $3R$ configuration is characterized by matching edge orientation between subsequent layers.

way to controllably tune optical properties [21]. However, the impact of stacking order on the complete band structure is not fully addressed. In this context, the direct visualization of the electronic band structure and its correlation with the full crystal structure of bilayer $3R$ WSe_2 using angle-resolved photoemission spectroscopy (ARPES) is very appealing [22]. Nevertheless, in the absence of large-scale samples, it is not possible to separate contributions from areas with different stacking sequence or/and number of layers. To overcome this issue, we use here nano-ARPES to restrict the analysis to nanometric sized regions from well-controlled $3R$ -bilayer WSe_2 grown using chemical vapor deposition (CVD) [23]. The experimental band structure from nano-ARPES is then compared to theoretical calculations obtained in the density functional theory (DFT) framework. Additional investigations are carried out using micro-PL/Raman spectroscopy and synchrotron-based x-ray photoemission spectroscopy (XPS) for chemical and electronic-structure properties.

II. RESULT AND DISCUSSION

Bilayer TMDs present two highly energetically favored stacking sequences: $2H$ (AA') and $3R$ (AB) [18] but the difficulties in fabricating large area WSe_2 bilayer have hindered the analysis of the less stable $3R$ configuration. Here, we compare the properties of two monocrystalline bilayer WSe_2 with two different stacking orders ($2H$ and $3R$), which are synthesized during the same chemical CVD experiment (see details in the Methods section). The as-grown WSe_2 crystals are identified on the silica substrate as triangular shaped (50 microns large) whose thickness (1 or 2 MLs) is easily assessed by optical contrast. Optical images in Figs. 1(a) and 1(b)

show respective WSe_2 crystals with ML (light color) and bilayer regions (darker color) in $2H$ and $3R$ configurations, characterized by 60° and 0° rotational alignments of the upper WSe_2 layer with respect to the base WSe_2 MLs respectively [15,21,24]. The corresponding atomic configurations are illustrated in Figs. 1(c) and 1(d). In order to confirm the stacking order of the bilayer, we then performed micro-PL and micro-Raman measurements at room temperature.

In Fig. 2, we recorded reference micro-PL spectra of ML WSe_2 and compared it to the two different types of WSe_2 bilayer, $2H$ and $3R$. The type of stacking orders can be deduced directly from the mutual orientation of edges in the first and second WSe_2 layers. The $3R(AB)$ stacking features parallel edges in both layers, while $2H(AA')$ stacking has edges rotated by 60° [12,15,21]. An optical image of $3R$ - WSe_2 bilayer is shown Fig. 2(a). The corresponding integrated PL/Raman intensity maps are reported in Figs. 2(b) and 2(c), respectively (see also Supplemental Material S1 [25]). The variations in the intensity contrast in the PL intensity map confirms that the number of layers varies across the flake: a higher PL intensity scale actually indicates a decrease in the number of layers and vice versa. Individual local PL spectra acquired in each region of each flake are shown Figs. 2(d) and 2(e) and confirm this analysis. The 1-ML WSe_2 shows a much higher PL intensity than both types of WSe_2 bilayers, with about a factor of 10 enhancement compared to 2-ML WSe_2 [Fig. 2(d)]. The PL quenching is measured for both $2H$ and $3R$ configurations in comparison with the PL of 1 ML. The maximum intensity decreases by $83 \pm 3\%$ and $95 \pm 3\%$ for $2H$ and $3R$, respectively. Such large PL quenching at room temperature has been previously reported for many TMD bilayers made by stacking CVD-grown or exfoliated MLs [8,26] and is associated with the indirect band gap transition with increasing ML thickness [7]. Nevertheless, some differences between the WSe_2 $2H$ and $3R$ bilayers are revealed by PL measurements: the different peak wavelengths (1565 and 1572 meV) and peak full width at half maximum (FWHM, 62 and 44 meV, for the $3R$ and $2H$ stacking respectively). For both stacking the stark decrease in intensity indicates that the PL is dominated by momentum-indirect transitions at room temperature, but temperature-dependent measurements are required to study the details of the possible optical transitions [21,24,27].

