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Controlling phase separation in vanadium dioxide thin films via substrate engineering

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The strong electron-lattice interactions in correlated electron systems provide unique opportunities for altering the material properties with relative ease and flexibility. In this Rapid Communication, we use localized strain control via a focused-ion-beam patterning of TiO_2 substrates to demonstrate that one can selectively engineer the insulator-to-metal transition temperature, the fractional component of the insulating and metallic phases, and the degree of optical anisotropy down to the length scales of the intrinsic phase separation in VO_2 thin films without altering the quality of the films. The effects of localized strain control on the strongly correlated electron system are directly visualized by state-of-the-art IR near-field imaging and spectroscopy techniques and x-ray microdiffraction measurements.

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Understanding, designing, and manipulating phase transitions at microscopic length scales in a chemically homogeneous system is a central goal of strongly correlated electron material (SCEM) research. Several pathways exist for modifying the properties of SCEM and controlling their phase separation at relevant length scales. Such paths include local electric-field gating [1–5], ion-beam implantation [6], and light-induced phase transitions [7–9]. The electric-field and light-induced effects alter the local phases with inhomogeneous carrier doping or localized heating while ion implantation allows for spatially selective phase engineering by introducing local defects and substitutions of atoms. While these methods have generated considerable interest for controlling SCEM, they have their own limitations: The high gate voltages often require the fabrication of nanoelectrodes, making control of local fields across macroscopic length scales difficult. Ion implantation can induce local structural defects, which may add extrinsic factors to the phase transitions. Light-induced inhomogeneities are restricted to the diffraction-limited spot size and are often transient effects—the inhomogeneities disappear when the light is switched off. In contrast to previous methods which rely on directly altering the SCEM films, we demonstrate a different route for controlling phase inhomogeneities down to the nanoscale: interface facet engineering. As will be shown below, this method possesses several unique advantages for modifying the local properties in SCEM, including reduced chemical defects,

no postfabrication processing, and direct control of the extent of the anisotropy and inhomogeneity.

This research is motivated by previous investigations of the nanoscale phase separation of vanadium dioxide (VO_2) thin films at elevated temperatures [5,10–12]. IR near-field imaging studies revealed that the phase transition temperature and the microscopic phase separation patterns of VO_2 films vary considerably when grown on different substrates [10,13]. These distinct substrate orientation- and film-thickness-dependent mesoscopic patterns in highly oriented transition-metal oxide thin films originate from the epitaxial strain due to a lattice mismatch between the film and substrate [10,14–16]. For example, in VO_2 films on a (110)_R TiO_2 substrate with an in-plane rutile c_R axis, striplike unidirectional phase separations are evident [11]. These metallic stripes can extend throughout the entire sample for a typical sample size of around 10 mm × 10 mm. The width and periodicity of the stripes are typically sub- μm , depending on the sample thickness [11]. In VO_2 films on sapphire substrates (e.g., $c\text{-Al}_2\text{O}_3$), however, a randomized metallic nucleation takes place rather than pattern formation [17,18], while in VO_2 films on (001)_R TiO_2 substrates, the mesoscopic self-organized phase inhomogeneities are totally absent [10]. The substrate-dependent phase inhomogeneities for the different films are mainly the result of distinct elastic coupling and epitaxial strain between the VO_2 and the substrate lattice. The complex, self-organized elastic interactions at the interface lead to variations in transition temperature and electronic phase separation in VO_2 films.

