A Novel Investigation of the Formation of Titanium Oxide Nanotubes on Thermally Formed Oxide of Ti-6Al-4V

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Traditionally, titanium oxide (TiO₂) nanotubes (TNTs) are anodized on Ti-6Al-4V alloy (Ti-V) surfaces with native TiO₂ (amorphous TiO₂); subsequent heat treatment of anodized surfaces has been observed to enhance cellular response. As-is bulk Ti-V, however, is often subjected to heat treatment, such as thermal oxidation (TO), to improve its mechanical properties. Thermal oxidation treatment of Ti-V at temperatures greater than 200°C and 400°C initiates the formation of anatase and rutile TiO₂, respectively, which can affect TNT formation. This study aims at understanding the TNT formation mechanism on Ti-V surfaces with TO-formed TiO₂ compared with that on as-is Ti-V surfaces with native oxide. Thermal oxidation-formed TiO₂ can affect TNT formation and surface wettability because TO-formed TiO₂ is expected to be part of the TNT structure. Surface characterization was carried out with field emission scanning electron microscopy, energy dispersive x-ray spectroscopy, water contact angle measurements, and white light interferometry. The TNTs were formed on control and 300°C and 600°C TO-treated Ti-V samples, and significant differences in TNT lengths and surface morphology were observed. No difference in elemental composition was found. Thermal oxidation and TO/anodization treatments produced hydrophilic surfaces, while hydrophobic behavior was observed over time (aging) for all samples. Reduced hydrophobic behavior was observed for TO/anodized samples when compared with control, control/anodized, and TO-treated samples. A method for improved surface wettability and TNT morphology is therefore discussed for possible applications in effective osseointegration of dental and orthopedic implants.

Key Words: thermal oxidation, anodization, nanotubes, anatase-rutile, osseointegration, Ti-6Al-4V alloy

INTRODUCTION

Titanium-6Al-4V alloy (Ti-V) is favored for orthopedic and dental implants because of its superior biocompatibility, density, mechanical resistance, corrosion resistance, atoxicity, chemical inertia, and osseointegration.1–8 Specifically, osseointegration is a strong indicator of implant success, which can be influenced by surface composition, energy, and topography.1,4,5,9–18 Surface titanium oxide (TiO₂) has been shown to improve osseointegration by means of enhanced bone morphogenetic protein signaling, apatite formation, up-regulation of collagen II, osteocalcin, biglycan, collagen I, osteopontin, and TGF-β1.12,14,19–25 TiO₂ is also known to readily react with water to form hydroxyl groups that increase surface energy and promote cellular attachment.13–16,26 In addition, the native TiO₂ layer prevents dissolution of bulk Ti-V components, aluminum (Al), vanadium (V), and titanium (Ti), into the biological environment (corrosion resistance).6,27 Aluminium ions promote scar tissue,6 V ions prevent cell division,6 and Ti ions promote osteolytic cytokines,27 which, together, can lead to implant loosening.22,28 In addition, Al and V are mutagenic, cytotoxic, and allergenic.6,29

The aforesaid surface properties can be improved by enhancing the surface native TiO₂ using atomic layer deposition (ALD), chemical vapor deposition (CVD), and thermal oxidation (TO), sandblasting, acid-etching, and anodization.3,4,6,25,26,30–38 Atomic layer deposition is a cyclic self-limiting deposition technique that is highly conformal to surface topography; using
tetrakis (diethylamino) titanium (TDEAT) precursor and H2O as oxidants, a growth rate of 0.5 Å/min of TiO2 can be achieved.39–42 Compared with ALD, CVD is a 1-step mechanism, and using TDEAT precursor and NH3 as catalysts and co-reactants results in a growth rate of up to 10 nm/min of TiO2.43 Alternatively, TO treatment can yield growth rates of TiO2 similar to CVD without the requirement of a precursor or vacuum condition and is governed by the diffusion of oxygen into bulk Ti-V.44–47 Oxygen from the air readily reacts with Ti-V at room temperature to form an oxide layer (native oxide) 2- to 8-nm thick, and a TiO2 layer 10 s of micrometer (µm) thick can be achieved within hours with TO treatment at temperatures of up to 1200°C.7,27,38,47–50 In addition, anatase and rutile crystalline structures of TiO2 formed above 200°C and 400°C, respectively, have been shown to promote basic OH/C0 groups in a biological environment (pH/C21 ≤ 6.2); basic OH/C0 groups enhance the cellular response as compared with acidic OH/C0 groups.26,51,52