The corresponding micro-Raman spectra collected at the same position are shown in Fig. 2(e). The observed Raman features are the degenerate in plane E'_{2g} and out-of-plane A_{1g} first-order Raman modes around 247 and 249 cm^{-1} , and the double resonance $2LA(M)$ mode at 258 cm^{-1} . Figure S1 of the Supplemental Material shows integrated Raman maps centered on the A_{1g} and E'_{2g} modes of the $3R$ WSe_2 homobilayer on the SiO_2/Si substrate. The signal is homogeneous over each mono- and bilayer regions. Note that the degeneracy of the E'_{2g} and A_{1g} modes, characteristic of WSe_2 1 ML, is fully lifted in the Raman spectra of WSe_2 $3R$ 2 ML as shown in Fig. 2(e). The peak at 308 cm^{-1} , absent in 1-ML WSe_2 , is the hallmark of thicker 2-ML regions, identical for both the $2H$ and $3R$ stackings. For both $2H$ and $3R$ bilayers, the E'_{2g} and A_{1g} modes now exhibit a ~ 2 - cm^{-1} frequency split, in agreement with theoretical prediction for $2H$ WSe_2 bilayers [28] and experimental reports on $2H$ and $3R$ WSe_2 homobilayers [29,30]. Due to the change in the crystal structure, the

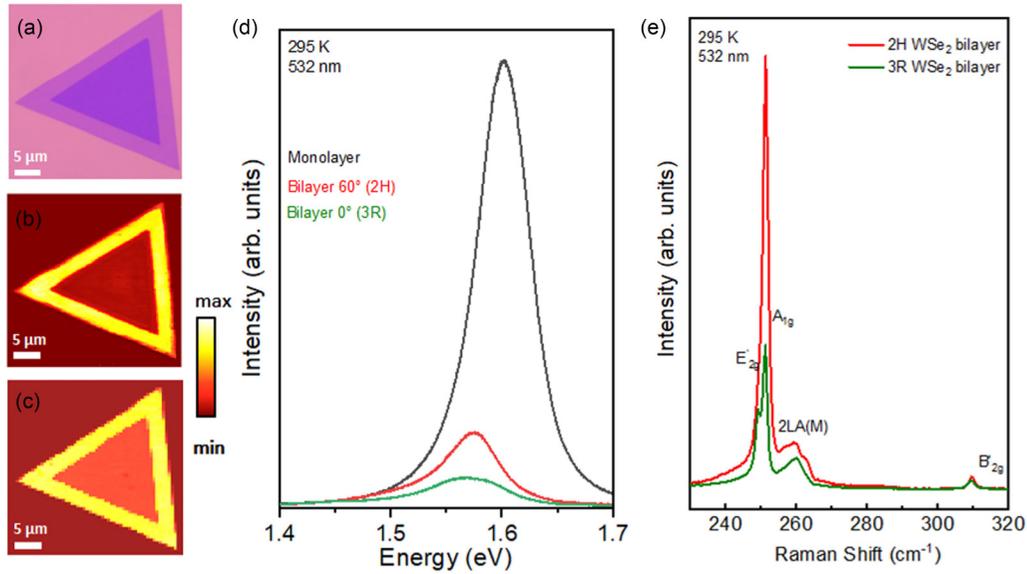


FIG. 2. Structural properties of 3R homobilayer $\text{WSe}_2/\text{SiO}_2$: (a) optical image, (b) room temperature micro-PL intensity map. (c) Integrated room temperature micro-Raman map centered on the E'_{2g} mode of the 3R WSe_2 bilayer. (d),(e) Comparison of individual micro-PL/Raman spectra acquired in the bilayer regions of 3R and 2H WSe_2 crystals (532-nm laser).

relative intensity between E'_{2g} and A_{1g} are different for room temperature Raman (532-nm laser) and the 3R exhibits a clear E'_{2g} signal, while it is near absent in the 2H configuration, as reported by others [29,31]. Our DFT calculations show that the frequency splitting is almost the same in both cases (see Fig. S2 of the Supplemental Material). These results confirm that the A_{1g} mode in the 2H stacking has a Raman signal which is higher by one order of magnitude with respect to the E'_{2g} one, but this intensity ratio is greatly reduced when moving to the 3R stacking due to symmetry-breaking arguments.

