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Spontaneous formation of nanostructures during pulsed laser deposition of epitaxial half-Heusler TiNiSn on MgO(001)

The half-Heusler alloy TiNiSn is a promising material for thermoelectric applications that is inexpensive and non-toxic. We demonstrate the epitaxial growth of smooth TiNiSn thin films on MgO(001) single crystal substrates by pulsed laser deposition, using transmission electron microscopy to investigate nanoscale structural and chemical inhomogeneities. In particular, an interfacial reaction results in the formation of a titanium-rich oxide nanolayer at the substrate interface, and this maintains epitaxy but perturbs film composition. Segregation effects are observed throughout the film, producing a narrow range of off-stoichiometric, half-Heusler compositions. The propensity of titanium to oxidise also leads to the commensurate formation of parasitic titanium oxide nanostructures which may be overlooked in bulk techniques but are a key feature that drives both the formation of Ni-enriched, full-Heusler phases and the local decomposition of the half-Heusler to form additional nickel-stannide phases.

Heusler alloys form a remarkably versatile materials system that supports functionalities suitable for a multitude of applications, including spintronics and magneto-optics. They have been the focus of substantial materials exploration research, and their optimised synthesis, as bulk materials or thin films, is of active concern. This versatility, and capacity for tunability, is of particular interest for the fabrication of thermoelectric generators, which are proposed as a means of scavenging waste heat that could tap into the 60% of energy generated that currently goes unused. In this regard, the semiconducting half-Heusler TiNiSn has emerged as a candidate n-type thermoelectric with a reasonable power factor and which—crucially—exploits abundant, non-toxic resources. Improvements in performance have been achieved by doping to independently modulate the electronic character and introduce structural complexity that can impede thermal transport. Such doping, however, also raises questions of phase segregation and stability, particularly when the material synthesis processes may be kinetically limited and conducted far from thermodynamic equilibrium. For example, intrinsic doping with excess Ni has in some studies driven nanoscale segregation into lattice-matched TiNiSn and TiNi₂Sn phases that scatter phonons. Our own recent studies of bulk extrinsic doping with Cu, on the other hand, revealed complex structural characteristics across several length-scales, from micron-scale grain-by-grain variations to nanoscale inclusions. There is, to date, no obvious link between the choice of the bulk synthesis method and the resulting nanostructures that form in doped or off-stoichiometric Heusler alloys. We therefore propose that thin film growth presents a useful
Films were deposited at substrate temperatures from room temperature up to 970 K with pressures ranging from a chamber base pressure of $10^{-8}$ Torr up to an argon partial pressure of $10^{-2}$ Torr. The deposition rate was typically 0.01-0.02 nm/s. In situ reflection high-energy electron diffraction (RHEED) patterns were recorded throughout deposition using a 30 kV electron source. Films were then characterised ex situ by AFM using a Digital Instruments Dimension 3100 atomic force microscope in tapping mode. A protective Pt layer (approx. 30-50 nm thick) was deposited on samples after AFM measurements to avoid subsequent surface oxidation effects. X-ray diffraction (XRD) measurements ($\theta - 2\theta$) were acquired using a PANalytical XPert3 diffractometer with a monochromated source of Cu Ka radiation.

Thin film samples were first prepared for transmission electron microscopy (TEM) using an FEI Nova Dualbeam focused ion beam (FIB) instrument to prepare cross-sectional lamella by routine methods. TEM and scanning transmission electron microscopy (STEM) data were then acquired using a JEOL ARM200cF probe–corrected instrument that was operated at 200 kV and is equipped with a cold field emission source. Spatially resolved chemical mapping was performed by electron energy loss spectroscopy (EELS) using a Gatan 965 Quantum ER spectrometer to acquire STEM dual–EELS spectrum images. Elemental quantification of EELS spectrum images was aided by a method of absolute quantification that employs a multi-linear least-squares fit to experimentally obtained reference spectra, as described elsewhere. The reference-based analysis was essential for accurate quantification since more standard routines were compromised by the proximity of EELS edges, which complicated background subtraction routines, and the known difficulties in using the Hartree-Slater scattering cross sections, whose uncertainty can be as large as 20% for quantification of M edges. Including the considerations detailed by Bobynko et al., contributions from atmospheric oxidation of the lamella surfaces were isolated and amount to <5 nm of encapsulating oxide on the surfaces of the film. Using these methods, compositions are determined with a confidence of 2% accuracy. Some spectra were further analyzed using non-negative matrix factorisation (NMF), which was implemented within the Hypespy L3 package and used to identify and separate common components in the spectra.