TiO2 nanotubes (TNTs) increase the surface area on a cellular scale, enhancing human mesenchymal stem cell adhesion.6,25,53 TiO2 nanotubes are created by an anodic oxidation process using an electrochemical cell and an electrolyte.6,54–60 TiO2 nanotubes grow as a result of oxidation and dissolution kinetics of TiO2.59,60 For a fluorine (F)–containing electrolyte, the 4 key processes are: (1) formation of a fluoride-rich layer at the oxide-metal interface due to ion migration, (2) displacement of this layer toward the cell boundaries due to a flow mechanism, (3) dissolution of the fluoride-rich cell boundaries, and (4) formation of separated TNTs as a result of dehydration.6,60 The TNT aspect ratio (diameter/length) is affected by the applied voltage, fluorine wt. % electrolyte composition, and anodization duration.6,61,62

In addition, thermal treatment of TNTs has been shown to provide a superhydrophilic surface that improves cellular response.26,56,63 Traditional anodization studies investigate TNTs formed on Ti-V with an amorphous TiO2 surface. Ti-V, however, is often subject to TO treatment to improve mechanical properties. TO treatment initiates anatase and rutile structured TiO2 at greater than 200°C and 400°C, respectively, which may affect TNT formation. This study aims to understand the TNT formation mechanism on TO-treated Ti-V with crystalline surface TiO2 compared with that on nontreated Ti-V with amorphous surface TiO2.

MATERIALS AND METHODS

Sample preparation

Twelve Ti-V disks (diameter = 15 mm, height = 1 mm) were cut and polished from Ti-V rods (McMaster-Carr). A smooth surface (Rav = 10 ± 2 nm) was achieved by (1) wet grinding using a series of Carbimet 2 silicon carbide grinding papers (Nos. 200, 320, 400, 600, and 800; Buehler, Lake Bluff, Ill); (2) polishing using a TexMet Polishing Cloth (Buehler), MetaDi 9-micron diamond paste (Buehler), and MetaDi Fluid lubricant (Buehler); and (3) fine-polishing using a Chemomet I polishing cloth (Buehler) with MasterMed colloidal silica polishing suspension (Buehler). Samples were washed with deionized (DI) water (17.7 MΩ cm, Barnstead NANOPure) followed by drying with N2 gas spray (grade 4.8, 99.998%; Progressive Industries, Inc, Chicago, Ill).

A Lindberg furnace (S# 54032) was used for TO treatments; a k-type thermocouple was placed at the center of the furnace quartz tube (35 cm from opening) and connected to a
temperature controller (JLD-612) to maintain working temperatures of 300°C and 600°C (Figure 1a). The samples were loaded into the quartz tube to the center of the furnace at a gradual insertion rate of 5 cm every 5 minutes to prevent micro-cracks from thermal shock. Similarly, samples were gradually extracted after TO treatment at a rate of 5 cm every 5 minutes to prevent micro-cracks from an abrupt cool down. Using existing computational and experimental data, the diffusion depth of oxygen was calculated for the working temperatures of 300°C and 600°C. Based on calculations, oxygen is expected to diffuse 60 nm into Ti-V at 300°C TO treatment, and 2000 nm into Ti-V at 600°C TO treatment. Between 10% and 20% of the calculated diffusion depth is expected to be composed of TiO₂, or 12 nm and 400 nm for 300°C and 600°C TO treatment, respectively. 

For anodization, a voltage source (Keithley 2400 Source-Meter) was connected to the Ti-V samples (working electrode) and a copper counter electrode (Figure 1b) in the electrolyte of ethylene glycol (96.15 mL), 4.0 vol. % DI water (3.85 ml), and 0.2 wt. % NH₄F (0.21 g). Ti-V samples were anodized at 60 V at room temperature for 15 minutes. After anodization, samples were washed with DI water, air dried, wrapped in sterile tissue (Kimwipe, Kimtech Science, Roswell, Ga), and stored in a glass petri dish (KIMAX® Petri Dish).

The experimental groups are control, 300°C TO treated (TO-300), and 600°C TO treated (TO-600), control/anodized (CTNT), 300°C TO/anodized (300TNT), and 600°C TO/anodized (600TNT).

**Surface characterization**

Sample surface characterization was carried out by means of field emission scanning electron microscopy (FESEM), energy dispersive x-ray spectroscopy (EDS), water contact angle (WCA) measurements, and white light interferometry (WLI).

