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# Corrosion Performance of Medical Grade NiTi after Laser Processing

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## Abstract

Laser processing of shape memory alloys (SMAs) has great potential to enable the multifunctionality required for complex applications. Achieving this improved functionality often requires laser processing of the SMA's with a high-energy pulse coupled with an observable decrease in the corrosion resistance, however there are discrepancies in the post-processing corrosion performance. The current study conducts a systematic investigation of Ni-44.2 wt.% Ti SMA with differing numbers of laser pulses and post-process surface treatments. The localized electrochemical characteristics were investigated by scanning electrochemical microscopy. X-ray photoelectron spectroscopy and Raman spectroscopy was used to determine surface composition and oxide crystallinity, while oxide stability was determined via potentiodynamic cyclic polarization. Results showed that fewer pulses were not significantly detrimental to the corrosion performance. However, increasing the number of pulses had a significant impact on oxide stability in the heat affected zone due to increased crystallinity. The post-process surface treatments restored corrosion performance to pre-processed capabilities; however, further optimization is required to achieve maximum corrosion resistance.

Keywords: Nickel-Titanium; Laser Processing; Corrosion; NiTi; Shape Memory Alloy (SMA); Surface Treatments;

## 1.0 Introduction

The shape memory effect (SME), pseudoelastic effect (PE), and outstanding mechanical properties of NiTi shape memory alloys (SMAs) have made them valuable engineering materials [1, 2]. Industries interested in applying shape memory alloys (SMA) range from automotive to aerospace to microelectronics. Additionally, the excellent biocompatibility of NiTi based SMAs makes them increasingly sought-after for numerous biomedical applications [3-6]. Devices implementing SMA's are becoming increasingly complex and smaller. This requires various laser modification techniques designed to combine different SMA properties into one system [7-10].

Prior studies examining effects of laser processing on the corrosion performance of NiTi SMAs have reported different, and sometimes contradictory, findings. Several publications have shown corrosion performance of NiTi to improve significantly after laser processing [11-13], while some publications report only marginally better performance [14, 15]. In contrast, publications have also shown reduced corrosion performance as compared to the original base material [16, 17]. Only a few of these studies attempt to characterize the surface after laser processing to determine how the oxide has changed to confirm the mechanisms behind these changes in corrosion performance [12, 16, 17] and very little research has been done to restore the corrosion performance to pre-processed conditions. For example, Cui et al. [13] characterized mechanically polished NiTi before and after laser processing by XPS and cyclic polarization. They found that laser processing increased the percent of Ti oxide present on the surface which increased corrosion resistance. Chan et al. [17] also characterized NiTi after laser processing by XPS and cyclic polarization, but they found little compositional differences between the processed region and the base material with only a slight decrease in the corrosion performance. However, another study by Chan et al. [16] observed grain recrystallization in the heat affected zone (HAZ) after

laser processing; leading to decreased corrosion resistance in the HAZ region. They also observed that the corrosion performance was enhanced slightly due to an increase in the thickness of the oxide layer [17]. However, this post-process heat treatment was only tested on the base material or weld zone separately, never in a combined system, neglecting the corrosion performance in the heat affected zone (HAZ). In a study by Pequegnat et al. [18] preferential corrosion was observed adjacent to overlapping pulsed laser spots. Therefore, it is necessary to determine what factor leads to these differing results in corrosion performance.

In order to assess the reported discrepancies in corrosion performance after laser processing of NiTi, [13, 16, 17] we provide an in-depth study that tracks the surface oxide evolution after pulsing with a Nd:YAG laser. Individual laser spots with increasing numbers of pulses were examined to determine how increasing the total energy input to the work material alters the local surface oxide properties. The topography variations were analysed by optical profilometry whereas the localized corrosion performance of the oxide layer was quantified by scanning electrochemical microscopy (SECM) and surface morphology was analyzed by X-ray photoelectron spectroscopy (XPS) and Raman spectroscopy. Potentiodynamic cyclic polarization testing was performed on samples processed with overlapping spots, which provided a large enough surface area to properly determine the overall corrosion performance of the laser pulse material and the ability of post-process surface treatments to restore corrosion resistance.

## **2.0 Experimental Section**

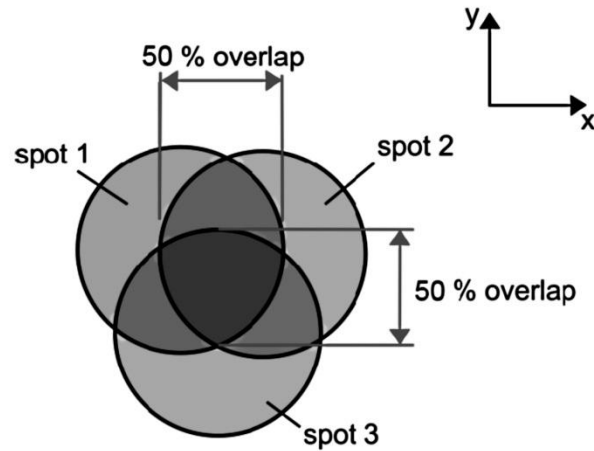
### **2.1 Sample Preparation**

This study used 0.37 mm thick by 12 mm wide SE508 NiTi strip commercially available through Nitinol Devices and Components (NDC) Incorporated. The nominal chemical composition was

55.8 wt.% (50.7 at.%) Ni and 44.2 wt.% (49.3 at.%) Ti with a maximum O and C content of 0.05 wt.% and 0.02 wt.%, respectively, as reported by the manufacturer. Prior to testing, the NiTi strip was ultrasonicated in a 7.5 vol. % HF, 20 vol. % HNO<sub>3</sub>, bal. H<sub>2</sub>O solution for 2 minutes, flipping half way, to remove the thick black oxide layer that forms during manufacturing. After chemically etching, the cross-sectional thickness was measured to be 0.35 mm. For the remainder of this study, the chemically etched state is referred to as ‘CE.’