The deposition rate for TiNiSn thin films was not observed to vary significantly across the Ar pressure range of $10^{-5}$ mTorr to 10 mTorr, but the RMS roughness (measured across a $5 \times 5 \mu m$ area by AFM) approximately doubled, with films deposited at base pressure ($10^{-7}$ mTorr) appearing smoothest. This may be because low gas pressures enable more energetic species to arrive at the substrate, facilitating intermixing and diffusion. As the aim of this study was to promote epitaxial, smooth films, base pressure was used for subsequent depositions. Deposition temperatures between 300 K and 970 K were assessed, and AFM scans indicated a surface morphological change for films grown above 650 K, with films dewetting from the substrate (and presumably losing Sn) above 800 K. Morphological changes can be seen in panels (a), (b), and (c)
FIG. 1. AFM images of the surfaces of ∼50–100 nm-thick TiNiSn films grown at substrate temperatures of (a) 630 K, (b) 720 K, and (c) 800 K with (d) and (e) the corresponding RHEED patterns for (a) and (b) observed at the end of film growth, collected with the electron beam at a grazing angle to the (100) direction of the MgO substrate. Due to the roughness of the film in (c), no corresponding RHEED pattern was observed.

The images in (a) and (b) highlight a transition from rounded, roughly hemispherical-topped islands with a typical diameter of 30 nm to large, flat islands that span 100–200 nm. Note that the peak-to-peak range in both images is of order 10 nm, substantially less than the film thickness, so that it is clear that the islands have coalesced to completely cover the substrate. At higher temperatures, this is no longer the case, and in Fig. 1(c), dark regions can be seen in the image where the film fails to coalesce, resulting in a peak-to-peak range of 50 nm, which is comparable with the film thickness. Such three-dimensional island formation often arises from limited adatom mobility across step edges during the initial stages of growth.

The RHEED patterns in Figs. 1(d) and 1(e) are consistent with the AFM images shown in (a) and (b). In both patterns, the diffraction streaks have a spotty profile that is produced by a three-dimensionally rough surface where the RHEED electron beam is transmitted through the uppermost islands. Neither pattern is sharp, and the diffractive features are broadened in both images, but the diffuse background is strongest for lower deposition temperatures, because of the rougher, less ordered surface. In particular, films grown below 670 K produce a more diffuse specular reflection, indicative of poorer quality crystal growth. Both patterns, however, retain clear diffractive scattering that demonstrates an epitaxial relationship between the substrate and the growing film. In terms of film flatness, we find the optimal deposition temperature to be ∼720 K, in agreement with a previous molecular beam epitaxy study which is interesting, given the potential differences in adatom kinetic energies between these two techniques. In that study, although evidence was presented for interfacial roughening, it was not discussed in detail but does suggest that the roughening observed in the present work is not driven by highly energetic sputtering processes, which can occur when PLD is performed under low pressure conditions.

To better understand the thin-film crystal structure, Fig. 2 shows further in situ RHEED patterns recorded during deposition of TiNiSn at 720 K. Significant surface roughening and island growth occur during the first 2 monolayers (ML) of growth, as indicated by the appearance of distinct spots and an increase in diffuse scattering in the pattern [Fig. 2(b)]. (Note that 1 ML is defined here as a single atomic layer of lattice-matched TiNiSn.) At a coverage of 2 ML, new spots appear at one third of the MgO(01) lattice rod spacing, indicating a surface reconstruction, or formation of surface structures, with a periodicity three times that of the bare substrate. This pattern was observed to some degree for all epitaxial films grown above 630 K. Its spotty, transmission-like nature, formed after a single monolayer of deposition, suggests substantial surface re-organisation and roughening that is more than a monolayer thick and which is therefore likely to involve several atomic layers of the substrate. In fact, we interpret the data as indicating the formation of a titanium oxide interfacial layer, which will be discussed in the context of TEM evidence below.