**Field Emission Scanning Electron Microscope**

A JEOL JSM-6320F FESEM was used to characterize sample surfaces after TO and TO/anodization treatments. Anodized sample surfaces were scraped to remove TNTs onto double-sided conductive carbon tape, which was mounted on an aluminum stub for imaging. All TNT images were produced with 30,000 magnification and a scale of 100 nm. Image software was used to measure TNT dimensions.

**Energy Dispersive X-Ray Spectroscopy**

An EDS attachment of the JEOL JSM-6320F FESEM was used for the chemical characterization of TO-treated and TO/anodized surfaces. Random locations on control and TO-treated samples and the top, center, and bottom of randomly selected TNTs from anodized surfaces were analyzed. All scans were carried out with 30,000 magnification. An acceleration voltage of 10 keV was used for all EDS scans; emission energies of aluminum (Al), titanium (Ti), and vanadium (V) are 1.486, 4.508, and 4.949 keV, respectively, and 2 times the highest emission energy of elements of interest is recommended.

**WCA Measurements**

A micro-syringe (Hamilton, 802RN) was used to place 5 μL DI water sessile droplets on TO-treated and TO/anodized Ti-V surfaces, and a goniometer (Rame-Hart NRL CA) was used to measure the WCA of the droplets. After TO and TO/anodization treatments, all samples were wrapped in sterile tissue for storage. Before WCA measurement, samples were removed from storage and sprayed with N₂ gas for 30 seconds to remove sterile tissue residue. After WCA measurements, samples were returned to sterile tissue wrap for storage. The WCA measurements were taken at the following stages: (1) immediately after TO treatment; (2) after a 4-week aging period following TO treatment; (3) after DI water wash following the aging period; (4) immediately after anodization; (5) over a 2-week aging period following anodization on days 1, 2, 7, and 14; and (6) after DI water wash following day 14 of the aging period. For the aging period TO/anodized samples (stage 5), samples were removed from sterile tissue wrap and sprayed with N₂ gas for 30 seconds to remove residue prior to WCA measurements. After WCA measurements, samples were returned to sterile tissue wrap for storage.

**White Light Interferometry**

White light interferometry (NewView 6300, Zygo Corporation, Middlefield, Conn) was used to carry out surface roughness measurements before and after TO treatment, provided as average roughness values (Ra). Surface roughness measurements were not carried out on anodized samples. Because of the morphology of TNTs, light does not propagate to the bottom of the TNT structure; therefore, accurate roughness measurements cannot be taken.

**Statistical analysis**

OriginPro 9.0 was used for Tukey 1-way analysis of variance to determine the statistical significance in surface wettability, surface roughness, and TNT dimensions. For all analyses, P < 0.05 was considered statistically significant.

**Results**

The FESEM images of TO-treated and TO/anodized samples are presented in Figure 2. No difference in surface roughness was observed for control and TO-300 samples (Figure 2a and b, respectively); a difference in surface roughness was observed between control and TO-600 samples (Figure 2a and c, respectively). The WLI data show a significant difference between the TO-600 surface and both control (P < .001) and TO-300 (P < .001) surfaces; Ra was measured to be 10 ± 2, 11 ± 2, and 29 ± 4 nm for control, TO-300, and TO-600, respectively (Figure 3). While surface pores were visible on control and 300TNT surfaces, very few pores were visible on 600TNT surfaces, forming a cover layer with TNTs present underneath (Figure 4a–c). At the top of all TNTs, a fuzzy-textured layer of 100 ± 15-nm thickness was observed (Figure 4). The TNT separation within this layer was less distinct, and surface pore diameters (surface inner diameter) were smaller compared with inner diameters below this layer (ID).

Energy dispersive x-ray spectroscopy was used to quantify the elemental composition of TO/anodized surfaces. Scans of top, center, and bottom of randomly picked TNTs were conducted, and no difference in elemental composition was...
The TNT O:Ti ratios of 2.0, 2.1, and 1.8 were seen for CTNT, 300TNT, and 600TNT surfaces, respectively, indicating the presence of TiO$_2$ in all of the analyzed TNTs. An example EDS scan is given in Figure 5. Energy dispersive x-ray spectroscopy of TO-treated surfaces was not successful.