The epitaxial strain-initiated local texture formation provides an opportunity for nanoengineering of functional SCEM materials. By introducing a nanoscale roughness and exposing different facets of the substrates before depositing the films, one can effectively control the interfacial strain and thus the

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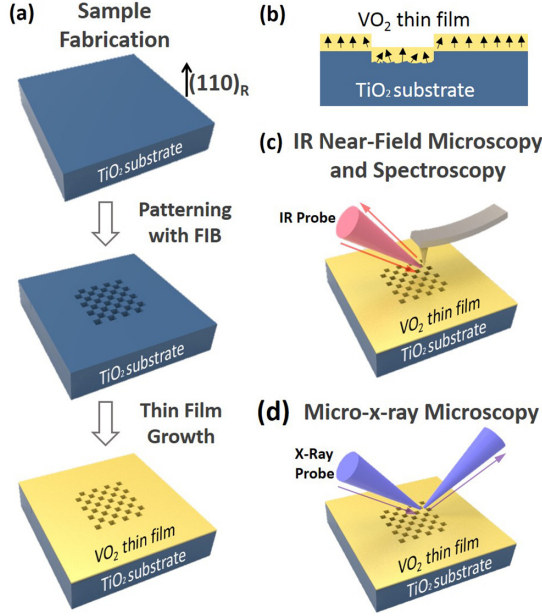


FIG. 1. Experimental design. (a) The samples are fabricated by patterning TiO_2 $(110)_R$ substrates with a focused-ion beam to produce a checkerboard pattern. After patterning, a VO_2 thin film (~ 71 nm) is grown over the substrate. The strain relief caused by the patterning of the substrate locally disrupts the VO_2 overgrowth, producing microscopic regions of different crystallinity. (b) Schematic cross section of the VO_2/TiO_2 interface showing the effects of etching and structural modification. The growth on the patterned regions produces a less coherent VO_2 film as different TiO_2 crystal facets are exposed by etching. (c) The local electronic phase transitions of the patterned samples are investigated with imaging and broadband spectroscopic s-SNOM with ~ 10 nm resolution. (d) The local crystallinity and structural phase transition of the VO_2 film are explored with micro-XRD with $\sim 1 \mu\text{m}$ resolution.

phase transition properties of the subsequent SCEM films. We demonstrate that the strain-induced electronic anisotropy in VO_2 can be locally transformed into isotropic conducting regions in a controlled manner. The engineered mixed electronic phases are studied with nanoscale resolution using near-field imaging and spectroscopy. The crystal structure of the VO_2 film is further characterized with micro-x-ray diffraction (μ -XRD) and reciprocal space mapping to provide a direct characterization of the local lattice parameters and structural phase composition.

We use a focused-ion beam (FIB) to create checkerboard patterns on $(110)_R$ TiO_2 substrates before VO_2 film deposition. Within the patterned region, 50% of the surface is masked, and the remaining 50% is exposed to the beam as squares. The typical applied voltage for the ion-beam lithography is ~ 30 keV, with a beam current ~ 24 pA (strata FIB 201, FEI Company, Ga ions). Patterns of different sizes (e.g., 200 nm, 500 nm, 1 μm , 5 μm) and different alignment angles [e.g., 0° , 30° , 45° with respect to the $(001)_R$ axis] are created on the same substrate. The entire fabrication procedure is schematically shown in Fig. 1(a) and the actual patterns can be seen in the Supplemental Material Figs. S1 and S2 [19]. The FIB etches the substrate surface to a depth of a few nm to a hundred nm. The high energy ions roughen the surface of the TiO_2 and modify the

structure by exposing different crystal facets in the TiO_2 along their collision path. The size of the structurally modified region depends on the dosage of the incident ion beam and the degree of beam spot dispersion [20]. At high doses, ion bombardment from FIB milling can relieve the implantation-induced substrate strain through localized lattice distortion [21].

A VO_2 thin film is then grown on the patterned substrate using reactive bias target ion-beam deposition at a substrate temperature of 400°C under 5.5 sccm Ar/O_2 80/20 mixture flow. The details on the sample growth and characterization can be found in Supplemental Material Fig. S3 [19] and Refs. [22,23]. The VO_2 film thickness is estimated to be 71 nm with a < 5 nm variation across the entire pattern. As a result of the FIB patterning, the VO_2 displays increased roughness in the checkerboard regions as compared to the single crystal thin film in the unpatterned areas (see the Supplemental Material Fig. S4 [19]). The disruption of the $(110)_R$ TiO_2 surface alters the orientation of the VO_2 in the patterned regions and is shown schematically as an interface cross section in Fig. 1(b).