A LW50A Miyachi Unitek pulsed Nd:YAG laser system with a 1.06  $\mu\text{m}$  wavelength, 600  $\mu\text{m}$  nominal spot diameter was used in this study. A flow rate of 14.2 L min<sup>-1</sup> (30 CFH) of argon shielding gas was found to be sufficient to prevent excessive oxidation of the sample. The CE strip was cleaned with acetone, ethanol, and then de-ionized water prior to laser processing to remove any potential surface contaminants. A pulsed laser profile with a peak power of 0.8 kW for 6 ms followed by a linear ramp down to 0.0 kW over 4 ms was used [10, 18].

The laser spots were pulsed one, three and five times to investigate the change in surface reactivity with increased laser pulsing. This material state will be referred to as the “laser processed” (LP) state for the remainder of this study with a number preceding ‘LP’ indicating number of pulses used. Additional samples were processed such that each spot was pulsed 5 times, with consecutive spots overlapping by approximately 50 % to eliminate gaps between spots, as shown in Figure 1. This melts the sample to a depth of  $\sim 300 \mu\text{m}$  and alters the local composition, as was shown in a previous study [18]. These samples were made to test the effectiveness of post-process surface treatments to restore corrosion resistance. This material state will be referred to as the “laser pulsed” (LP) state for the remainder of this study. After laser processing, samples were cleaned again with acetone, ethanol and de-ionized water prior to any testing or post-processing treatments to remove any surface deposits from laser processing.



**Figure 1: Illustration of laser spot configuration with 50 % overlap.**

Some of the overlapped LP sample were treated after laser processing to restore the corrosion resistance. First, a mechanically polished surface was tested to determine if the reduced corrosion performance was a result of the surface morphology or the change in composition. After laser processing, strips were cut in  $8 \times 8 \text{ mm}^2$  pieces and progressively ground with 600, 800, 1200 coarse, and 1200 fine grit silicon carbide paper. Polishing was then performed with a  $0.03 \text{ }\mu\text{m}$  colloidal silica suspension and 30% hydrogen peroxide mixture (9:1 ratio). Extreme care was taken to use consistent grinding/polishing protocol for each individual specimen. This material state will be referred to as the “laser processed and mechanically polished” (LPMP) state for the remainder of the study. The second post-process treatment tested was chemically etching. The samples were immersed in a mixture of 15 vol.% HF, 40 vol.%  $\text{HNO}_3$ , and 45 vol.% de-ionized water for 30 seconds, and then immediately ultrasonicated in de-ionized water for 10 minutes. This material state will be referred to as the “laser processed and chemically etched” (LPCE) state for the remainder of the study.

## 2.2 Instruments and Procedures

The surface topography was characterized using the Nanovea ST400 Optical Profilometer, which is a 3D non-contact profiler that was not affected by sample reflectivity. This is important due to the variation in reflectivity across the sample surface from retained base material to laser spots. Three-dimensional topographical surface maps of the samples were obtained by rastering the optical pen over a  $1.2 \times 1.2 \text{ mm}^2$  area using a step size of  $5 \text{ }\mu\text{m}$ . High-magnification optical microscopy was done with an Olympus BX51M upright inspection and research microscope. Low-magnification optical microscopy was performed on a Leica A60 S Stereomicroscope and a PAX cam LM2 digital camera using Seagull image capture software.

The SECM experiments were carried out using a HEKA scanning electrochemical microscopy (model ELP3, HEKA Elektronik, Germany). The laser processed NiTi sample was fixed to the bottom of a Teflon cell. A  $25 \text{ }\mu\text{m}$  diameter platinum microelectrode (ME) was used as the working electrode. A chloridized silver wire served as a quasi-reference electrode (Ag|AgCl-QRE) and a  $0.5 \text{ mm}$  diameter platinum wire was used as a counter electrode (Goodfellow Cambridge Limited, Huntingdon, England). MEs were fabricated in-house by sealing Pt wires ( $25 \text{ }\mu\text{m}$  diameter, Delta Scientific Laboratory Products Ltd., Canada) into borosilicate glass capillaries with an outer diameter of  $1.5 \text{ mm}$  and an inner diameter of  $0.7 \text{ mm}$  (Sutter Instrument, USA). The probe tip was subsequently sharpened to a ratio  $RG = r_{\text{glass}} / r_{\text{T}} \approx 10$ , where  $r_{\text{glass}}$  is the radius of the insulating sheath and  $r_{\text{T}}$  is the radius of the active ME. The resulting Pt MEs were polished on a microcloth pad (Struers MD Chem cloth) using a series of water-based alumina slurries (i.e.  $1 \text{ }\mu\text{m}$ ,  $0.3 \text{ }\mu\text{m}$  and  $0.05 \text{ }\mu\text{m}$ ). The working solution was composed of  $1 \text{ mM}$  FcMeOH as redox-active mediator and  $0.1 \text{ M}$   $\text{K}_2\text{SO}_4$  as the supporting electrolyte in ultra-pure water (Millipore MiliQ water  $18.2 \text{ M}\Omega$ ).



SECM feedback images ( $1.2 \times 1.2 \text{ mm}^2$  area) were acquired with the ME probe positioned at a tip to substrate distance of  $10 \text{ }\mu\text{m}$  relative to highest topography feature on the surface and using a  $x, y$ -scan rate of  $5 \text{ }\mu\text{m s}^{-1}$ . SECM quantitative feedback approach curves used a  $z$ -scan rate of  $2 \text{ }\mu\text{m s}^{-1}$ . During imaging and acquisition of quantitative feedback approach curves in  $1 \text{ mM FcMeOH}/0.1 \text{ M K}_2\text{SO}_4$  solution, the NiTi was not polarised. The ME was polarized at  $E_T = 0.4 \text{ V vs Ag|AgCl-QRE}$  to oxidize FcMeOH at a mass-transfer limited rate. The diffusion coefficient  $D$  of the redox-active mediator was experimentally determined to be  $7.6 \times 10^{-6} \text{ cm}^2 \text{ s}^{-1}$ .

