Tungsten oxides are a family of transition metal oxides with fascinating physical and chemical properties that make them suitable for application in many research fields, including photocatalysis and photoelectrochemistry, gas and pH sensing, or energy storage. W oxides can form in a wide range of crystalline structures and compositions, including substoichiometric compounds, conventionally denoted as \( \text{WO}_{3-x} \) (0 < x < 0.4). Such suboxides are characterized by crystallographic defects, namely, oxygen vacancies ("planar defects"), which provide unique properties.

The presence of oxygen vacancies in \( \text{WO}_{3-x} \) structures leads to the creation of new discrete energy levels below the conduction band that can act as doping states. Moreover, the simultaneous existence of \( W^{5+} \) and \( W^{6+} \) sites is beneficial for enhancing the electrical conductivity, for harvesting light in the near-infrared region through small polaron absorption and for improving the oxide ability to detect specific gas molecules (e.g., \( \text{NO}_2 \) sensing).

Several efforts were given to develop methods to grow \( \text{WO}_{3-x} \) nanostructures, which can provide even more advanced properties compared with bulk counterparts. In particular, 1D nanostructures (e.g., nanowires [NWs], nanotubes) are a most explored type of morphology because they can enable improved charge transport, electronic, and optical properties.

In this context, tungsten oxide NWs have been widely investigated and several synthetic strategies were proposed in the literature. Hydrothermal and solvothermal methods, thermal evaporation, and thermal reduction are most commonly adopted techniques. These procedures, however, can be quite expensive and laborious or may require the use of dangerous W precursors or solvents. In the latter case, undesired by-products may be generated by side reactions (between organic solvents and W-precursors) and the presence of a catalyst is often required. Consequently, such techniques are hardly scalable.

The direct thermal oxidation of W metal has also been proposed as an alternative. Thermal oxidation, which is based on the exposure of a W metal piece or a W-coated surface to a controlled oxygen-deficient atmosphere at high temperatures, can in fact enable the growth of oriented \( \text{WO}_{3-x} \) NWs. The absence of oxygen, e.g., the use of vacuum conditions or inert gases, is key to enable the NW growth, therefore to avoid the conversion of W metal into a "compact" W oxide layer.

The morphology of thermally grown NWs depends on process parameters such as the annealing conditions (atmosphere, temperature, and time) and nature of the substrate (morphology and chemical composition). The directional growth mechanism provides the resulting NWs high degree of crystallinity, i.e., a single crystalline nature, which can further improve properties such as electron conductivity and consequently the material performances in many applications.

Although various thermal oxidation approaches are reported in the literature, most of the them are, also in this case, based...
on relatively drastic conditions, i.e., require vacuum, high temperatures (>600 °C), the use of catalysts, or combinations of such measures.\[6,12,28–30,33\] This limits the employment of large scale or thermally sensitive materials as substrates for the NW growth. Therefore, the growth of WO$_{3-x}$ NWs has so far been reported only on W specimen, or from W films deposited on Si/SiO$_2$\[12,28,29,33\] and alumina\[6,30\] substrates.

Herein, we demonstrate the feasibility of the thermal-oxidative growth under mild conditions of WO$_{3-x}$ NW arrays from W films sputtered on fluorine-doped tin oxide (FTO) slides. We operate at temperatures as low as 500–550 °C, in argon and at atmospheric pressure. We explore the effects of parameters such as the annealing environment, temperature, treatment time, gas flow rate, and W film thickness on morphology, composition, and crystallinity of the formed NWs. In conclusion, we show how this approach can be transferred to a variety of different substrates, including metal surfaces, glass, or porous oxide layers.