In order to probe the electronic properties at the nanoscale, scattering scanning near-field optical microscopy (s-SNOM; NT-MDT NTEGRA-IR) is used to image the IR properties of the samples at single frequencies (10–11 μm) or collect spectra at specific locations with a broadband light source (ALS SINS beamline). The infrared near-field apparatus has ~ 10 nm spatial resolution ($\sim \lambda/1000$)—far beyond the light diffraction limit ($\sim \lambda/2$)—ensuring optical imaging and spectroscopic identification without integrating the information from mixed phases [17,24–36]. This fine resolution is especially helpful in studying phase transitions close to the critical temperature since the areal averaging in traditional far-field measurement techniques can hinder the data interpretation in systems with complex phases. Here, s-SNOM is able to distinguish different phases in the coexistence region and directly accesses fundamental properties of the materials, enabling direct mapping of the IR electronic structure at extreme subwavelength scales [34–36]. The unique sensitivity of s-SNOM facilitates the strain nanoengineering presented in this Rapid Communication.

In order to probe the local crystallographic structure variations in the VO_2 film, μ -XRD is performed using a monochromatic 11 keV x-ray beam with submicron resolution at the 7-ID-C beamline of the Advanced Photon Source [37]. The x-ray beam is focused down to a spot size of 1 μm [full width at half maximum (FWHM)] by a Fresnel zone plate in our experiments, which has sufficient spatial resolution to characterize the $5 \mu\text{m} \times 5 \mu\text{m}$ checkerboard patterns. Scanning x-ray imaging by diffraction intensity at the film peak position and local x-ray scattering specular scans are acquired at different temperatures from 295 to 350 K. The variations with strain of the spatial lattice constants for different phases are obtained by analyzing the XRD specular scans.

We first studied the VO_2 films on patterned $(110)_R$ TiO_2 substrates at room temperature. Three regions on the sample can be identified: the unpatterned single crystal VO_2 (UP), the substrate-unetched region of the pattern (UE), and the substrate-etched region of the pattern (EP). These three regions are clear in atomic force microscopy (AFM) topography, which is shown in Fig. 2(a). The line profile in Fig. 2(b) shows the height variation is within several nm, as the result of