X-ray photoelectron spectroscopy (XPS) was conducted using a K-Alpha XPS manufactured by ThermoFisher Scientific using monochromatized Al K $\alpha$ 1 radiation. The source was run at  $50 \text{ eV}$  with a spot size of  $30 \text{ }\mu\text{m}$ . Survey scans were taken at a pass energy of  $200 \text{ eV}$  and a binding energy range of  $0 - 1300 \text{ eV}$ . The Raman spectrometer used for this experiment was a Renishaw Raman Microscope with a resolution of  $\leq 1 \text{ cm}^{-1}$ . The spectra were obtained with a He-Ne laser operating at  $633 \text{ nm}$  with an incident power of  $7.5 \text{ mW}$  and a  $5 \text{ }\mu\text{m}$  spot size. The spot size reported is the focal spot size of the laser used in this study. The integration time for all Raman spectra is  $10 \text{ s}$ . The wavelength calibration was made using the silicon emission line at  $520 \text{ cm}^{-1}$ . Raman spectra were replicated four times with good repeatability for each surface condition.

Potentiodynamic cyclic polarization tests were performed in a phosphate buffered saline (PBS, pH 7.4, MP Biomedicals) electrolyte. Using PBS for cyclic polarization testing is common [18-21], and B.G. Pound et al. [22] has shown that the breakdown potentials of NiTi in PBS are similar to Hanks' solution. Controlling and recording the test was done with a GAMRY Instruments potentiostat using Framework v4.35 software. A  $250 \text{ mL}$  double walled corrosion cell with a three-electrode configuration was used per the ASTM F2129-08 standard [23]. A saturated calomel electrode (SCE) was used for the reference electrode and a Pt mesh was used

for the counter electrode. All potentials in this study will be presented with respect to SCE. The working electrode consisted of an exposed area of the specimen ( $0.25 \text{ cm}^2$ ) using an area limiting gasket in a custom holder. Tests were performed at  $37^\circ\text{C}$  to simulate biological conditions. The solution was de-aerated with nitrogen for 30 minutes before the tests were initiated. The open circuit potential (OCP) was measured until it was stable within  $0.05 \text{ mVs}^{-1}$ , for a maximum of one hour. Testing was performed at a scan rate of  $0.167 \text{ mVs}^{-1}$  per the ASTM G61-86 [24], starting at  $-600 \text{ mV}$  and increasing to  $1,400 \text{ mV}$ , automatically reversing if the current density exceeded  $0.1 \text{ mAcm}^{-2}$ . Five specimens were tested for each material condition and representative curves are provided.

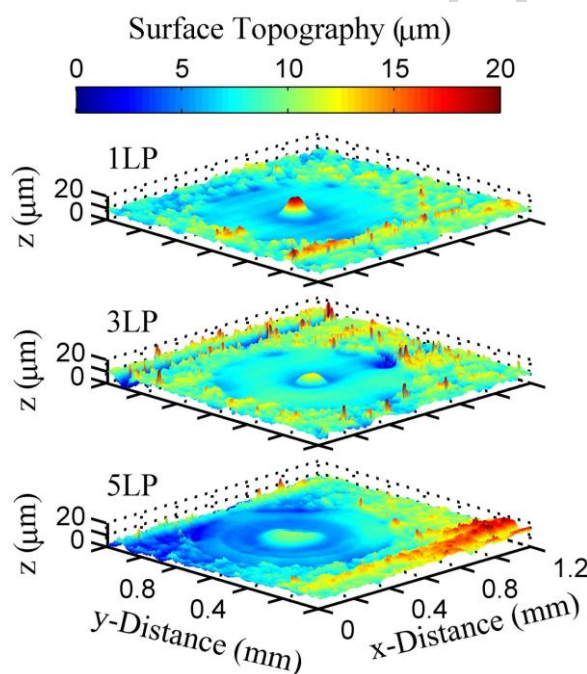
### 3.0 Results and Discussion

In this comprehensive study, the results presented were obtained from experiments designed to thoroughly discover the localized surface conditions of the CE and LP samples, followed by the resulting corrosion performance of CE and LP samples, and finally the effect of post-process surface treatments. First, the surface reactivity was characterized by SECM to identify localized regions of increased reactivity after laser processing. Next, the surface constituents were characterized by XPS and Raman spectroscopy to identify the species present and the degree of crystallinity. Finally, the overall corrosion performance was evaluated using cyclic potentiodynamic polarization testing.

#### 3.1 Surface Morphology and Reactivity

The 3D topographical profiles of the individual spots are shown in Figure 2. The surface roughness values ( $R_a$ ) values were calculated to be  $1.45 \text{ }\mu\text{m}$  (1LP),  $1.71 \text{ }\mu\text{m}$  (3LP),  $2.54 \text{ }\mu\text{m}$  (5LP). The surface roughness values of all the LP samples were comparable to the roughness

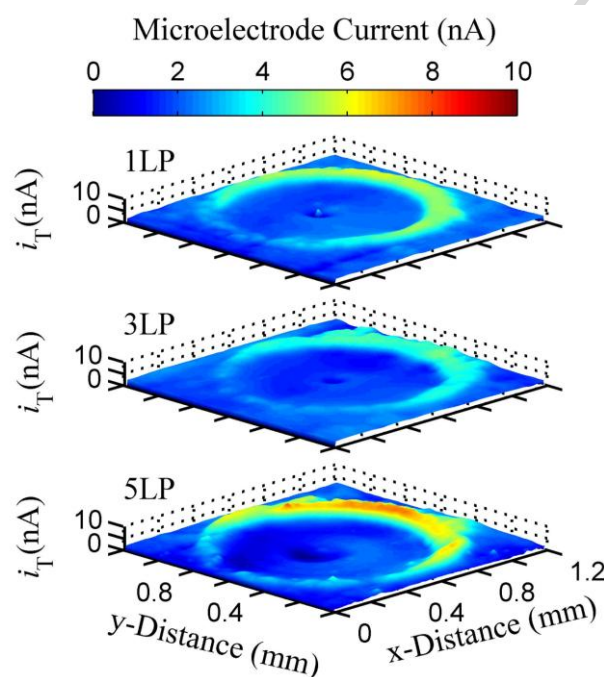
value of the CE surface ( $2.18\ \mu\text{m}$ ); however, the laser spots were found to be smoother than the chemically etched material surrounding it. This result is similar to the findings of other laser processing studies [12]. There was a slightly elevated region in the center of all the spots. This is due to the pressure drop associated with keyhole-mode laser processing. Ripples were also observed in the re-melted region, which increase in intensity with the number of pulses. Analysing surface morphology prior to SECM analysis was crucial to ascertain the extent that the surface topography affects the SECM feedback measurements.