To further characterize the two types of configurations, we carried out optical reflectance measurements. It was previously shown by McCreary *et al.* [29] that the WSe_2 2H and 3R bilayer have different reflectance spectra. Here, we use wide-field interference reflection microscopy (IRM) to directly measure and compare the reflectance of 2H and 3R WSe_2 bilayers at multiple wavelengths. In particular, we report in Fig. S3(a) the contrast of WSe_2 bilayers vs WSe_2 monolayer for both stacking configurations, in the range $\lambda = 400\text{--}700$ nm. The contrast difference between 2H and 3R bilayers is most pronounced at $\lambda = 450$ nm (0.21 ± 0.01) and $\lambda = 500$ nm (0.17 ± 0.04), while it is minimal at wavelengths above 590 nm. Interestingly, while $\lambda = 450$ nm yields the highest contrast difference, the case of $\lambda = 500$ nm is of higher practical interest [see the line profiles and images in Figs. S3(b) and S3(c) respectively], as the reflectivity contrast changes sign between the two configurations. This provides an efficient way to discriminate stacking types in large assemblies of bilayer crystals using wide-field microscopy images acquired at a fixed wavelength.

In order to perform nano-ARPES experiments and to avoid charging effects of our TMD samples during the photoemission process; the CVD-grown WSe_2 crystals were transferred on a graphene/SiC substrate [30,32] [Fig. 3(a)]. Indeed, graphene is often used as a substrate for TMD heterostructures due to its conductive properties [33]. The transferred 2-ML

WSe_2 flakes were then identified by their optical contrast with respect to the graphene substrate [Fig. 3(b)]. The optical image shows large (lateral size about $50 \mu\text{m}$) triangular flakes of WSe_2 on the graphene layer and also confirms the

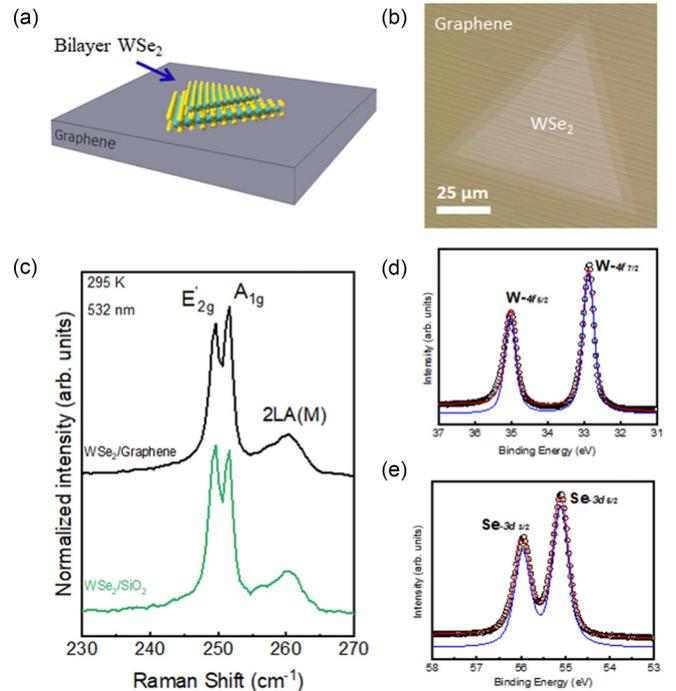


FIG. 3. Structural and electronic properties of bilayer WSe_2/Gr heterostructure: (a) 3D view of WSe_2 on top of graphene. (b) Optical image of the bilayer WSe_2 transferred onto the graphene layer; (c) room temperature micro-Raman spectra of bilayer before transfer (2-ML WS_2/SiO_2 ; green line), and after transfer (2-ML $\text{WS}_2/\text{graphene}$; black line); (d),(e) XPS spectra of W-4f and Se-3d core levels recorded in the bilayer region of sample in panel (b).