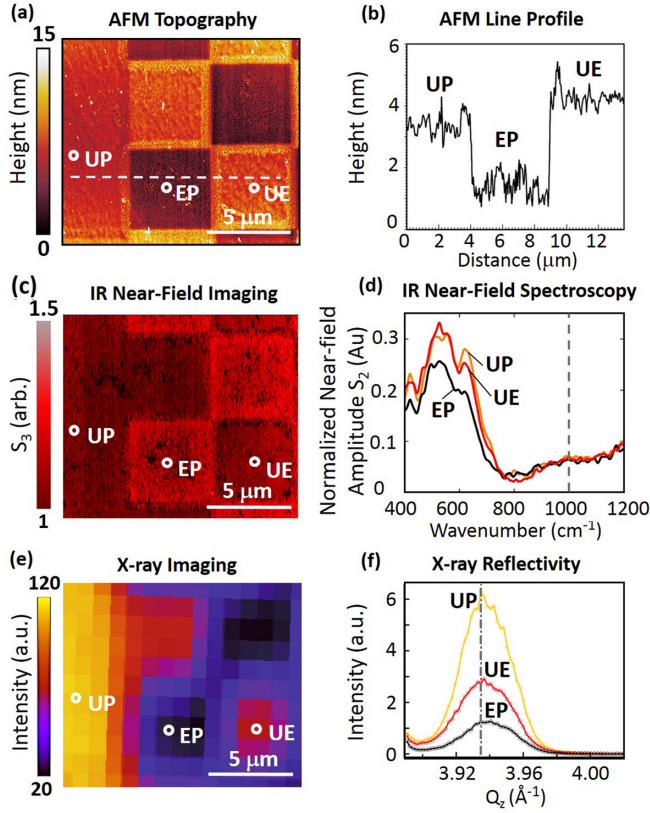


FIG. 2. Near-field nanospectroscopy and micro-x-ray diffraction at room temperature. (a) AFM topography at the edge of a checkerboard pattern with $5\ \mu\text{m} \times 5\ \mu\text{m}$ boxes. Three distinct locations on the pattern are identified: (UP) unpatterned region, (UE) unetched-substrate patterned region, and (EP) etched-substrate patterned region. (b) Topography profile along the white dashed line in (a). (c) Room-temperature near-field image collected at $\sim 10\ \mu\text{m}$ showing the IR response of the same pattern as in (a). (d) Broadband near-field spectra (referenced to gold) collected at each of the regions identified in (a) and (c), showing a shift in the phonon amplitude within the patterned region (curve EP). The unpatterned regions (as indicated by curves UP and UE) show nearly identical responses. The vertical dashed line indicates the $10\ \mu\text{m}$ wavelength where the near-field image (c) was collected. (e) Room-temperature μ -XRD intensity map where the diffraction intensity at the monoclinic $(022)_M$ Bragg peak position has been used to generate the contrast. (f) Specular scans measured at various locations as labeled in (e).

the substrate etching. Near-field microscopy is performed on the same area simultaneously with the AFM measurements. At room temperature, very little contrast is evident in the third harmonic near-field signal [Fig. 2(c), S_3 measured at $10\ \mu\text{m}$], indicating the entire sample is insulating. The exposed crystal facets in the substrate-etched patterned region (EP) lower the local transition temperature, resulting in a slightly higher optical conductivity compared to the unpatterned region (UP). The three regions identified in the topography are examined with broadband near-field Fourier-transform infrared microscopy (FTIR) in Fig. 2(d) using the SINS synchrotron beamline at the ALS, Berkeley [32]. The second harmonic (S_2) of the near-field response of a known phonon mode of VO₂ [38] is shown to decrease by $\sim 15\%$ in the substrate-etched

patterned region (EP), indicating a variation in the local crystal orientation due to the strain nanoengineering. To characterize the local crystallinity, a μ -XRD intensity map of the patterned region at the $(022)_M$ peak is shown in Fig. 2(e), highlighting the change of the Bragg peak as a result of the substrate patterning. The three representative locations of the pattern are indicated in Fig. 2(e) and the corresponding diffraction peaks are shown in Fig. 2(f). Moving from the unpatterned area (UP) to the substrate-unetched area within the pattern (UE), it is clear the diffraction intensity decreases. Moving into the substrate-etched patterned area (EP), the diffraction intensity is significantly reduced (Fig. S5 in the Supplemental Material [19]). The reduction of diffraction intensity is not attributed to polycrystallization as the VO₂ powder diffraction ring was not observed on the x-ray area detector (PILATUS 100K; see Fig. S6 in the Supplemental Material [19]), nor can the reduction in diffraction intensity be attributed to amorphization, as indicated by Raman measurements (Supplemental Material Fig. S7 [19]), which show a reduction in signal of well-known modes, consistent with previous reports [39–42]. Instead, the reduced diffraction intensity in area (EP) suggests that the out-of-plane crystal axis of the majority VO₂ film on the etched TiO₂ substrate is not aligned along $(110)_R$. Although these regions of the VO₂ film are not aligned, their orientations are not randomized either, due to the lack of a powder diffraction ring. The results shown in Fig. 2 indicate that the modified interface alters the local epitaxy of the VO₂ film, producing regions of single crystal VO₂ [unpatterned (UP)], less-aligned VO₂ [substrate-unetched region within the pattern (UE)], minimally aligned VO₂ [substrate-etched pattern (EP)], and the intermediate areas at the border between the different regions. In addition, we find two monoclinic $(022)_M$ Bragg peaks, M and M' , at room temperature which can be resolved with a double Gaussian fitting (see Supplemental Material Fig. S8 [19]). The M peak is close to the bulk value of the monoclinic VO₂ $(022)_M$ peak while the M' peak has a smaller out-of-plane lattice constant due to epitaxial strain. The peak width of the M phase gradually increases from region (UP) to (EP), suggesting the coherence length (as measured by the inverse of the Bragg peak width and defined as the length over which crystalline order is maintained) of the M phase along the out-of-plane direction gradually decreases from region (UP) to (EP). This reduction of the coherence length of the M phase with patterning is consistent with a relaxed (unstrained) VO₂ film, while the strained VO₂ film component M' shows no discernable coherence length differences in these three regions (see Supplemental Material Fig. S8 for details [19]).