**Figure 2: Surface topography maps of 1LP, 3LP, and 5LP.**

Figure 3 shows the SECM images of the NiTi sample surfaces recorded in the feedback mode in a solution containing  $1\text{mM FcMeOH}/0.1\text{ M K}_2\text{SO}_4$ . When positioned in the bulk solution (greater than  $1\text{ mm}$  from the substrate) a steady state limiting current of  $3.15\text{ nA}$  was recorded. It was observed that CE/re-melted regions of the NiTi surface provide small ME current and that of the HAZ region provide a large current response, with a measured current in excess of the bulk

solution. In the centre of the laser spots, located at approximately 0.60 mm by 0.65 mm (x, y) topography variations on the order of 7  $\mu\text{m}$  are present in the topography map (Figure 2). The general patterns observed in the SECM images (Figure 3) are consistent with the topography images, and the current variations reflect the combined effect of reactivity and topography at the centre of the spot.



**Figure 3: SECM feedback images 1LP, 3LP, and 5LP.**

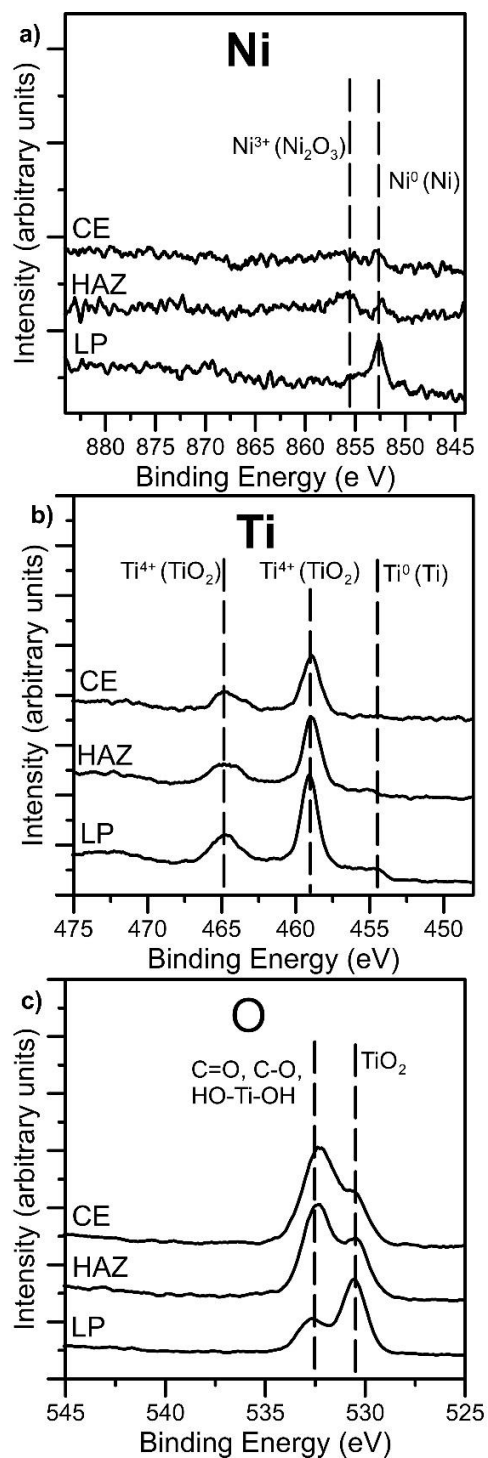
### 3.2 XPS and Raman Spectroscopy

The SECM mapping of the laser spot (Figure 3) was used to direct the XPS and Raman analysis to determine what compositional and structural differences exist between the oxides present in the HAZ, CE and LP regions. First, XPS results for each of the regions in 5LP samples were analyzed and the identified species and their respective binding energies are summarized in Table 1 to aid in further discussion.

Figure 4 shows the XPS spectra of 5LP NiTi sample in the retained CE region, inside the LP region and the HAZ region prior to corrosion testing. Analysis showed the retained CE and the HAZ regions contained trace amounts of elemental Ni. Additionally, in the HAZ region,  $\text{Ni}_2\text{O}_3$  was also identified, which is in agreement with previous studies [18]. Although a binding energy of 855.66 eV can be associated with constituents other than  $\text{Ni}_2\text{O}_3$ , laser processing is known to deposit a black plume on the surface [10], which suggests that it is in fact  $\text{Ni}_2\text{O}_3$ . Both the retained CE and HAZ regions contained titanium primarily in the form of  $\text{TiO}_2$ , with relatively higher intensity in the HAZ region. In these regions, oxygen was mostly in the form of carbon bonding or possible metallic hydroxides with some metallic oxide ( $\text{TiO}_2$ ). XPS analysis inside the LP region showed a greater amount of Ni and Ti compared to the retained CE and HAZ regions. This may be due to the presence of inclusions, such as  $\text{TiC}$  and  $\text{Ti}_2\text{NiO}_x$ , in the CE and HAZ regions, which would be the result of the material manufacturing process [15]. Similar to the retained CE and HAZ regions, the region inside the LP also contained titanium primarily in the form of  $\text{TiO}_2$ . Additionally, the presence of metallic Ti can be observed inside the LP region at a binding energy of 454 eV. In this region oxygen was mostly in the form of metallic oxide ( $\text{TiO}_2$ ), with some carbon bonding or possible metallic hydroxides.