FIG. 2. RHEED diffraction patterns recorded throughout the growth of TiNiSn thin film on MgO(001) at 720 K. The images correspond to (a) the bare MgO substrate and approximate coverages of (b) 2 monolayers (ML), (c) 20 ML and (d) 200 ML. The electron beam was directed at a grazing angle to the (100) direction of the substrate.
The superstructure spots are apparent only at the 2 ML stage and disappear thereafter. Island growth persists throughout much of the subsequent deposition, but the diffuse background diminishes so that by a coverage of 20 ML [Fig. 2(c)], the surface crystal quality seems much improved. Figure 2(d) then shows the RHEED pattern after 200 ML deposition: this has partially recovered the streaky quality expected from a smooth, two-dimensional surface, which suggests a tendency toward gradual island coalescence, consistent with the broad, flat islands in Fig. 1(b). Throughout the entire deposition, the film retains a clearly crystalline and epitaxial character.

XRD measurements presented in Fig. 3 show the $\theta - 2\theta$ diffraction pattern for a film deposited at 720 K. Only (00L) reflections are observed for the film, consistent with a single-crystal Heusler film with a (001) out-of-plane orientation that is coherent with the substrate. The remaining major reflection is accounted for by the MgO substrate, and minor contributions to the pattern close to the MgO(002) peak arise from the polycrystalline Pt capping layer deposited after thin film fabrication. Heusler reflections are found to be peaked between the expected positions for the bulk full and half-Heusler lattices, suggesting a non-stoichiometric mean composition. The position of the (004) reflection at 62.03° would suggest a mean composition of TiNi$_{1.3}$Sn, showing a similar level of excess Ni to that discerned by EELS, but with an uncertainty relating to the approximation of applying a Vegard’s law comparison to relaxed bulk values, in addition to the clear spatial inhomogeneities described below. Asymmetrical broadening of this peak, sloping in the direction of the bulk full-Heusler reflection, is attributed to the incoherent summation of diffraction from regions with different lattice constants due to variations in Ni concentration. Unlike the data discussed in the MBE study, which described observed XRD reflections using a pair of Gaussian profiles, at least three pseudo-Voigt functions are required for a reasonable fit to the asymmetric peak, which is shown in Fig. 3(b). Although not a unique fit, it suggests the presence of at least three Heusler phases and may be more indicative of a narrow range of Ni-rich half-Heusler compositions. Accordingly, further improvements to the fit could be made by adding further components and the choice of three peaks here is justified on the basis of subsequent EELS measurements. The approximate compositions inferred from the fit are TiNi$_{1.5}$Sn, TiNi$_{2}$Sn, and TiNi$_{2.15}$Sn, with relative breadths attributed to the quantity and geometric extent of each phase. Regions of the highest Ni content are expected to be the smallest and the sharper TiNi$_{1.15}$Sn peak corresponds to the majority phase. The Scherrer relation gives a mean domain...
size of 31 nm, 14 nm, and 4 nm, respectively for these phases, indicating that the largest grains are those closest to the target composition.