We sputtered W metal in the form of conformal films (nominal thickness 120 nm) on FTO slides (Figure 1a,b). The W-coated FTO slides were then thermally treated in Ar grade 5 (99.999%) for 1 h at different temperatures (450–600 °C) to identify the temperature onset for the thermal-oxidative growth of WO$_{3-x}$ NWs. As shown in the SEM images in Figure 1c,d, a compact oxide layer was obtained at 450 °C, whereas the growth of short NWs is initiated at 500 °C. The homogeneous formation of well-defined WO$_{3-x}$ NWs was observed at 550 °C (Figure 1e). On the contrary, the structure of samples treated at higher temperatures (e.g., 600 °C; Figure 1f) suggests that the NWs may initially form and then undergo substantial agglomeration and loss of their 1D structure likely due to a phase transition from monoclinic WO$_{3-x}$ to orthorhombic WO$_3$, as also shown in previous work.\[9\]

Apart from ultrapure Ar gas (99.999%), thermal treatments at 550 °C for 15 h were also performed in 4.6 grade Ar (99.996%), low concentration O$_2$/Ar mixture (O$_2$ content 0.03%), and air (O$_2$ content ≈20%) to assess the effect of oxygen on the NW formation (see Figure 2a–d). The gas flow rate was in any case 10 NL h$^{-1}$. The formation of suboxide NWs was observed only in a pure Ar atmosphere, i.e., Ar 99.999% or 99.996% (Figure 2a,b). The dark blue color of these samples is in line with their substoichiometric nature (see optical pictures in the insets). On the contrary, compact oxide layers are formed even in the presence of concentrations of O$_2$ as low as 0.03% (Figure 2c,d). The color of such layers is yellow, which suggests that even traces of O$_2$ (0.03%) in the annealing environment can cause the conversion of the sputtered W film into compact W oxide layers.

A comparison of NWs grown for 1 and 15 h is shown in Figure 3a–d. Regardless of the annealing time, the conversion of the W film into suboxide NWs is only partial, as from the cross-sectional SEM images (Figure 3b,d) it is still possible to see the presence of a compact sublayer underneath the NWs—likely a W oxide sublayer, as discussed later. Under the same conditions (pure Ar, 15 h) but at 525 °C, densely packed NWs can also be formed (Figure S1a–c, Supporting Information).

The size distribution statistics reported in Figure 3e,f show a typical length of ≈460 ± 160 and 450 ± 130 nm for NWs grown at 550 °C for 1 and 15 h, respectively. Therefore, no significant differences in the NW length could be observed when increasing the treatment duration. On the contrary, the average diameter of the NWs grown for 15 h is almost 3 times larger than that of structures grown for 1 h, i.e., 36 ± 9 nm versus 11 ± 3 nm, respectively. This suggests that, after nucleation, the WO$_{3-x}$ nanostructures grow along their length ((010) direction, discussed later) and gradually thicken (lateral growth) along the (020) direction.

We also explored the effect of the Ar flow rate by carrying out a thermal treatment (under optimized conditions: Ar grade 5, 550 °C, 15 h) with a substantially reduced flow rate, i.e., 2 NL h$^{-1}$. As shown in Figure S2a, Supporting Information, the gas flow rate seems to have a negligible effect on the NW structure (compare with Figure 3c,d). Moreover, no substantial differences in the NW morphology was found when varying the nominal W film initial thickness from 120 to 180 nm (see Figure S2b,c, Supporting Information).

![Figure 1. Top-view SEM images of a) FTO substrate, b) as-sputtered W film on FTO (W film nominal thickness = 120 nm) and c–f) W films on FTO thermally treated at c) 450 °C, d) 500 °C, e) 550 °C, and f) 600 °C. All the thermal treatments were performed for 1 h in pure Ar (99.999%) using a flow rate of 10 NL h$^{-1}$.

---

*Phys. Status Solidi RRL 2020, 14, 2000235* © 2020 The Authors. Published by WILEY-VCH Verlag GmbH & Co. KGaA, Weinheim
Thus, based on these results, we assume that the NWs form via a solid-state process at the W surface where $\text{WO}_{3-x}$ nuclei, probably formed due to the presence of surface adsorbed oxygen species (discussed later with X-ray photoelectron spectroscopy [XPS] data), act as seeds for the directional growth of the NWs. Material transport occurs from the W metal substrate (“seed” sublayer) toward the NWs, and enables the bottom-up growth of the NWs from the $\text{WO}_{3-x}$ nuclei. The NW growth seems mainly influenced by the annealing temperature, environment, and time.