The temperature-dependent s-SNOM characterization of the infrared properties of the VO₂/TiO₂ patterns, crucial for future applications of the proposed substrate nanoengineering technique, is presented in Fig. 3, where the insulating conductivity is in red and the metallic conductivity is in cyan. Figures 3(a)–3(f) demonstrate the response of a $5\ \mu\text{m} \times 5\ \mu\text{m}$ checkerboard and Figs. 3(g)–3(l) show a $1.5\ \mu\text{m} \times 1.5\ \mu\text{m}$ checkerboard. The near-field images at $11\ \mu\text{m}$ demonstrate that the VO₂ on the etched regions of the TiO₂ substrate [region (EP)] undergo an insulator-to-metal transition (IMT) at lower temperatures than the unpatterned, highly oriented region of the film [region (UP)]. Region (EP) also lacks the characteristic anisotropic metallic stripes (cyan color) evident

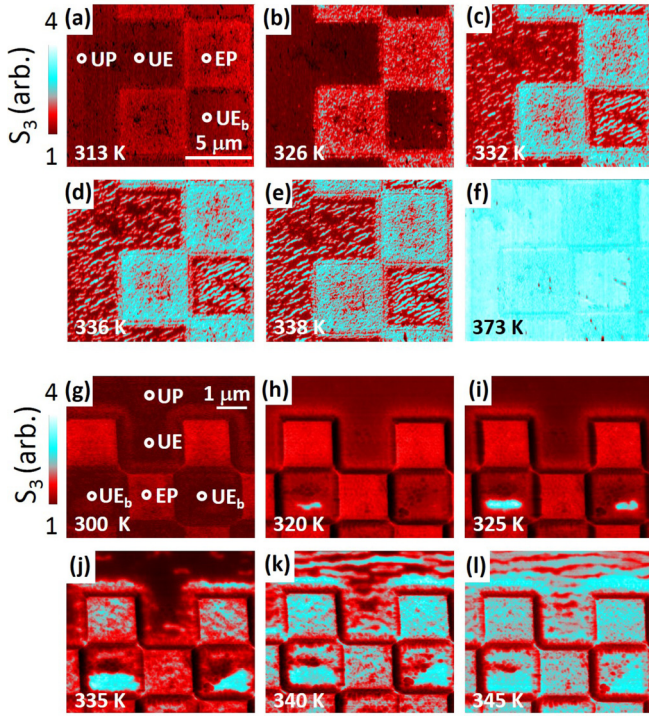


FIG. 3. Temperature-dependent infrared near-field images of patterned VO_2/TiO_2 at $11\ \mu\text{m}$, revealing area-dependent insulator-to-metal phase transitions. The metallic phase is shown in cyan and the insulating phase is in red. (a)–(f) $5\ \mu\text{m} \times 5\ \mu\text{m}$ checkerboard patterns at (a) 313, (b) 326, (c) 332, (d) 336, (e) 338, and (f) 373 K. (g)–(l) $1.5\ \mu\text{m} \times 1.5\ \mu\text{m}$ checkerboard patterns on the same sample, at (g) 300, (h) 320, (i) 325, (j) 335, (k) 345, and (l) 350 K. The smaller scale of the pattern shown in (g)–(l) exhibits strain-induced confinement effects, especially in the fully bounded UE regions (UE_b).

in the region (UP). This is due to the fact that the film on the unetched TiO_2 surface is subject to a highly oriented epitaxial growth, yielding a typical striplike insulator-to-metal phase transition as observed in strained single crystalline thin films [10,11,15], while the region of the film grown on the etched substrate region is less oriented and experiences strain relief. At the highest temperatures (above 360 K) the entire sample becomes metallic, indicating the quality of the VO_2 film and the robustness of the phase transition even with the etched TiO_2 substrate.