**Table 1: Species identified via XPS analysis on the CE, LP, Edge of spot and LP-Pol surfaces and their respective binding energies (eV)**

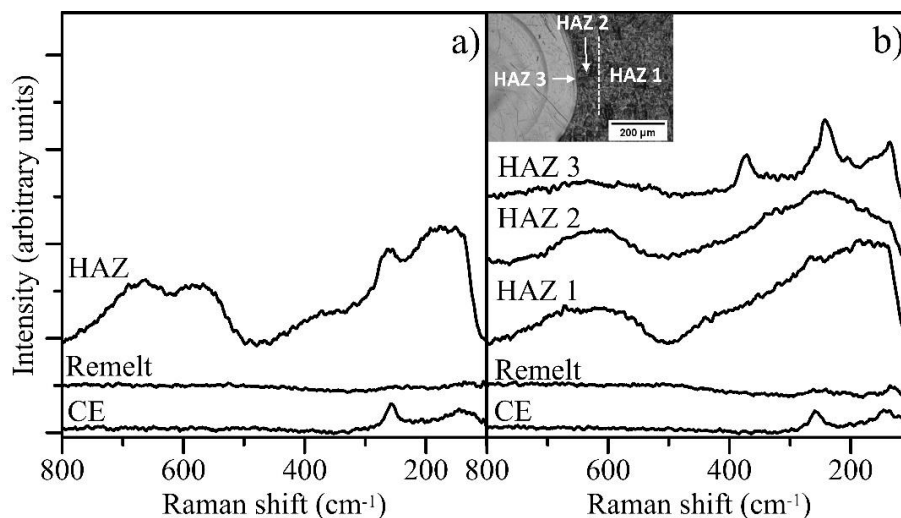
	Peak Energy (eV)	Constituent	Region		
			CE	LP	HAZ
Ni	852.68	Ni 2p <sub>2/3</sub>	no	yes	yes
	855.66	Ni <sub>2</sub> O <sub>3</sub>	no	no	yes
	859.51	Ni 2p <sub>2/3</sub>	no	no	yes
Ti	454.85	Ti 2p <sub>2/3</sub>	no	yes	no
	458.76	TiO <sub>2</sub> 2p <sub>2/3</sub>	yes	yes	yes
	460.48	Ti 2p <sub>2/3</sub>	no	no	no
	464.40	TiO <sub>2</sub> 2p <sub>2/3</sub>	yes	yes	yes
O	530.24	TiO <sub>2</sub>	yes	yes	yes
	532.40	C=O, C-O, HO-Ti-OH	yes	yes	yes



**Figure 4: XPS spectra of Ni, Ti and O from the 5LP sample in the retained CE region, inside the LP region and the HAZ region prior to corrosion testing.**

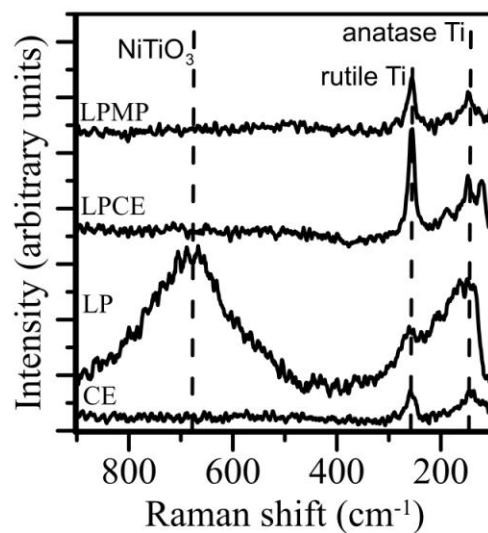
The Raman spectra for 1LP and 5LP samples are shown in Figure 5. The 1LP sample had only one distinct HAZ region where partial crystallinity was observed. Due to the broadness of the peaks, the exact species present could not be verified, but the peaks present on the HAZ of the 1LP sample correspond closely with the anatase and rutile forms of  $\text{TiO}_2$ , which correlates with previous findings [25, 26]. The Raman spectrum of outer HAZ region of the 5LP sample, labelled HAZ 1 in Figure 5b, was similar to the HAZ region observed in the 1LP sample. The inner HAZ region, labelled HAZ 3 in Figure 5b, displayed significantly sharper peaks, suggesting an increase in crystallinity and/or oxide thickness. The peaks observed in the HAZ 3 tend to correspond more closely to  $\text{NiTiO}_3$ , which has several peaks including  $\sim 250\text{ cm}^{-1}$  and  $\sim 350\text{ cm}^{-1}$  [26]. However, the absence of a peak above  $700\text{ cm}^{-1}$  suggests the presence of  $\text{Ni}_2\text{O}_3$ . The XPS analysis shown in Table 1 also confirms the presence  $\text{Ni}_2\text{O}_3$ . This is an indicator that some form Ni or Ni oxides may have been entrapped in the HAZ region after repeated pulsing. This result correlates well with the SECM maps in Figure 3, which showed increased reactivity on the right side of the laser spot corresponding to Ni plume deposits. The region labelled HAZ 2 appears to be a transition from the outer HAZ region to the inner HAZ region.