Measurements of TNT length, surface inner diameter, inner diameter, and outer diameter (OD) are given in Figure 6. Surface inner diameters were measured as $30 \pm 6$ nm and $34 \pm 7$ nm for the CTNT and 300TNT groups, respectively, and surface inner diameters on 600TNT samples were too few and too small to be accurately measured. While the surface inner diameters of CTNT and 300TNT were not significantly different ($P = .151$), surface pores were visibly fewer on 300TNT. Inner diameters under the surface were not significantly different between the groups ($72 \pm 13$ nm, $76 \pm 8$ nm, and $78 \pm 14$ nm for CTNT, 300TNT, and 600TNT groups, respectively). Outer diameters under the surface for 600TNT ($136 \pm 20$ nm, $P = .932$) samples were not significantly different from CTNT samples ($134 \pm 13$ nm); however, a significant difference was observed for 300TNT samples ($120 \pm 19$ nm, $P = .004$). The TNT lengths of the 300TNT group ($2.01 \pm 0.12 \mu m$) were observed to be significantly longer as compared with the CTNT and 600TNT groups ($1.36 \pm 0.10 \mu m$, $P < .001$, and $1.51 \pm 0.18 \mu m$, $P < .001$, respectively).

The WCA measurements for TO-treated and TO/anodized samples are given in Figure 7. Immediately after TO and TO/anodization treatments, hydrophilic behavior was observed on all sample surfaces ($<10^\circ$). Aging, however, increased hydrophobic behavior for both TO-treated and TO/anodized groups. Hydrophobic behavior was observed for TO-treated samples within days and remained steady up to 4 weeks for all TO-treated groups ($61^\circ \pm 2^\circ$). DI water wash after the aging period led to a decrease in WCA to $41^\circ \pm 2^\circ$ ($P < .001$), $35^\circ \pm 3^\circ$ ($P < .001$), and $15^\circ \pm 4^\circ$ ($P < .001$) for control, TO-300, and TO-600 samples, respectively. Over 2 weeks, the 300TNT and 600TNT groups aged to lesser extents ($49^\circ \pm 5^\circ$ and $57^\circ \pm 11^\circ$, respectively) than the CTNT group ($73^\circ \pm 9^\circ$). For CTNT and 300TNT samples, the DI water wash after the aging period led to a significant decrease in WCA to $39^\circ \pm 4^\circ$ ($P < .001$) and $34^\circ \pm 3^\circ$ ($P < .001$), respectively. The DI water wash after the aging period of the 600TNT samples did not lead to a significant decrease; WCA decreased from $57^\circ \pm 11^\circ$ to $48^\circ \pm 12^\circ$ ($P = .107$).

**DISCUSSION**

**Surface roughness**

The FESEM images and WLI data of TO-treated samples (Figures 2a–c and 3, respectively) indicate a correlation between TO treatment temperature and roughness, as seen in the literature.\textsuperscript{64} Previously, majority anatase and majority rutile in TO-300 and TO-600 Ti-V samples, respectively, were observed with Fourier transform infrared spectroscopy (not shown). The surface roughness value of the TO-600 surface was significantly higher than that of control ($P < .001$) and TO-300 ($P < .001$) surfaces because rutile TiO$_2$ formation results in significant
changes in surface morphology while no changes occur for as-is amorphous and anatase TiO₂ of TO-300 samples.

**Variations in elemental composition**

Energy dispersive x-ray spectroscopy was used to confirm the elemental composition of the TO-treated and TO/anodized surfaces: there was no observable difference in the TNT composition between the anodized groups. The O:Ti ratio was calculated to be 2.0, 2.1, and 1.8 for CTNT, 300TNT, and 600TNT, respectively, indicating the presence of TiO₂. Energy dispersive x-ray spectroscopy of TO-treated–only surfaces was not possible because of the low thickness of the TiO₂ layers; the electron beam (e-beam) of 600 nm was greater than the oxides of control and TO-treated surfaces; therefore, only the bulk Ti-V components (Ti, Al, and V) were present in the spectrums.⁶⁵,⁶⁶

**Variations in WCA**

The WCA measurements carried out immediately after TO treatment were superhydrophilic (WCA ≤ 5°) as expected (Figure 7).¹³⁻¹⁶ The crystalline structure of TiO₂ promotes dissociation of H₂O molecules, leading to hydroxylation of the surface.¹⁶,⁶⁷,⁶⁸ Hydroxylation attracts H₂O molecules, forming OH-H₂O complexes, which are water adsorption sites.⁶⁸ Stored control and TO-treated samples experienced hydrophobic behavior, or aging, (WCA = 61° ± 2°), as reported in the literature,⁶⁹ with a significant decrease in WCA after DI water wash; WCA decreases to 41° ± 2° (P < .001), 35° ± 3° (P < .001), and 15° ± 4° (P < .001) for control, TO-300, and TO-600 samples, respectively. Similar to hydroxylation, TiO₂ crystallinity also delays dehydroxylation (aging) of the surface TiO₂ because of the higher affinity to maintain hydroxylation. This phenom-
enon is also evident in TO/anodized samples and is discussed subsequently.