We now turn to the TEM characterisation of a cross section through a typical film, grown at 720 K to a thickness of ∼80 nm. A low magnification overview of the film is presented in Fig. 4(a) and shows a fully dense, continuous film that lacks the obvious granularity of, for example, azimuthally disordered columnar growth. The trapezoidal features at the top of the film are identified as the broad, flat islands observed by AFM. Subtle variations in intensity running vertically throughout the film are attributed to either compositional variations or slight differences in diffractive scattering, both of which could arise from the intergrowth of islands that grew independently from distinct nucleation sites in the first few monolayers. Triangular features circled in Fig. 4(a) are likely nickel-stannide phases that form upon decomposition of the half-Heusler structure and are discussed further below. Convergent beam electron diffraction (CBED) patterns collected from three distinct regions are presented in Figs. 4(b)–4(d), using a focused beam of 1 nm diameter to acquire local diffraction information. The diffraction pattern from the substrate [Fig. 4(b)] shows the expected square symmetry with in-plane reciprocal lattice vectors that are coincident with those of the TiNiSn film [Fig. 4(d)], indicating epitaxy. The out-of-plane diffraction spots collected from the film around 10 nm from the interface can be indexed to the expected, unstrained F-43m structure. The pattern collected at the interface, shown in Fig. 4(c), is dominated by a superposition of the patterns arising from the substrate and bulk film but with weak distinct spots, indicated by arrows, that are consistent with a slight out-of-plane expansion of the MgO (real space) lattice, together with a more subtle contraction of the in-plane lattice spacings. The diffraction peak spacings indicate an in-plane lattice contraction to 4.17 Å accompanied by an out-of-plane expansion to 4.26 Å, thereby preserving the unit cell volume of MgO. The reduction in the substrate’s in-plane lattice constant at the interface is somewhat surprising as strain is generally expected to be accommodated in the growing thin film; the contraction is also more substantial than required to accommodate the nominal lattice mismatch and in fact switches from a situation where TiNiSn is deposited onto the 0.32% larger lattice of MgO to one where it is bonded to a substrate lattice that is 0.64% smaller. Both observations are attributed to chemical variations revealed by EELS, as discussed below. As the lattice spacings and strain are likely to evolve to their equilibrium values during the initial stages of deposition, this may explain the proclivity for three-dimensional island growth observed here and elsewhere. A final important observation of Fig. 4(c) is that it shows no evidence of the surface reconstruction observed using RHEED during growth [Fig. 2(b)], which would again appear as additional spots at a third of the spacing of the in-plane lattice spots. The absence of a superstructure indicates that the modulation observed in RHEED reflected a transient surface reconstruction that was overcome upon subsequent deposition so that the overlayer retained a cubic structure. Lattice coherence is more evident in STEM imaging, and Fig. 5 shows a STEM High-Angle Annular Dark Field (STEM-HAADF) cross-sectional image of the MgO/TiNiSn interface that resolves atomic columns directly. Image contrast is dominated by differences in the atomic number, and the bright spots in the top half are Sn atoms, whilst the dim spots in the substrate are Mg. Figure 5(c) is a Fourier-filtered version of the same data that highlights the in-plane lattice spacings across the interface. Both images indicate a continuity of lattice lines with only four dislocations in the field of view.

STEM-EELS spectrum imaging was conducted in a similar region to that of Fig. 5(a) in order to provide spatially resolved chemical maps of the MgO/TiNiSn interface. The results are presented in Fig. 5(d) and indicate the relative proportions of the main elements, revealing the effects of substantial chemical reactions, including the presence of an interfacial nanolayer. Moving from the bottom to top through the

![FIG. 4.](image-url) (a) TEM image of a cross section of a TiNiSn film deposited on MgO(001) at 720 K with features identified as nickel stannides encircled in red. [(b)–(d)] Nanoprobe convergent beam electron diffraction (CBED) patterns collected along a MgO (010) direction and from positions marked approximately in the TEM image. Diffraction patterns correspond to (b) the MgO substrate, (d) the TiNiSn film, and (c) the interface between the film and the substrate. The scaling is the same for all images.
plot, the substrate has a clear 50:50 MgO composition but with a small quantity of Ti that extends up to 6 nm from the interface. The MgO signal drops to zero over a distance of 2 nm, forming a region that is substantially wider than would be expected for the extremely flat MgO substrates measured by AFM prior to deposition. It is therefore taken as evidence of substantial structural reorganisation and elemental diffusion in the initial stages of deposition. There is then a concentration of Ti at the interface, and this appears to be partially oxidised, having presumably drawn oxygen from the substrate. The oxygen signal itself extends further than that of the MgO and well into the half-Heusler film, only diminishing to zero outside of this data set. Ti enrichment at the interface produces complementary enhancement of the Ni and Sn concentrations immediately above. As can be seen in Fig. 5(d), the Sn concentration only drops to be equal to that of Ti about 5 nm beyond the interface. The enhancement in Ni, which here results in a local mean composition of TiNi\textsubscript{1.38}Sn, persists further, beyond the range of this line profile and up to 10 nm from the interface.