In view of a perspective use of the NWs/FTO composites as optoelectronic elements, an adequate transparency is required, which translates in a minimized thickness of the “seed” sublayer (compact layer underneath the NWs).

Thus, we explored the growth of $\text{WO}_{3-x}$ NWs from thinner, i.e., nominally 60 nm-thick, W films. The results, compiled in Figure 3g–j, show that the formed NWs feature a morphology comparable with that of structures grown from thicker W films (nominally 120 nm-thick W films; Figure 3c). These samples are transparent as one can see in Figure 3h, which enables their use under backside illumination conditions.

By comparing structures grown from W films of different thicknesses, we noted that when growing NWs from a nominally 120 nm-thick W film, the sublayer thickness is $\approx 250–300$ nm (Figure 3b,d). When the nominal initial thickness of the W film is 60 nm, the sublayer thickness is $\approx 100–150$ nm (Figure 3i,j).

We hypothesize that during the thermal oxidation, part of the sputtered W film is “consumed” for the suboxide NW growth, whereas the rest undergoes, meanwhile, conversion into a suboxide “compact” layer. This may also explain why the W seed film is not fully converted into NWs, that is, the mobility of W atoms in a W oxide lattice in negligible with respect to that of W atoms in a W metal lattice. In other words, once the sublayer is oxidized, W atom transport becomes limited. Arguably, the samples are transparent because the sublayer undergoes oxidation to W suboxide—differently, a W metal film of comparable nominal thickness (e.g., 120 nm) is not transparent and shows the typical metal shine appearance (see Figure S3, Supporting Information).

The increase in the sublayer thickness with respect to the nominal initial thickness of the W film is likely due to a volume expansion of the W film when it undergoes oxidation during the NW growth. The volume expansion is ascribed to oxygen incorporation during oxidation. Such volume expansion in terms of volume of the resulting oxide versus the volume of the parent metal film can be quantified by the Pilling–Bedworth ratio, which in the case of $\text{WO}_3$/W systems is reported to be $\approx 3.3–3.6$. This would approximately correspond to an increase in the film thickness by a factor 3.3–3.6 upon oxidation. We, however, observed that the film thickens by a smaller factor, i.e., $\approx 2.2–2.5$. This can be explained by considering that part of the parent W film is converted to NWs.

This sublayer oxidation is also supported by the X-ray diffraction (XRD) data (Figure 4a): the XRD pattern of the as-sputtered metallic W film shows a signal at $2\theta = 38.9^\circ$ that can be assigned to the 110 reflection of W metal. When applying the Scherrer equation, we determined for such as-sputtered W films an average crystallite size of 1.6 nm, which is well in line with the rather broad XRD peak. This peak, however, disappears after the thermal-oxidative growth, and only peaks of the W suboxide phase can be seen, confirming that no metallic W phase is left after the NW growth.

Moreover, when comparing structures grown for 1 h versus 15 h (from nominally 120 nm-thick W films), we observed that a short (1 h) thermal-oxidative treatment forms thinner and less packed NWs (Figure 3b). The sublayer thickness is $\approx 250$ nm. The nominal volume expansion (resulting sublayer thickness vs initial W film thickness) is $\approx 2.1$. A longer treatment (15 h) leads to thicker and more densely packed NWs (Figure 3d). The sublayer thickness is in this case $\approx 300$ nm, and the nominal volume expansion of the sublayer is $\approx 2.5$. One can explain these results assuming that 1) the NWs grow and thicken via consumption of W from the sublayer. This occurs via W atom
Figure 3. a,c) Top-view and b,d) cross-sectional SEM images of NWs grown at 550 °C for a,b) 1 h and c,d) 15 h (W film nominal thickness = 120 nm, Ar flow rate = 10 NL h⁻¹). Insets: Magnified top-view SEM images of the WO₃ NWs. e) Diameter and f) length statistical distributions of NWs grown for 1 h (light blue) and 15 h (green). g) Top-view and i,j) cross-sectional SEM images and h) optical image of NWs grown at 550 °C for 15 h (W film nominal thickness = 60 nm, Ar flow rate = 10 NL h⁻¹).
Figure 4. a) XRD patterns of as-sputtered W film (nominally 120 nm thick, blue line) and NWs grown in pure Ar at 550 °C for 1 h (red line) and 15 h (black line). Fitting of b,d,f) W 4f and c,e,g) O 1s XPS spectra of (b,c) as-sputtered W film on FTO and NWs grown in pure Ar at 550 °C for (d,e) 1 h and (f,g) 15 h.
diffusion from the sublayer. Simultaneously, 2) the sublayer incorporates increasingly larger amounts of oxygen, which makes the sublayer grow thicker. As for the NW growth, the oxygen source is presumably surface adsorbed oxygen species (see XPS data in Figure 4b,c). The net result of increasing the thermal-oxidative time is thicker/more densely packed NWs and a thicker sublayer.