As can be seen in Figure 7, TO/anodization treatments also produced hydrophilic surfaces \((<10°)\). Anodized samples show hydrophilic behavior due to the increased surface area available for water contact on the nanotubes.\(^6,\)\(^25,\)\(^53\) While aged anodized groups displayed hydrophobic behavior, 300TNT and 600TNT groups aged to a lesser extent \((49° ± 8° \text{ and } 57° ± 11°, \text{ respectively})\) compared with the CTNT group \((73° ± 9°)\). In addition, the hydrophobic behavior of TO-treated samples has been observed within a few days as compared with the gradual increase in hydrophobic behavior of TO/anodized groups over 2 weeks; hydroxylation and dehydroxylation equilibrium takes longer to reach on the greater TNT surface area.\(^6,\)\(^25,\)\(^53\) It may also be possible that the 300TNT TNTs may be of crystalline nature, therefore diminishing the aging process\(^13–16\) as compared with CTNT and 600TNT; the observed
difference in surface morphology (Figure 2d–f) may also be due to the crystalline TiO$_2$.

The DI water wash after the aging period led to a significant decrease in WCA for CTNT and 300TNT samples (WCA decreased from 73° ± 9° and 49° ± 5° to 39° ± 4° [P < .001] and 34° ± 3° [P < .001], respectively). The greater surface area of the CTNT and 300TNT surfaces is attributed to the decrease in the WCA; CTNT and 300TNT after-wash WCA are similar, indicating an equivalent air entrapment behavior that limits how low WCA decreases, irrespective of the anatase or amorphous TNT structure. The decrease in WCA after the DI wash for 600TNT samples was not significant (WCA decreased from 57° ± 11° to 48° ± 12°); the reduced hydroxylation from the DI water wash was due to less surface area available on the 600TNT surface.

**TNT morphology**

A denser distribution of surface pores is observed on the CTNT surface as compared with the 300TNT surface (Figure 2d and e); the 600TNT surface has very few surface pores and is nearly covered by a TiO$_2$ cover, possible remnants of the TO-formed TiO$_2$ (Figure 2f). The TNTs formed on the 300TNT surfaces acquire lengths of 2.01 ± 0.12 μm as compared with 1.36 ± 0.10 μm and 1.51 ± 0.18 μm for TNTs on the CTNT and 600TNT surfaces, respectively. As seen in Figure 8, an equilibrium between oxidation/dissolution of the barrier-like oxide layer (BOL) in the presence of fluoride ion leads to the formation of TNTs (see steps 4 and 5 below). For the oxidation part, a BOL formed as a precursor to TNT formation may explain the difference in morphology (Figure 8a). From Kowalski et al., Minagar et al., and Chen et al., TNTS likely form due to the following anodic reactions occurring at the Ti/oxide and oxide/electrolyte interfaces:

\[
\text{H}_2\text{O(l)} \rightarrow \text{H}^+ + \text{OH}^- \quad (1)
\]

\[
\text{OH}^- \rightarrow \text{H}^+ + \text{O}_2^- \quad (2)
\]

\[
\text{Ti}(s) + 4\text{OH}^- \rightarrow \text{Ti(OH)}_4^+ + 4e^- \quad (3)
\]

\[
\text{Ti}(s) + 2\text{O}_2^- \rightarrow \text{TiO}_2(s) + 4e^- \quad (4)
\]

\[
\text{TiO}_2(s) + 6\text{F}^- + 4\text{H}^+ \rightarrow [\text{TiF}_6]^{2-} + 2\text{H}_2\text{O} \quad (5)
\]

At the end, TNTs are composed of a mixture of TiO$_2$ and Ti(OH)$_4$; the dehydralation of anodized samples after removal from electrolyte leads to TNT separation. In preliminary work, the BOL is observed to form during the first 5 minutes of anodization until the oxidation/dissolution equilibrium condition is met (Figure 8c). Compared with CTNT samples, a shorter BOL formation step or accelerated TNT formation may explain the longer TNTs of 300TNT samples. H$^+$ ion production/accumulation occurs during BOL formation, attracting F$^-$ ions necessary for dissolution of the BOL surface to form a nanotubular structure (Figure 8a). Dissolution of the BOL at the bottom of the nanotubular structure exposes the underlying bulk Ti-V to the electrolyte, where new TiO$_2$ complexes are formed (Figure 8c) and the cycle of oxidation/dissolution is repeated. Namely, there are 2 steps in which TO-formed TiO$_2$ may affect the anodization mechanism: (1) H$^+$ production and (2) the transport of F$^-$ ions to TiO$_2$ complexes.