Formation of a Ti-rich interfacial layer implies that the nature of the substrate’s epitaxial template differs from that expected for clean MgO and might, therefore, lead to a loss of epitaxy and cause disordered, polycrystalline growth. However, the layer adopts a very similar lattice spacing to that of MgO and the Fourier filtered HAADF image in Fig. 5(c) does not indicate a disruption to the epitaxial relationship. It is also clear that the MgO subsurface region is chemically modified by Ti implantation up to 6 nm beneath the centre of the nanolayer. This is observed in both the EELS data and the STEM-HAADF image, which shows a clear band of intermediate intensity at the interface rather than an abrupt step between the relatively heavy TiNiSn and the relatively light MgO. It is possible that implantation arises from the impact of energetic species during PLD, but this would be expected to also apply to both Ni and Sn, which is not observed. We therefore attribute the Ti implantation to diffusion, in agreement with previous observations\textsuperscript{40} that showed facile diffusion of Ti into MgO with cation substitution even at room temperature. In Fig. 6, variations in the Ti L\textsubscript{2,3} EELS edge across the nanolayer indicate a continuous change in the near-edge structure and as a consequence a change in the titanium oxidation state, moving from half-Heusler to interface to substrate. The chemical shift and fine structure of the Ti L\textsubscript{2,3} edges in the half-Heusler are very similar to those of elemental Ti, in agreement with the covalent, uncharged nature of Ti within the half-Heusler alloy. Moving to the
interface layer, there is a small shift in the peak maxima to higher energy-losses as the proportion of oxide increases. In the interface and the substrate below, the position of the Ti L3 peak shifts to 457.4 eV, in excellent agreement with expectations for Ti2+ arising from a cubic TiO phase41 (and inconsistent with the formation of TiO2). The near-edge structure observed for all positions below the half-Heusler are similar although with poorer signal-noise ratios due to the diminishing titanium content.

The formation of TiO which has a similar cubic structure to that of MgO and a very similar lattice constant (a0 = 4.176 Å)42 would explain the epitaxial relationship between TiNiSn and MgO. Although thin films of TiO on MgO have not been studied in detail previously, the surface termination of bulk titanium oxides is known to reconstruct to form a number of superstructure phases with the potential to lattice-match to the MgO substrate here.43,44 Incorporation of Ti into the MgO substrate further increases the possibilities. A Mg-Ti-O spinel phase has been previously identified to have a coincidental lattice match with MgO (with a lattice parameter approximately twice that of MgO),45 and it is feasible that the transient reconstruction observed in RHEED is either a TiO reconstruction or a transient reconstructed Mg-Ti-O metastable phase. The substitution of Ti in the MgO lattice also explains the measured changes in lattice parameters at the interface if interpreted as an intermediate MgO:TiO composition driving the in-plane lattice constant toward the smaller value for TiO.

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Incorporation of Ti into oxide inclusions perturbs the surrounding alloy composition, and the false-coloured spectrum image of Fig. 7 shows complementary green (Ni-rich) stripes immediately beneath each red (TiO) region. If the flux of atomic species arriving at the substrate is stoichiometric and

STEM-HAADF, on the other hand, are nano-inclusions of titanium oxide. In Fig. 7(c), absolute quantification of EEL spectra shows an equal increase in the quantity of titanium and oxygen in the inclusion (at around 10 nm in the plot) compared to the half-Heusler phase, suggesting TiO. Assuming an isotropic form, we approximate the inclusions to be clusters of roughly spherical volumes that likely decorate the surface during deposition and will extend out of the plane of the TEM image by a similar amount to their observed diameter in the image plane. In the case of the inclusion shown in the line profile, absolute quantification results indicate approximately 1270 Ti atoms and a TiO volume of 23 nm3 with a diameter of 3.5 nm. Performing a similar analysis across the entire spectrum image indicates that approximately 2.5% of the film volume is occupied by TiO. This level of oxide is consistent with that observed in the bulk target material, which subsequent study (not shown) indicates to exist as predominantly TiO2 and with a different particle size distribution to that observed in the films here. We surmise that it is likely that the target is the source of oxygen during growth and that the oxides are not transferred as complete nanoparticles during deposition. The presence of TiO rather than TiO2 is interesting because it has a structure that is coherent with the surrounding crystal matrix and allows epitaxy to continue, just as described above for the interfacial TiO layer. Perhaps this relatively unusual oxide phase is stabilised by lattice strain. A HAADF image resolving the atomic positions, dominated by Sn, which is heaviest, is presented in Fig. 8(a) with a Fourier-filtered image in Fig. 8(c). The latter clearly demonstrates continuity of the half-Heusler lattice around the oxide inclusion, with a small number of dislocations to accommodate strain. As the nanoinclusions are small enough to be embedded within the TEM lamella, their unique EEL spectra were isolated from that of the matrix using non-negative matrix factorisation (NMF). The two extracted spectra are presented in Fig. 8 and show a clear shift between the Ti within the inclusion and that of the matrix, with the inclusion again being assigned as TiO with Ti2+.