The thermally formed WO$_{3-x}$ NWs and their parent sputtered W films were further characterized by XRD and XPS. The XRD patterns of NWs grown for different times are shown in Figure 4a. NWs grown at 550 °C for 1 and 15 h show main signals at 2θ = 23.5° and 48.0°, indexed to the 010 and 020 reflections of monoclinic W$_{18}$O$_{49}$ phase (P2/m, a = 18.334 Å, b = 3.786 Å, c = 14.044 Å; JCPDS Card No. 71-2450), respectively. Similar XRD patterns were reported, for example, for W$_{18}$O$_{49}$ NWs thermally grown at 650 °C from W films sputtered on alumina. It is difficult to unequivocally assign the XRD pattern to a specific stoichiometry, as various tungsten oxide phase (e.g., W$_{2}$O$_{19}$ or W$_{12}$O$_{55}$ among others) have similar structures, i.e., exhibit a monoclinic geometry with analogous unit cell parameters. In addition, as the most intense diffraction signal is at 23.5°, we can conclude that the NWs grow along the (010) crystallographic direction, which is consistent with previous studies on thermally grown WO$_{3-x}$ NWs. This proves that our approach forms single crystalline substoichiometric W oxide NWs that are virtually identical in terms of physicochemical and morphological features to those grown in previous work under harsher conditions.

The XPS results are in good agreement with the XRD as discussed earlier. The XPS spectra in the W 4f and O 1s regions for NWs (growth time of 1 and 15 h) and untreated W films sputter-deposited on FTO are shown in Figure 4b-g. The W 4f spectrum of W-coated FTO in Figure 4b shows peaks of W$^{6+}$ (W 4f$_{7/2}$ and W 4f$_{5/2}$ at 35.8 and 37.9 eV, respectively) and W metal (W 4f$_{7/2}$ and W 4f$_{5/2}$ at 31.6 and 33.6 eV, respectively), indicating that as-sputtered W films feature a thin air formed surface oxide layer. The thickness of this oxide layer is a few tens of nm (as determined by XPS sputter profiling; Figure S4, Supporting Information). Considering that even small traces of O$_2$ gas in the annealing atmosphere were proved detrimental for the NW growth, one can conclude that such surface air-formed oxide layer provides the oxygen source for the growth of the WO$_{3-x}$ NWs.

XPS spectra for samples treated at 550 °C for 1 and 15 h show no peak attributable to W metal, whereas a clear broadening of the W oxide signals can be observed (Figure 4d,f). This indicates the NWs to be composed of not only W$^{6+}$ but also W$^{5+}$ species. The fitting of the W 4f signals along with considerations on the different oxygen coordination of W$^{6+}$ and W$^{5+}$ centers (an average of 3 oxygen atoms for each W$^{6+}$ center and 2.5 oxygen atoms for each W$^{5+}$ center is assumed) allows to estimate the O/W ratio and stoichiometric of the NWs (more details in the Supporting Information). The resulting NWs are composed of a WO$_{3.88}$ phase, which corresponds to an oxygen vacancy concentration of ≈4.0%.

The as-sputtered W film features O 1s signals at 530.9 and 532.9 eV (Figure 4c) that correspond to the WO$_{3}$ (native oxide layer) lattice O signal and adsorbed –OH, respectively. The O 1s signals for heat-treated samples (Figure 4c,e,g) further confirm the formation of substoichiometric tungsten oxide NWs. The oxygen signals for the NWs are significantly shifted to lower binding energies (530.4 and 532.5 eV; Figure 4e,g). This indicates the presence of oxygen vacancies and defects as well as a weaker W–O interaction.