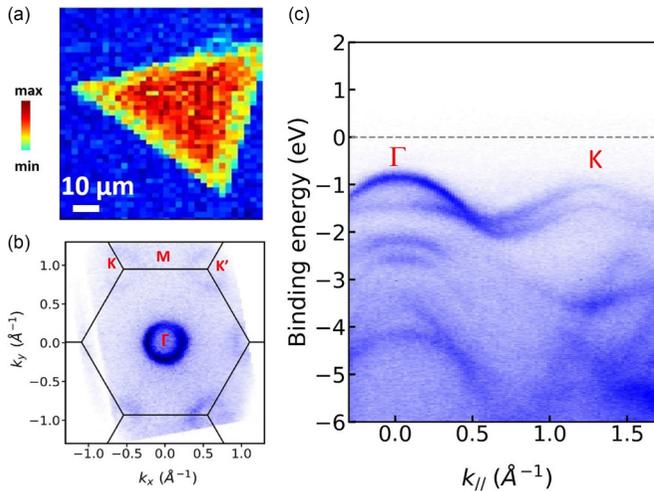


FIG. 4. (a) Spatially resolved ARPES map of the bilayer 3R-WSe₂ sample; (b) isoenergy cut in the ΓK plane obtained in surface-sensitive conditions ($h\nu = 100$ eV) at -2 eV binding energy, (c) high resolution map of the electronic band structure of bilayer WSe₂ collected at $h\nu = 100$ eV and $T = 70$ K along the $K\Gamma$ high-symmetry direction.

absence of large-scale damage which could be a consequence of the transfer process. Each stripe on the optical image is attributed to an atomic step on the SiC(0001) substrate [34]. In Fig. 3(c), we confirm that micro-Raman spectra of WSe₂/SiO₂ and WSe₂/graphene substrate are similar before and after the transfer to the graphene substrate (Fig. 2), which further confirms the absence of any structural damage during the transfer process. Chemical analysis by XPS is shown in Fig. 3(d), with W-4*f* and Se-2*p* XPS peaks deconvolution. The W component of WSe₂ is well fitted by a doublet peak at binding energy (BE) of 33.4 and 35.6 eV, corresponding to W-4*f*_{7/2} and W-4*f*_{5/2} core energy levels, respectively. The selenium Se-3*d* peak consists of a single doublet (Se-3*d*_{5/2} at BE = 55 eV and Se-3*d*_{3/2} at BE = 56 eV) corresponding to Se-W bonding, and validates the stoichiometry of WSe₂. The absence of any oxygen O-2*p* signal with binding energy around 5–10 eV marks the high quality of the interface between the WSe₂ and the graphene layer.

We now turn our attention to the electronic structure of the above described 3R-bilayer WSe₂. To do so, we have performed nano-ARPES measurements at the ANTARES beamline of the synchrotron SOLEIL at low temperature (70 K), with a spatial resolution of 600 nm. For this XPS measurement, the conductivity of the substrate is important to avoid the electronic charging of the film during the XPS/ARPES. Graphene/SiC(0001) may be a suitable material to combine with TMD due to its lack of dangling bonds and chemical inertness. In order to limit any charge transfer between the bilayer WSe₂ and graphene substrate, we have used the 3–4 layer of graphene/SiC, for which the Dirac point is located at the Fermi level [22]. In Fig. 4(a) we show the angle and energy integrated photoemission map coming from the 3R-WSe₂ bilayer crystal shown in Fig. 3(b). The photoemission intensity around the Γ point was acquired along the ΓK direction for a photon energy $h\nu = 100$ eV as a function