As the dimensions of the pattern are reduced, the effects of the adjacent strain distribution in a specific region of the pattern become increasingly important. This is evident in the fully bounded unetched substrate regions (UE_b) in Figs. 3(c)–3(e) and 3(h) and 3(i), where metallic stripes are observed to form at lower temperatures than the isotropic phase change of the etched-substrate patterned region (EP), indicating an intermediate state exists as a result of strain-induced confinement effects. The existence of this intermediate state is most pronounced when the length scales are comparable to the intrinsic limit of the self-organized stripes, in which case the adjacent strain environment influences the phase separation, as discussed in the Supplemental Material [19]. By measuring the checkerboard patterns with different lateral sizes and orientations fabricated on the same film, we have

reached the following conclusions: Regions (UP), (UE), and (UE_b) (as shown in Fig. 3) exhibit a striped phase coexistence, while region (EP) supports localized nucleation of the metallic state; region (UP) has a higher transition temperature than region (EP) and (UE_b); all regions complete the IMT above 350 K and the signal levels are close to each other at the fully metallic state; and the in-plane orientation of the checkerboard pattern does not play a major role in the mesoscopic phase separation.

The corresponding temperature-dependent structural variation across the same checkerboard pattern in Figs. 3(a)–3(f) is presented in Fig. 4. The x-ray reflectivity data, fit of the component phases, and the peak locations of the component phases as a function of temperature are plotted for region (UP) [Figs. 4(a)–4(c)] and region (EP) [Figs. 4(d)–4(f)]. The x-ray intensity map taken at the $\text{VO}_2(022)_M$ film peak position at 300 K is shown in the inset of Fig. 4(a). Figures 4(a) and 4(d) show the specular scans measured in the vicinity of the $\text{VO}_2(022)_M$ peak and $(220)_R$ peak as a function of the out-of-plane scattering vector $Q_z = 4\pi \sin(2\theta/2)/\lambda$, where 2θ is the scattering angle and λ is the x-ray wavelength. We find that in regions (UP) and (EP), two sets of monoclinic $(022)_M$ peaks (below ~ 340 K) and two sets of rutile $(220)_R$ peaks (above ~ 330 K) can be identified. As discussed earlier, at room temperature, two monoclinic $(022)_M$ peaks in the x-ray reflectivity, noted as M and M' , can be clearly resolved with the fitting. The M peak is close to the bulk value of the monoclinic $\text{VO}_2(022)_M$ peak while the M' peak has a smaller out-of-plane lattice constant. Above ~ 340 K, two rutile $(220)_R$ peaks, R and R' , can be identified. The R peak is close to the bulk value while R' has a smaller out-of-plane lattice constant. At 335 K, in the vicinity of the phase transition temperature, all four peaks (M , M' , R , R') are observed in Figs. 4(b) and 4(e) as a result of the strain-induced phase separation. Although the absolute intensity of the x-ray reflectivity at region (EP) is much lower compared to that observed at region (UP), the two-peak features in both locations are similar. The vertical lines in Figs. 4(c) and 4(f) are the calculated Q values— 3.9025 and $3.9297\ \text{\AA}^{-1}$ —which correspond to the bulk out-of-plane lattice constants $3.1978\ \text{\AA}$ $(110)_M$ and $3.2201\ \text{\AA}$ $(110)_R$, respectively [43,44].