Finally, we evaluated the feasibility of growing W substoichiometric oxide NWs on other substrates. For this, we sputter-coated W metal films (thickness in the 20–120 nm range) on a W metal foil, a microscope glass slide (soda lime glass), and a layer of anodic, short aspect-ratio TiO$_{2}$ nanocavities (for the latter, see refs. [42-44]). We also coated W metal films by e-beam evaporation on a glass slide (experimental details in ref. [45]). The samples were then thermally treated in pure Ar (99.9999%) at 550 °C. The resulting structures are shown in Figure S5, Supporting Information. While it is evident that NWs could be formed on every substrate, the morphology of nanostructures, e.g., grown on porous titania or produced from evaporated W films, still requires to be improved, likely by optimizing the experimental conditions as shown earlier for the case of W films sputtered on FTO. It is reasonable to assume that the nature of the substrate (composition, crystallographic features, surface morphology) can affect some properties of the sputter-deposited W layer, e.g., the resulting surface roughness and crystallographic properties, which can, in turn, influence aspects such as the nucleation of the NWs and consequently their packing density and morphology.

These results prove nonetheless that the approach can be transferred to a palette of different substrate materials including flat metal and glass surfaces, or even porous oxide scaffolds, coated using different methods (sputtering or evaporation). Particularly, porous oxides could be of potential interest to design hybrid semiconductor nanostructures and heterojunctions for photocatalysis or photoelectrochemistry.

In conclusion, we reported on a thermal-oxidative growth approach to form arrays of W substoichiometric oxide NWs on various substrates, such as FTO, metal or glass surfaces, or nanoporous TiO$_{2}$ scaffolds. Aside from the evident versatility in terms of substrate type, the method offers various advantages: it can be conducted in a two-step approach, i.e., sputter and anneal under mild conditions (relatively low temperature, atmospheric pressure), and it requires neither a catalyst to initiate the NW growth nor the use of toxic W precursors or solvents. The possibility to operate under mild conditions could enable the NW growth on thermally sensitive materials (e.g., FTO glass preserves adequate electronic conductivity only up to 600 °C and for few minutes long treatments) or on large-scale substrates, for application in sensors, smart windows, or photoelectrodes.

**Experimental Section**

Fabrication of WO$_{3-x}$ NWs on FTO Electrodes: FTO glass slides (1.5 × 2.5 cm$^2$, Solaronix, 7 Ω cm$^{-2}$) were degreased by sonication for 15 min in acetone, ethanol, and deionized water and were then dried in a N$_2$ stream.

A plasma-sputtering machine (EM SCD 500, Leica) was used to sputter-coat W metal thin films on the FTO glasses. A 99.8% pure W target (Hauner Metallische Werkstoffe) was used for the sputtering process. The applied sputtering current was 16 mA, and the pressure of the sputtering chamber was set at 10$^{-2}$ mbar of Ar. The amount of sputtered material in terms of nominal film thickness was in situ determined by an automated quartz crystal monitor mounted inside the sputtering chamber.

After W sputter-deposition, the samples were annealed at 450–600 °C in Ar (grade 5 Ar, i.e., 99.9999% pure) with a gas flow rate of ≈10 NL h$^{-1}$ to
induce the growth of WO$_3$ NWs. The tubular furnace was purged with pure Ar (flow rate 15 NL h$^{-1}$) for 30 min prior to each thermal treatment to condition the chamber and remove atmospheric oxygen; 4.6 grade Ar (i.e., 99.996% pure), low concentration O$_2$/Ar (O$_2$ content 0.03%), and air were also used for control experiments.

Characterization of WO$_3$ NWs: A field-emission scanning electron microscope (FE-SEM; Hitachi S4800) was used to characterize the morphology of the samples. XRD with an X'pert Philips MPD (equipped with a Panalytical X'celerator detector) was used to examine the crystallographic properties of the materials. The chemical composition of the samples was analyzed by XPS (PHI 5600, USA) and peak positions were calibrated with respect to the C 1s peak at 284.8 eV. Fitting of the XPS spectra was performed by using Origin 2016 software (OriginLab, Northampton, USA) and the signals were fitted with symmetric Voigt profiles.