of coordinate on surface [35]. The mono- and bilayer regions of the flake can be directly identified as low and high intensity areas on the integrated intensity map, and match the data obtained by optical microscopy or Raman analysis. The large and homogeneous over surface intensity confirms the uniform thickness of the 2-ML region. Figure 4(b) presents an isoenergy cut obtained at BE = -2 eV binding energy of the 3R-WSe₂ BL (*AB* stacking). We have chosen this particular binding energy in order to highlight the valence-band maximum (VBM) dispersions both at the K and Γ points of the Brillouin zone (BZ). We observe a circular (isotropic) hole pocket at the Γ point and a triangular hole pocket at each K point which is typical for the hexagonal WSe₂ 2-ML crystals. The presence of a unique set of six pockets at well-determined K positions in the reciprocal lattice confirms that the WSe₂ crystal is of a single domain (i.e., monocrystal).

The energy-momentum cut along the high-symmetry ΓK direction of 3R-WSe₂ bilayer is shown Fig. 4(c). From the experimental ΓK spectrum, we can evaluate the number of branches at the Γ point, the energy position of the valence-band maximum (VBM), and the valence-band splitting at the K point of the band structure. We first note the presence of two separated bands near the Γ point confirming the band structure to originate from the 3R-bilayer region (2 MLs, *AB*). The energy difference between the Fermi level (E_F) and the VBM is approximately equal to 0.7 eV. Considering the optical band-gap energy (1.57 eV) of bilayer 3R WSe₂ determined in Fig. 2(b), we can conclude that our sample is hole doped, since the VBM is closer to E_F compared to the conduction-band minimum (CBM). This observation persists in the 1-ML region (not shown) and consequently suggests that the observed *p*-type character possibly relates to the growth conditions [36]. We have also fitted the experimental dispersions in the vicinity ($\pm 0.12^\circ \text{Å}^{-1}$) of the K point using a parabolic approximation in order to extract the effective mass values of the hole's bands. The experimental dispersion has been fitted with a parabolic model $E(k) = E + \hbar k^2/2m^*$ where m^* is the effective electron mass and \hbar is the reduced Planck. We find $m^* = 0.48m_e$ for the low energy band and $m^* = 0.73m_e$ for the high energy band. The observed splitting of the valence band along the ΓK direction of our 3R bilayer WSe₂ is due to the strong spin-orbit coupling originating from the high mass of the constituent elements and the lack of inversion symmetry [37,38].

To better understand the band structure as well as the specificities of the 3R-WSe₂ 2-ML configuration, we have performed out band structure calculations using DFT with the hybrid functional HSE06. We were also able to compare them with previous calculations at the G_0W_0 level, to consider many-body effects, already reported in Ref. [31]. It is known that the more symmetrical 2*H* configuration shows two energetically degenerate bands near the K points, while the 3*R* stacking presents four different bands ($k = 1.2\text{Å}^{-1}$) [21]. The band structure of 3*R* bilayer thus presents different transition energies at the K point. This is in contrast to 2*H* stacking, where both MLs, and their associate intralayer states, are energetically degenerate [21]. Comparing with the experimental data for a monolayer of WSe₂ or WS₂ [39,40], we see an overall broadening (30 meV) of the band structure at the K point. The upper and lower bands are theoretically