The main structural differences between region (UP) and region (EP) are as follows. First, at 335 K, the R phase in region (UP) has an out-of-plane lattice constant slightly larger than the bulk value [Fig. 4(c)]. An in-plane lattice constant smaller than the bulk value along the $(001)_R$ axis is expected due to the Poisson effect [45], which explains the occurrence of the metallic stripes in region (UP) below the bulk T_c (~ 340 K) [46] as in-plane strain drives the phase separation. In region (EP), the rutile phase has a slightly smaller lattice constant than the bulk value [Fig. 4(f)] and the occurrence of the metallic nucleation in region (EP) is mostly due to the parts of the VO_2 film with out-of-plane orientations other than $(110)_R$. The second difference between region (UP) and (EP) is the fraction of the component phases at the same temperature. In region (EP), the R' peak strengthens at a lower temperature than in region (UP), while the R peak strengthens at a higher temperature than in region (UP).

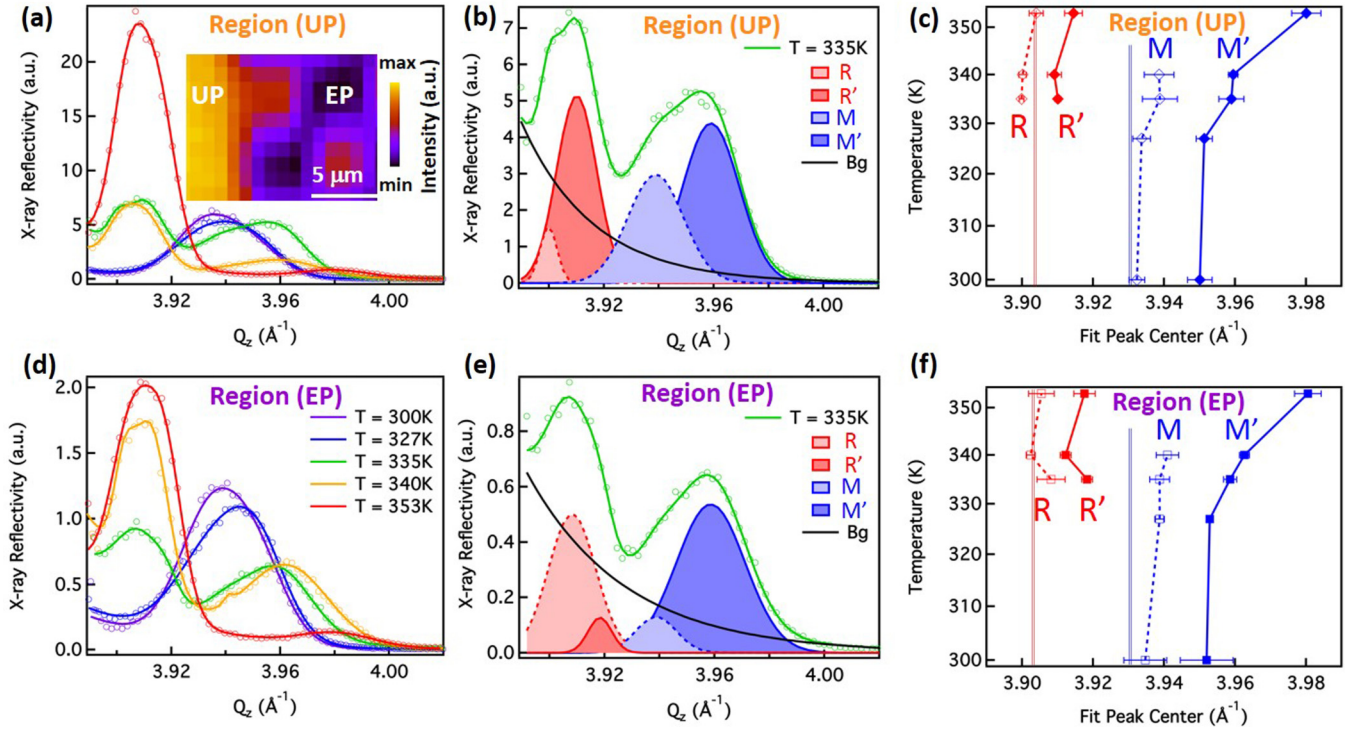


FIG. 4. μ -XRD analysis of the VO_2/TiO_2 patterned area. Micro-XRD intensity map taken at the film peak position at 300 K is shown in the inset of (a). The measurements in region (UP) are summarized in (a) the specular x-ray diffraction scans at various temperatures corresponding to the legend in (d), (b) the representative fitting of diffraction peaks at $T = 335$ K, and (c) the extracted temperature-dependent Bragg peak positions. (d)–(f) show the corresponding plots for region (EP).