Supporting Information
Supporting Information is available from the Wiley Online Library or from the author.

Acknowledgements
The authors would like to acknowledge ERC, DFG, and the DFG cluster of excellence EAM for financial support, as well as H. Hildebrand, A. Friedrich, A. Knoop, and U. Marten-Jahns for valuable technical help. D.S. and S.R. gratefully acknowledge access funding enabled and organized by Projekt DEAL.

Conflict of Interest
The authors declare no conflict of interest.

Keywords
1D nanostructures, sputtering, thermal oxidation, tungsten oxide nanowires, WO$_3$ NWs.

Received: May 7, 2020
Revised: June 2, 2020
Published online: July 26, 2020

[1] G. Zheng, J. Wang, H. Liu, V. Murugadoss, G. Z. Wu, H. Che, C. Lai, H. Li, T. Ding, Q. Gao, Z. Guo, Nanoscale 2019, 11, 18968.
[2] D. Spanu, S. Recchia, S. Mohajernia, P. Schmuki, M. Altomare, Appl. Catal. B, Environ. 2018, 237, 198.
[3] M. Altomare, N. T. Nguyen, S. Hejazi, P. Schmuki, Adv. Funct. Mater. 2018, 28, 1704259.
[4] C. Fenster, A. J. Smith, A. Abts, S. Milenkov, A. W. Hassel, Electrochem. Commun. 2008, 10, 1125.
[5] C. T. Pan, C. Y. Su, C. Y. Luo, Microsyst. Technol. 2017, 23, 2113.
[6] D. Zappa, A. Bertuna, E. Comini, M. Molinari, N. Poli, G. Sberveglieri, Anal. Methods 2015, 7, 2203.
[7] K. K. Upadhyay, M. Altomare, S. Eugénio, P. Schmuki, T. M. Silva, M. F. Montemor, Electrochem. Acta 2017, 232, 192.
[8] L. Zhang, H. Wang, J. Liu, Q. Zhang, H. Yan, J. Mater. Sci. Mater. Electron. 2020, 31, 861.
[9] K. Thumavichai, N. Wang, F. Xu, G. Rance, Y. Xia, Y. Zhu, R. Soc. Open Sci. 2018, 5, 171932.
[10] J. Y. Zheng, Z. Haider, T. K. Van, A. U. Pawar, M. J. Kang, C. W. Kim, Y. S. Kang, CrystEngComm 2015, 17, 6070.
[11] C. M. Wu, S. Naseem, M. H. Chou, J. H. Wang, Y. Q. Jian, Front. Mater. 2019, 6, 49.
[12] B. Behera, S. Chandra, Mater. Sci. Semicond. Process. 2018, 86, 79.
[13] H. Zheng, J. Z. Ou, M. S. Strano, R. B. Kaner, A. Mitchell, K. Kalantar-Zadeh, Adv. Funct. Mater. 2011, 21, 2175.
[14] M. Altmare, O. Pfoch, A. Tighineanu, R. Kirchgeorg, K. Lee, E. Selli, P. Schmuki, J. Am. Chem. Soc. 2015, 137, 5646.
[15] H. G. Choi, Y. H. Jung, D. K. Kim, J. Am. Ceram. Soc. 2005, 88, 1684.
[16] X. Guo, X. Qin, Z. Xue, C. Zhang, X. Sun, J. Hou, T. Wang, RSC Adv. 2016, 6, 48537.
[17] C. Guo, S. Yin, Y. Huang, Q. Dong, T. Sato, Langmuir 2011, 27, 12172.
[18] C. H. Lu, M. H. Hon, C. Y. Kuan, I. C. Leu, RSC Adv. 2016, 6, 1913.
[19] B. J.-W. Liu, J. Zheng, J.-L. Wang, J. Xu, H.-H. Li, S.-H. Yu, Nano Lett. 2013, 13, 3589.