**Figure 5: Raman spectra for a) 1LP and b) 5LP. Inset shows the area from which the Raman spectra were taken, indicating the different HAZ regions observed on 5LP.**

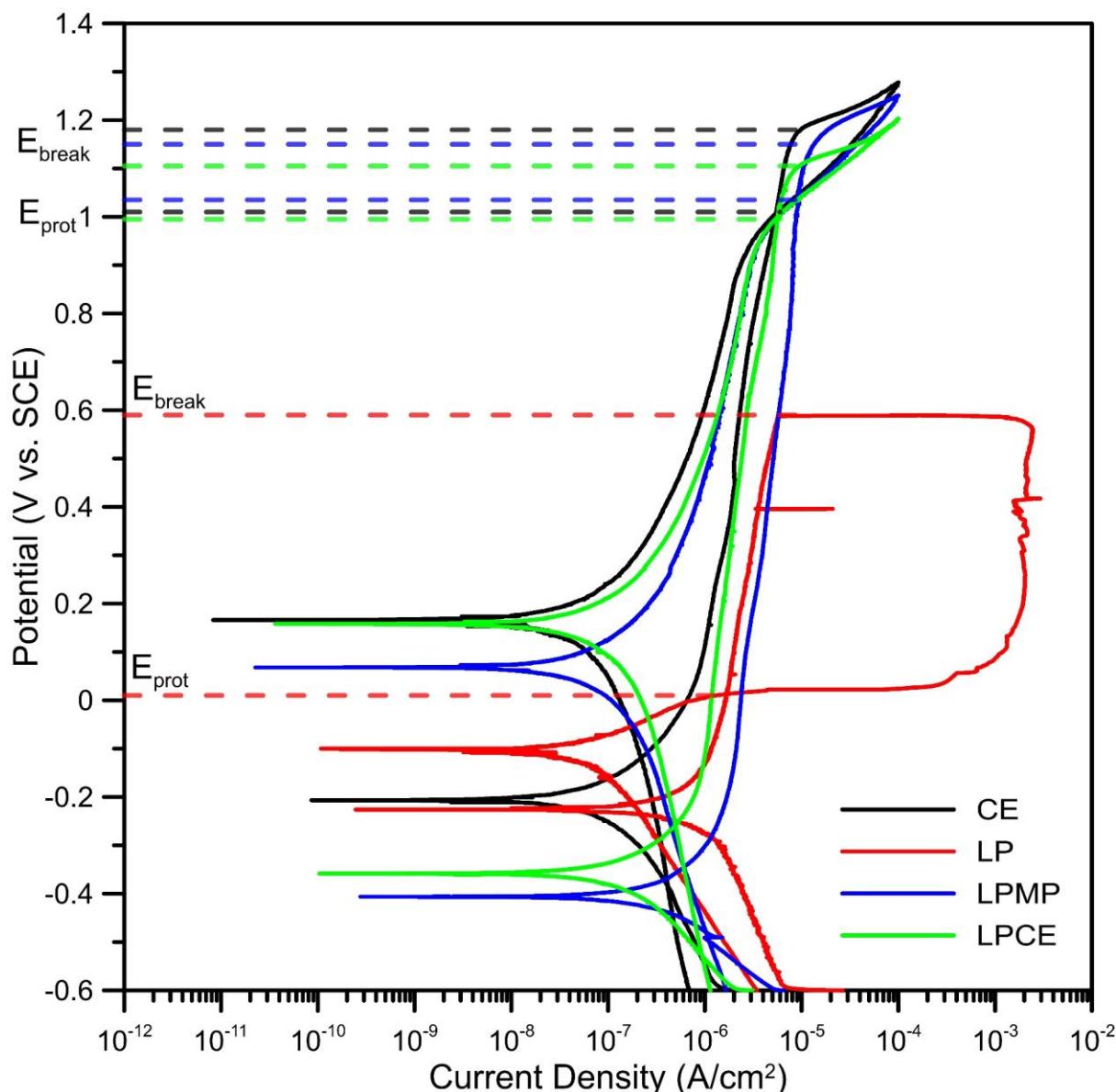
Figure 6 compares the Raman spectra for LPMP and LPCE with the baseline CE and unfavourable HAZ region from the LP samples. The HAZ region examined on the LP samples was comparable to the HAZ regions observed in the 1LP and 5LP samples. After mechanically polishing, the degree of crystallinity is reduced and comparable to the original CE surface. After chemical etching, the LPCE surface does not show any trace of Ni based crystals; however, the Ti oxide peaks shows an increase in intensity compared the original CE surface. This may be an indication that some areas on the surface are close to being over etched, as it was etched both before and after laser processing. Potentiodynamic cyclic polarization tests were performed to determine the effect of this crystallinity on the stability of the surface oxide.



**Figure 6: Representative Raman spectra for the CE surface and untreated HAZ region compared with LPMP and LPCE surfaces.**

### 3.3 Potentiodynamic Cyclic Polarization Curves

Representative cyclic polarization curves for CE, LP, LPMP and LPCE surfaces are shown in Figure 7. The current densities in the passive region ( $i_{\text{pass}}$ ) for all materials are of the same magnitude, ranging from  $1.99 \times 10^{-6} \text{ A/cm}^2$  to  $4.56 \times 10^{-6} \text{ A/cm}^2$ . However, the LP and LPMP samples tended to be slightly higher, indicating a marginally faster rate of metal dissolution.



**Figure 7: Representative potentiodynamic cyclic polarization curves comparing CE, LP, LPMP and LPCE.**

It is important to note that the breakdown ( $E_{\text{break}}$ ) and repassivation ( $E_{\text{prot}}$ ) potentials for the LP sample are drastically lower than all of the other samples. This is not an anomaly, but a manifestation of the instability of the surface oxide after high-energy-input laser processing, and is consistent with results observed by Chan *et al.* in a previous study [17]. Sudden spikes were also observed in the LP curve, which is likely due to the sudden pitting and repassivation of the

material. This was the only sample that experienced this phenomenon and it is most likely related to the instability of the surface oxide after high-energy-input laser processing. After mechanical polishing,  $E_{\text{break}}$  and  $E_{\text{prot}}$  are restored to 1,137 mV vs SCE and 1,026 mV vs SCE, respectively, similar to the 1,170 mV vs SCE breakdown potential and 1,027 mV vs SCE repassivation potential of the CE surface; however, the corrosion potential ( $E_{\text{corr}}$ ) of the LPMP surface decreases significantly to -461 mV vs SCE compared to -194 mV vs SCE of the CE surfaces. After chemically etching,  $E_{\text{break}}$  and  $E_{\text{prot}}$  improve to 1,092 mV vs SCE and 974 mV vs SCE, respectively, which is only slightly lower than the breakdown and repassivation potentials of the original CE surface. Again, the  $E_{\text{corr}}$  of the LPCE sample is reduced to -319 mV vs SCE. The lowering of the corrosion potential for both of the post-processed surfaces indicates that they have a more reactive starting surface. All pertinent values are summarized in Table 2.