Segregation of Ti into oxide inclusions perturbs the surrounding alloy composition, and the false-coloured spectrum image of Fig. 7 shows complementary green (Ni-rich) stripes immediately beneath each red (TiO) region. If the flux of atomic species arriving at the substrate is stoichiometric and

![STEM-HAADF cross-sectional image of a ~100 nm-thick film grown at 720 K. A protective Pt capping layer was deposited immediately after deposition. (b) RGB colour-map determined from a spectrum image collected within the area marked by the red rectangle in (a), showing the distribution of Ti (red), Ni (green), and Sn (blue). (c) Plot of a single line profile through the quantification results marked by the white arrow in (b).](apm/2018/7/figures/fig7.jpg)
uniform throughout deposition, then it is likely that the segregation into Ni-rich and TiO phases occurs immediately after layers are deposited, producing the observed asymmetric elemental distribution. We speculate that trace oxygen arriving at the growing film is adsorbed and draws the Ti out of the underlying half-Heusler to decorate the growing surface with TiO nanoparticles that are subsequently buried. Sn does not appear to be as depleted in the region of the oxides [see Fig. 7(c)] and so is less accentuated in the residual underlying material. Determining an accurate composition for the Ni-enriched region is difficult because the region will, as with the oxides, be embedded and surrounded within the TEM lamella by matrix material; yet NMF decomposition is found to be unreliable because these two components differ only subtly in their Ni content. However, if the majority matrix phase is TiNi$_{1.12}$Sn and the Ni-enriched region is restricted to that immediately below the oxide inclusions, then the same geometric approximation applied to the TiO inclusions yields an average composition for the Ni-rich nano inclusions of ~TiNi$_{1.8}$Sn, which could account for the minority TiNi$_{1.85}$Sn fraction seen in XRD.

In some regions, greater extraction of Ti from the surrounding matrix cannot be supported without a collapse of the half-Heusler crystal structure, leading to decomposition to form a nickel stannide binary alloy phase, which occurs most notably at the substrate interface. One such region, taken from the same film as above, is illustrated in Fig. 9. In this region, the spectrum image yields a majority phase (i.e., excluding defects) of TiNi$_{1.15}$Sn, similar to that of Fig. 7(a), with the differences perhaps accounting for the vertical stripes in contrast seen in Fig. 4(a). Such variations in the majority phase Ni content across the film are reminiscent of the grain-by-grain variations seen in an earlier study of Cu-doped TiNiSn.\textsuperscript{23} If the elemental flux during deposition is uniform, then such variations indicate the post-deposition lateral diffusion of Ni, which is intriguing. Turning to the defects, the spectrum image in Fig. 9(a) highlights Sn-rich (blue) regions immediately above the interfacial TiO (red) that relate to similar features encircled in Fig. 4(a). A line profile through the region—shown in Fig. 9(b)—clearly indicates a composition rich in Ni and Sn. (Note that the 0% deviation in Fig. 9(b) corresponds to the local majority phase of TiNi$_{1.15}$Sn.) The structure
of this Ni-Sn alloy can be discerned through analysis of the lattice-resolved high-resolution TEM image shown in Fig. 9(c), where the defect region, false-coloured in blue, has a different lattice to either the surrounding matrix or the substrate. The Fourier transform shown in Fig. 9(d) was used to distinguish distinct crystal regions—this has the appearance of a diffraction pattern and includes Fourier components from the substrate (coloured red), the matrix (coloured green), and the inclusion (coloured blue). Note the geometric alignment of all the spots, indicating an epitaxial relationship between all three components. The Ni-Sn alloy’s Fourier components can be indexed using the structure of Ni$_3$Sn$_2$ observed on the [110] zone axis. This nickel stannide phase is one of two possible phases predicted in recent ab initio thermodynamic calculations; no evidence has been found for the other predicted phase, Ni$_3$Sn$_4$. The inhomogeneity represented by this relatively large, ~10 nm-high inclusion is difficult to reconcile with the uniform flux expected during PLD, and it is unclear if its formation is driven by an irregularity in the substrate, a fluctuation in the arriving flux during deposition to produce a NiSn “seed” island or the result of inhomogeneous diffusion post-deposition. However, the coincidence of an apparent dip in the surface profile above the inclusion—a gap between the flat islands observed by AFM—does suggest that the half-Heusler material on either side of the inclusion nucleated separately and merged together as the film grew.