[20] X. Zhang, L. Gong, K. Liu, Y. Cao, X. Xiao, W. Sun, X. Hu, Y. Gao, J. Chen, J. Zhou, C. Z. Wang, Adv. Mater. 2010, 22, 5292.
[21] K. Hong, M. Xie, R. Hu, H. Wu, Appl. Phys. Lett. 2007, 90, 173121.
[22] N. Van Hieu, H. Van Vuong, N. Van Duy, N. D. Hoa, Sens. Actuators B, Chem. 2012, 171–172, 760.
[23] A. Ponzoni, V. Russo, A. Bailini, C. S. Casari, M. Ferroni, A. Li Bassi, A. Migliori, V. Morandi, L. Ortolani, G. Sberveglieri, C. E. Bottani, Sens. Actuators B, Chem. 2011, 153, 340.
[24] F. Liu, X. Chen, Q. Xia, L. Tian, X. Chen, RSC Adv. 2015, 5, 77423.
[25] F. Ghasempour, R. Azimirad, A. Amini, O. Akhavan, Appl. Surf. Sci. 2015, 338, 55.
[26] H. Qi, C. Wang, J. Liu, Adv. Mater. 2003, 15, 411.
[27] J. Qin, G. Zhang, Y. Xing, Sci. China Technol. Sci. 2012, 55, 1503.
[28] W. Zhang, M. Hu, X. Liu, Y. Wei, N. Li, Y. Qin, J. Alloys Compd. 2016, 679, 391.
[29] D. Dellasega, S. M. Pietralunga, A. Pezzoli, V. Russo, L. Nasi, C. Conti, M. J. Vahid, A. Tagliaferri, M. Passoni, Nanotechnology 2015, 26, 365601.
[30] Y. Qin, W. Xie, Y. Liu, Z. Ye, Sens. Actuators B, Chem. 2016, 223, 487.
[31] G. Gu, B. Zheng, W. Q. Han, S. Roth, J. Liu, Nano Lett. 2002, 2, 849.
[32] Z. Liu, Y. Bando, C. Tang, Chem. Phys. Lett. 2003, 372, 179.
[33] M. Furubayashi, K. Nagato, H. Moritani, T. Hamaguchi, M. Nakao, Microelectron. Eng. 2010, 87, 1594.
[34] S. C. Cifuentes, M. A. Monge, P. Pérez, Corros. Sci. 2012, 57, 114.
[35] A. Mozalev, M. Bendova, F. Gispert-Guirado, Z. Pytlicek, E. Llobet, J. Mater. Chem. A 2016, 4, 8219.
[36] C. Bagnall, J. Capo, W. J. Moorhead, Metallogr. Microstruct. Anal. 2018, 7, 661.
[37] L. Sinha, P. M. Shirage, J. Electrochem. Soc. 2019, 166, A3496.
[38] L. Yue, J. Tang, F. Li, N. Xu, F. Zhang, Q. Zhang, R. Guan, J. Hong, W. Zhang, Mater. Lett. 2017, 187, 118.
[39] Y.-M. Hsu, C.-Y. Wang, P. Chang, T.-R. Yew, Nanoscale 2015, 7, 901.
[40] B. You, S. Robin, A. Donnadieu, G. Dufour, C. Maillot, H. Roulet, C. Senemaud, Mater. Res. Bull. 1984, 19, 1349.
[41] A. M. Smith, M. G. Kast, B. A. Nail, S. Aloni, S. W. Boettcher, J. Mater. Chem. A 2014, 2, 6121.
[42] J. E. Yoo, K. Lee, M. Altomare, E. Selli, P. Schmuki, Angew. Chem. Int. Ed. 2013, 52, 7514.
[43] L. Ji, D. Spanu, N. Denisov, S. Recchia, P. Schmuki, M. Altomare, Chem. Asian J. 2020, 15, 301.
[44] D. Spanu, A. Bestetti, H. Hildebrand, P. Schmuki, M. Altomare, S. Recchia, Photochem. Photobiol. Sci. 2019, 18, 1046.
[45] M. Altmare, O. Pfoch, A. Tighineanu, R. Kirchgeorg, K. Lee, E. Selli, P. Schmuki, J. Am. Chem. Soc. 2015, 137, 5646.