**Table 2: Corrosion properties measured by potentiodynamic cyclic polarization testing.**

Surface	$E_{\text{cor}}$ (mV vs. SCE)	$E_{\text{break}}$ (mV vs. SCE)	$E_{\text{prot}}$ (mV vs. SCE)	$i_{\text{pass}}$ (A/cm <sup>2</sup> )
CE	-194 ± 32	1,170 ± 25	1,027 ± 18	2.48 ± 0.55 × 10 <sup>-6</sup>
LP	-250 ± 16	535 ± 71	-35 ± 57	4.56 ± 0.55 × 10 <sup>-6</sup>
LPMP	-460 ± 46	1,137 ± 42	1,026 ± 7	3.96 ± 0.75 × 10 <sup>-6</sup>
LPCE	-319 ± 27	1,092 ± 7	974 ± 17	1.99 ± 0.28 × 10 <sup>-6</sup>

A large hysteresis between the breakdown potential and the corrosion potential was observed for the LP sample (570 mV). This large hysteresis has been observed previously from laser welded NiTi [17] and is an indication of a significant loss of material. Both of the post-process treated surfaces had somewhat reduced hystereses on average (LPMP 111 mV and LPCE 118 mV)

compared to the original base material (147 mV). This is a good indication that the treatments were effective at restoring the corrosion performance of the laser processed NiTi.

## 4.0 Discussion

### 4.1 Effects of laser processing on surface structure

The corrosion performance of NiTi SMAs is linked to the surface composition and chemistry, surface roughness and homogeneity [8, 27, 28]. Laser processing has been found to have a detrimental effect on these properties in some cases [16-18]. It is therefore necessary to discuss the evolution of the surface oxide during laser processing and how post-process surface treatments alter the surface oxide so that corrosion performance of laser processed NiTi may be optimised.

Localized surface reactivity gives great insight into how the surface oxide changes during laser processing. It was observed that after only one laser pulse a ring of increased reactivity existed adjacent to the re-melted region (*vide infra*), see Figure 3. This corresponds with the heat affected zone (HAZ) found around the weld pool region and was consistent with the studies by Chan et al. [16] and Pequegnat et al. [18] who observed preferential corrosion adjacent to pulsed laser spots. Increasing the number of pulses not only increases the intensity of ME current, it also increases the area in which higher current was observed, as shown in Figure 8. The high current areas were estimated to be 0.27 mm<sup>2</sup> for 1LP, 0.30 mm<sup>2</sup> for 3LP and 0.40 mm<sup>2</sup> for 5LP.

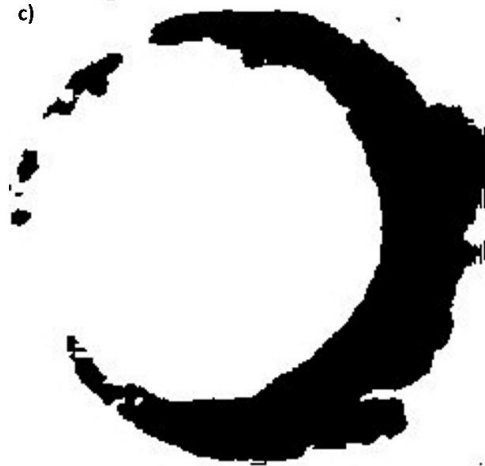
a)



b)

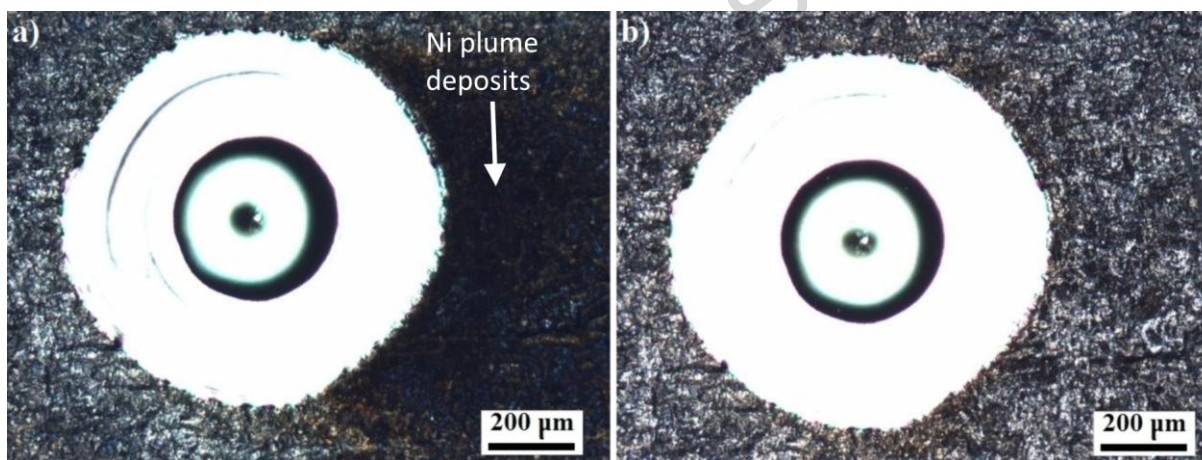


c)



**Figure 8: Surface areas of reactive regions estimated from SECM mappings for a) 1LP, b) 3LP, and c) 5LP. Black regions are conductive areas, white regions are insulating areas.**

However, the growth of the annular region of higher ME current was not uniform (Figure 3 and Figure 8) suggesting that it may not result entirely from the formation of the HAZ region. The trailing edge to the right of the re-melted region, see Figure 3, corresponds to the plume deposits that form on the surface during laser processing, as is clearly observed in the optical image of the laser spot prior to cleaning, as shown in Figure 9a. This agrees with previous studies that observed Ni-enriched plume deposits after laser pulsing of NiTi SMAs [10]. Note the removal of the black plume deposit to the right of the laser spot after cleaning (Figure 9b) prior to SECM analysis.