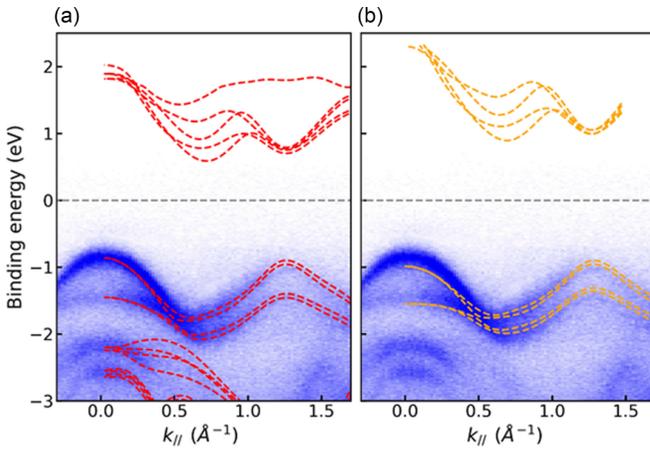


FIG. 5. Comparison between experimental and theoretical band structure of 3R homobilayer WSe₂. (a) HSE06 DFT (red) calculation for the present work. (b) G_0W_0 (yellow) taken from Ref. [31].

split by about 21 meV, which is larger than our experimental energy resolution and thus should have been resolved directly in our measurements. However, we do not clearly resolve four individual bands in the vicinity of K , only an increasing linewidth of the band structure in the direction of ΓK , which we attribute here to 3R-specific bands splitting. We conjecture that the sample roughness, induced in the growth or transfer process, does not allow for the full band structure determination at the nominal energy resolution. Figures 5(a) and 5(b) show the measured band structure and the calculated occupied states using HSE06. The distinctive features of band structure include the VBM at Γ (derived from the W d_{z^2} and Se p_z orbitals), the VBM at K (W $d_{x^2-y^2}/d_{xy}$ and Se p_x/p_y orbitals) and the valley between Γ and K (crossover from W d_{z^2} and Se p_z to W $d_{x^2-y^2}/d_{xy}$ orbitals) [32]. Supplemental Material Fig. S4 shows the EDCs profiles spectra at the K and Γ points with a Gaussian fit to estimate the splitting values. The valence-band splitting at the K point is about 550 ± 20 meV and the splitting at the Γ point is about 620 ± 20 meV. Those values agree with those calculated from the DFT calculations from the HSE06 and G_0W_0 correction, Figs. 5(a) and 5(b). The main difference between HSE06 and G_0W_0 calculations resides in the VBM: when HSE06 locates the maximum in Γ , by more than 40 meV with respect to the maximum in K , the G_0W_0 scheme provides the VBM in K by more than 100 meV. We recall that those energy differences are highly sensitive to lattice parameter and exchange-correlation functional choices as discussed previously for TMD monolayers [41], which can also be the case experimentally due to strain effect [42] in bilayer systems, for instance. The spin polarization of the depicted bands is shown in Fig. S5. We observe that the two highest VBMs possess the same spin state at K (spin down in red) while the next two lowest are of the opposite spin (spin up in black). This is directly linked to the 3R crystal structure and each pair of bands (spin up + down) corresponds to one of the two distinct WSe₂ layers. In the conduction band (CB), the energy ordering at K is different with spin up and down bands from the first layer, and then spin up and down from the second layer at slightly higher energy. The difference between CB and VB band ordering at K is due to a much

smaller spin splitting in CB, combined with a high degree of localization of those states, which corresponds to a weak hybridization between the layers. In the Q valley, the spin polarization is globally weaker, with a much stronger mixing of states between the two layers, but it remains sizable and leads to a distinct ordering of spin in the CBs.

In summary, we have successfully studied the growth and the electronic structure of a 3R stacked bilayer WSe₂. A minor PL quenching and peak shift were observed for the 3R WSe₂ bilayer structure with respect to the 2H stacking. By performing nano-ARPES experiments, we have spatially mapped the electronic band structure of 3R WSe₂ bilayer. We observed sharp ARPES bands, in particular for the VBM near the Γ and K points. We directly observed the strong valence-band splitting at the k point and its value is in good agreement with the DFT calculations. We have further extracted the effective masses at the K point of the BZ, finding hole effective mass of $0.48m_e$ (upper band) and $0.73m_e$ (lower band). The valence-band splitting of the VB at K is experimentally determined to 550 ± 20 meV. Our results provide a starting point for future studies of electronic properties of WSe₂ bilayers and their applications in spintronic and valleytronic devices.