We have demonstrated that pulsed laser deposition of off-stoichiometric TiNiSn on MgO(001) between substrate temperatures of 620 K and 720 K results in smooth, epitaxial films, albeit with some limited three-dimensional character that diminishes with increasing film thickness. Our results suggest that an initial three dimensional growth mode arises as a consequence of surface reactions that substantially roughen the substrate immediately upon deposition and thereby limit adatom mobility. These reactions most notably involve the extraction of oxygen from the substrate and titanium from the film to form a titanium monoxide phase and a Ti-implanted MgO surface region, both of which retain an epitaxial relationship with the bulk substrate and which therefore do not impede subsequent epitaxy of the deposited half-Heusler film. The introduction of a buffer layer, such as epitaxial vanadium, could act as a barrier to prevent this interaction with the substrate and will be the subject of future work. Growth beyond the interfacial layer leads to the formation of a narrow range of half-Heusler compositions that manifest in broadened XRD peaks. From EELS quantification, however, the distributions of phases is weighted to form spatially distinct regions with mean compositions of TiNi$_{1.25}$Sn for the matrix, TiNi$_{1.38}$Sn for a thin Ni-enriched layer above the interface, and TiNi$_{1.5}$Sn for the Ni-rich nano-inclusions, occupying volume fractions of 78.4%, 8.1%, and 11%, respectively. In addition, the TiO oxide phase, not including the interfacial oxide that likely draws oxygen from the substrate itself, occupies a 2.5% volume fraction that we attribute to oxides present in the PLD target. These Heusler compositions differ slightly from the predictions made from the XRD data, where the predicted compositions are TiNi$_{1.5}$Sn, TiNi$_{1.5}$Sn, and TiNi$_{1.5}$Sn. However, as has been discussed elsewhere, the TiNiSn half-Heusler lattice does not expand linearly with excess nickel concentration and interpreting the XRD data with Vegard’s law is therefore only indicative.

Trace oxygen present during PLD growth, most likely from titanium oxides in the target material, results in the formation of cubic titanium monoxide nano-inclusions which, curiously, do not disrupt epitaxy but do appear to drive the formation of Ni-rich half-Heusler phases and Sn-rich binary phases. It can be assumed that these inclusions will be less problematic if more pure PLD targets can be produced. It is interesting to note that the presence of neither TiO nor Ni$_3$Sn$_4$ was observed by XRD and both could therefore be easily overlooked without the spatial resolution of TEM techniques. Functionally, the presence of titanium oxides is not anticipated to impair performance, as indicated by experimental measurements and theoretical calculations.

Spontaneous phase segregation of the Ni-rich half-Heusler target material was to be expected from the literature, with a tendency to form crystals with compositions close to the nominal half-Heusler or full-Heusler. However, the nature of the segregation observed here is interesting. Although modest segregation into vertical stripe domains has been observed, the variation in the Ni content (as described by the “majority” phases discussed above) is slight. On the other hand, a clear motif of the films is the proximity of titanium oxides with regions of more pronounced Ni enrichment. TiNi$_{1.38}$Sn was observed immediately above the interfacial oxide layer, and TiNi$_{1.5}$Sn tended to form beneath TiO inclusions. The latter is consistent with the smaller enthalpy of formation of the full-Heusler phase, resulting in greater stability compared to the half-Heusler. Stabilisation of the TiNi$_{1.38}$Sn phase through a chemical interaction with the substrate and, perhaps, the epitaxial relationship with the Ti-doped MgO interfacial layer, however, highlights the difference between the thermodynamically equilibrated state and materials deposited under kinetically limited conditions and deserves further investigation.

The nanostructuring and oxide formation observed here are of key importance in the understanding of phase segregation effects in thermoelectric half-Heusler alloys, where the micro-/nanostructure of the material can have a key role in limiting the thermal conductivity and improving thermoelectric performance. Here, we observe phase segregation that appears to be driven by a proximity to TiO phases and the subsequent diffusion of Ti out of the majority matrix. We envisage both these results and the methods used for their characterisation, being of relevance to a broad range of other Heusler alloy systems.

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