**H$^+$ Production and Accumulation**

Compared to Ti-V with amorphous TiO$_2$, the increased H$_2$O dissociation on anatase and rutile surfaces of TO-300 and TO-600 surfaces, respectively, may produce more H$^+$ ions, attract more F$^-$ ions, and lead to oxidation/dissolution equilibrium more quickly. While this explains longer TNTs on the 300TNT surface, FESEM images of the 300TNT and 600TNT surfaces (Figure 2e and 2f) show insufficient dissolution on the surface; crystallinity is not the only factor in the observed difference in TNT morphology. The oxide formed by 600°C TO treatment may be too thick, delaying the point at which oxidation/dissolution equilibrium is reached; 400 nm of TiO$_2$ and 1600 nm of the oxygen diffusion zone are expected for 600°C TO treatment. The oxygen diffusion zone has less than the 2:1 O:Ti ratio. Because of the oxide and diffusion zone, oxidation may occur at lower rates (reduced H$^+$ ion production), dissolution is dominant, and therefore oxidation/dissolution equilibrium is delayed. As a result, TNTs on 600TNT samples are shorter compared to the presence of TNTs on 300TNT samples. The TNTs on 600TNT samples are longer than on CTNT samples, however, suggesting that anodization is faster due to the rutile TiO$_2$ once oxidation/dissolution equilibrium is reached.

**F$^-$ Ion Transport**

Crystalline TiO$_2$ is known to be more reactive than amorphous TiO$_2$, and the diffusivity of TiO$_2$ increases when doped with nonmetals such as fluorine. The minority anatase {001} facets experience dissociative adsorption of reactant molecules, are preferentially fluorinated, and therefore are energetically stabilized. F-terminated {001} facets reduce the activation energy of reactant molecules and affect the reaction mechanism at the molecular level. In addition, Wagemaker et al. suggests that diffusion of foreign ions into anatase occurs at a higher rate. Although rutile TiO$_2$ has higher diffusion rates for distances of μm when compared with amorphous and anatase TiO$_2$, F$^-$ ion transport is reduced for larger distances because of ineffective grain stacking (grain boundaries); the dissolution mechanism is therefore slowed, and oxidation/dissolution equilibrium is delayed.

**Surface pores**

As TNTs become longer, the F$^-$ ions accumulate at the bottom of TNTs, where new H$^+$ ions are produced via TiO$_2$ production and dissolution at the surface ceases. Based on these phenomena, the more TiO$_2$ that exists before anodization, the more TiO$_2$ will remain after anodization as this fuzzy layer. Even the highly water-dissociative property of rutile structure formed during 600°C TO treatment is not sufficient in the dissolution of the thick TiO$_2$ layer; water dissociation produces H$^+$ ions, which attract F$^-$ ions for the dissolution of TiO$_2$. Increasing the water content in the electrolyte solution, however, may accelerate the removal of this fuzzy layer before oxidation/dissolution equilibrium is reached.
In summary, rutile TiO₂ may provide the best wettability behavior for improved cellular response. While anatase TiO₂ (TO-300) produced the longest TNTs, a thin rutile TiO₂ layer may produce even longer TNTs. To compensate for the reduced anodization on thick rutile TiO₂ (TO-600), we may have to use TO duration as short as 15 minutes to produce rutile TiO₂ on Ti-V with a thickness similar to that of TO-300 samples. Surface pore size can be maximized by increasing electrolyte water content to enhance dissolution of TiO₂ on the TNT surface. The combined benefit of increased TNT lengths, surface pores, and improved wettability can improve cellular response and the shelf life of dental and orthopedic implants and create opportunities for drug and cellular media-loading applications.

**ABBREVIATIONS**

ALD: atomic layer deposition  
BOL: barrier-like oxide layer  
CVD: chemical vapor deposition  
EDS: energy dispersive X-ray spectroscopy  
FESEM: field emission scanning electron microscopy  
TDEAT: tetraakis(diethylamino)titanium  
Ti-V: Ti-6Al-4V alloy  
TNT: TiO₂ nanotube  
TO: thermal oxidation  
WCA: water contact angle  
WLI: white light interferometry

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**REFERENCES**