III. METHODS

WSe₂ crystals were grown by chemical vapor deposition (CVD) in a customized 4 inch. Nabertherm (RST 120) horizontal oven on SiO₂/Si substrates [283-nm thick SiO₂ on Si(100), Nova Wafers]. Argon (Ar, 200 sccm) and hydrogen (H₂, 10 sccm) are used as carrier gases; the pressure is 2.5 mbar for the whole growth procedure. The actual growth duration at high temperature is 30 min, excluding the temperature ramps from and to room temperature (60 min ramp up, 120 min ramp down). During growth, the selenium source is elemental Se (Neyco 99.9%) contained in a quartz crucible heated at 250° C. The tungsten source is H₂WO₄ (Acros Organics, 99%) mixed with KCl (Alfa Aesar, 99.9%) powder heated to 850° C, placed upstream to SiO₂/Si substrates. Both 3R and 2H bilayers are found over the SiO₂/Si substrate (5 × 2 cm), with an overwhelming fraction of 3R homobilayers compared to 2H stacking, in line with other experimental reports using comparable substrate temperature [15]. Graphene on 4H-SiC(0001) is used as conductive host substrate during the ARPES investigations, produced via a two-step process in another CVD system [43]. Prior to graphitization, the substrate was hydrogen etched (100% H₂) at 1550° C to produce well-ordered atomic terraces of SiC [34]. Subsequently, the SiC sample was heated to 1000° C and then further heated to 1550° C in an Ar atmosphere to obtain the desired single layer graphene. For the transfer process PMMA was spin coated onto the WSe₂ flakes and peeled them off from the SiO₂ substrate by wet etching in KOH solution. Afterward, the PMMA/WSe₂ layer was transferred onto the conductive graphene/SiC substrate. The PMMA was finally removed using acetone. After the transfer of WSe₂ onto graphene, an annealing process at $T = 300^\circ \text{C}$ for 30 min in UHV (base pressure better than $P \sim 10^{-10}$ mbar) was used to further clean the surface and interface of the WSe₂ graphene heterostructure. The shape of the flakes and their huge size was not modified after the transfer [44].

Micro-Raman spectroscopy measurements were conducted at room temperature using a Horiba Labram Raman spectrometer operating at laser of wavelength $\lambda = 532$ nm. The ARPES experiments were performed at the ANTARES beamline of the SOLEIL synchrotron light source (Saint-Aubin, France). The ARPES data were taken at a photon energy of 95 eV, using linearly polarized light. All ARPES measurements were carried out at a pressure of 10^{-11} mbar and at a temperature of 70 K. The band positions were determined by energy curve analysis (EDC) of the ARPES spectra as reported in the Supplemental Material [25]. Spin projected band structure calculation of 3R WSe₂ homobilayer and IRM measurements of 3R and 2H bilayer WSe₂ are also shown, as well as are Raman active frequencies and phonon dispersions of 2H and 3R WSe₂ homobilayers.

Density functional theory (DFT) calculations were performed using the VASP package [45] using the projector augmented wave formalism [25,46,47] to tackle the core electrons issue. Fourteen electrons for W atoms and six for Se ones were explicitly included in the valence states. A grid of $12 \times 12 \times 1$ k points has been used in conjunction with a Gaussian smearing of 0.05 eV of width for partial occupancies. The bilayer WSe₂ in-plane lattice parameter was fixed to 3.32 Å with a vacuum height of 21.6 Å between periodic images in the z direction, to reduce spurious interaction. The first optimization process of the geometry was performed at the PBE-D3 level [25,48] to include van der Waals interac-

tion between layers and then the Heyd-Scuseria-Ernzerhof (HSE06) hybrid functional [25,49–51] was used as an approximation of the exchange-correlation electronic term, to obtain a reliable band structure after a Wannier interpolation procedure performed with the WANNIER90 program [52].

The data sets generated during and/or analyzed during the current study are available from the corresponding author on reasonable request.

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There are no conflicts to declare.

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