The μ -XRD measurement is consistent with the electronic phase separation and the broad hysteresis of the VO_2 films observed in the near-field images. The alternating relaxed-tensile-relaxed strain environment modulates the VO_2 film crystal structure so that the M , M' , R , and R' phases coexist below and above 340 K. This is in agreement with the observation that the metallic and insulating states coexist below and above 340 K in the near-field [imaged as stripes in the highly oriented region (UP)]. The significantly lower XRD intensity and slightly different strain levels in region (EP) compared to region (UP) are in agreement with the expectation that the substrate-etched area has a multifaceted crystal domain structure, which leads to different metallic domain formation at the mesoscopic scale. This consistency between the near-field and μ -XRD measurements strongly suggests that the modification of the local structural phase coexistence through interface strain nanoengineering can lead to controlled reordering of the phase separation textures in the electronic response.

To explore the feasibility of substrate engineering with even smaller feature sizes, patterns with dimensions less than 5 and $1.5 \mu\text{m}$ have been fabricated (e.g., $1 \mu\text{m}$; see Fig. S9 [19]). For the $(110)_R$ substrate orientation, we found that the smallest length scale at which one can still reliably distinguish the etched versus unetched region is ~ 250 nm, irrespective of an etch depth between 5 and 110 nm. This can be understood given the fact that the periodicity of the spontaneous strain-induced striped phase separation of a ~ 50 -nm-thick unpatterned $\text{VO}_2/(110)_R\text{TiO}_2$ film is on the order of 200 nm (see the

Supplemental Material in Ref. [11]). In other words, the periodicity of the long-range ordered, self-organized stripes defines the ultimate “resolution” of the substrate’s impact on the local phase transitions in the films. In $\text{VO}_2/(001)_R\text{TiO}_2$ films, this restriction no longer applies since there is no in-plane anisotropy or conductive stripe state. As a result, the “resolution” of the pattern in $\text{VO}_2/(001)_R\text{TiO}_2$ films is only limited by the resolution of FIB writing, which can be reliably controlled down to 50 nm. We note that patterning a $(001)_R\text{TiO}_2$ substrate leads to the opposite trend in the transition temperature: The VO_2 transition temperature increases in the (EP) region as a result of the out-of-plane c -axis orientation.

In summary, we have demonstrated the feasibility of microscopical substrate engineering in vanadium dioxide thin films. By utilizing prepatterned substrates to control the local crystal growth and strain environment, we are able to manipulate the composite phases on a length scale close to the intrinsic self-organized phase separation in thin films without requiring postfabrication processing. Since the epitaxial strain-induced mesoscopic phase separation phenomenon is rather common in strongly correlated electron materials [14,47–50], this work provides a universal means for controlling the local electronic phase separation, especially in heterostructure systems.

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S.N.G.C. and J.J. contributed equally to this work. S.N.G.C., M.K.L., I.C.T., and H.W. wrote the manuscript with all of the authors contributing to editing. S.N.G.C., X.C., and J.Z. performed the near-field measurements. S.N.G.C. and M.K.L. analyzed the near-field data. H.A.B., M.C.M., and G.L.C. provided support and access to the nano-FTIR beamline at ALS. I.C.T., Z.C., Y.Z., and H.W. performed the XRD measurements at APS. I.C.T. and H.W. analyzed the XRD data. S.K., J.L., and S.A.W. deposited the films. S.K. analyzed the Raman data. J.J. and T.H.T. patterned and characterized the substrates. M.K.L., T.H.T., and H.W. designed the experiment.

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