**Figure 9: Optical images of a laser processed sample a) with post-process plume deposits on surface and b) after cleaning thoroughly with acetone, ethanol, and de-ionized water.**

To investigate the effects of the HAZ and nickel plume deposits better XPS analysis was performed to determine which species were present and Raman spectroscopy was performed to determine the degree of crystallinity. The XPS and Raman results confirmed the presence of Ni based oxides in the HAZ region after laser processing. Additionally, the spread of the crystalline structure and the extent to which it crystallizes seems to increase with the number of pulses. This makes sense, as more pulses input more energy into the sample, the oxide crystals are able to

grow bigger and further. Thermodynamics of solid state diffusion tells us that when sufficient energy is applied preferred crystal orientations will grow and absorb smaller crystals in order to reduce the overall energy of the crystalline structure. This is the concept applied when metals are annealed. It is well known that welding methods, including pulsed laser methods, provide enough energy for crystal growth to occur in the heat affect zone (that is the definition of a heat affected zone). If one laser pulse provides enough energy for crystals to grow in the HAZ, then multiple pulses will provide enough energy for crystals to grow multiple times.

The presence of these semi-crystalline regions surrounding the pulse is significant in two ways. Firstly, these crystal structures are known to provide a pathway for oxygen to reach the bulk NiTi material below the protective oxide and for Ni to reach the outer surface [25]. This agrees with the SECM findings that showed increased conductivity in the HAZ region. It is well known that an amorphous oxide layer is more robust than a crystalline oxide layer [28, 29], since the amorphous oxide lacks these quick diffusion pathways. Secondly, these semi-crystalline regions create non-homogeneity across the surface. It is also well known that a uniform, homogeneous surface oxide is more stable than a heterogeneous surface oxide layer [28].

#### **4.2 Effects of post-process treatment on restoring corrosion properties**

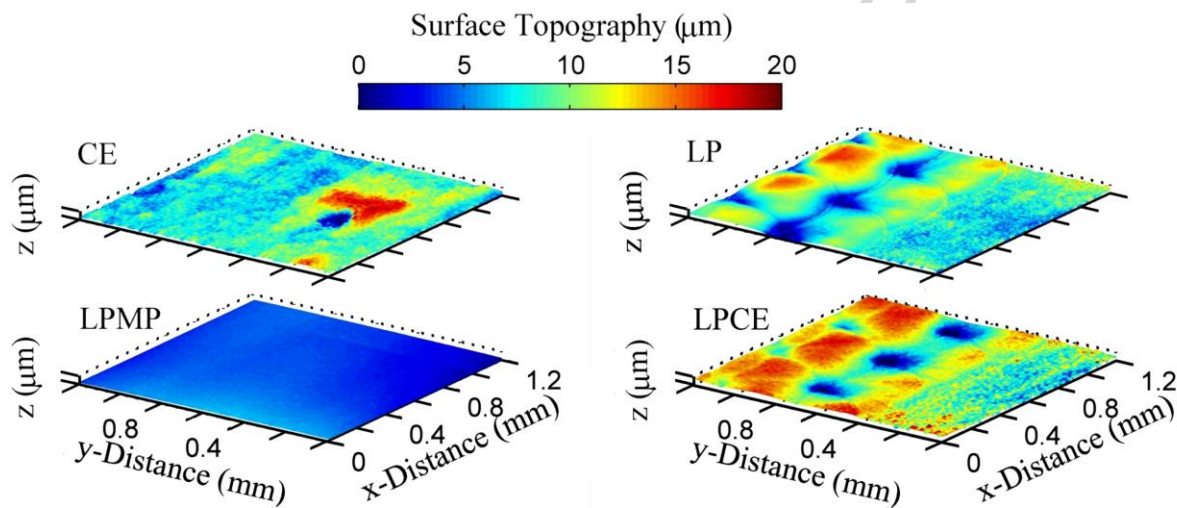
Even though laser processing can increase the crystallinity of the surface oxide in the HAZ region, Raman spectroscopy showed that the post-process treatments successfully removed this detrimental crystallinity. The method of mechanically polishing the surface physically removed the old surface oxide layer and rebuilt a new one. The method of chemically etching the surface preferentially removed the Ni ions and oxides present and built up the Ti oxide [29, 30].



There was a significant decrease in the corrosion potential, from  $-194 \pm 32$  mV vs SCE to  $-460 \pm 46$  mV vs SCE, after mechanical polishing for two primary reasons: (i) an increase in elemental Ni composition and (ii) the presence of elemental Ti. The mechanically polished surface contains more elemental Ni compared to the CE surface because this method does not preferentially remove Ni, whereas chemically etching does preferentially remove Ni from the surface. The Ti is only partially oxidized at room temperature so there is still some elemental Ti in the surface oxide of the LPMP sample [28]. This slight increase of Ni on the surface may also account the increase in the passive current densities of the LPMP ( $i_{\text{pass}} 3.96 \times 10^{-6}$  A/cm<sup>2</sup>) and the LP ( $i_{\text{pass}} 4.56 \times 10^{-6}$  A/cm<sup>2</sup>) samples compared to the CE ( $i_{\text{pass}} 2.48 \times 10^{-6}$  A/cm<sup>2</sup>) and the LPCE ( $i_{\text{pass}} 1.99 \times 10^{-6}$  A/cm<sup>2</sup>) samples. It has been shown that there is slow selective dissolution of Ni in the passive region [31].

There was also a marginal difference observed in the breakdown and repassivation potentials between the original chemically etched surface ( $E_{\text{break}} 1,170 \pm 25$  mV vs SCE,  $E_{\text{prot}} 1,027 \pm 18$  mV vs SCE) and the post-processed chemically etched surface ( $E_{\text{break}} 1,092 \pm 7$  mV vs SCE,  $E_{\text{prot}} 974 \pm 17$  mV vs SCE). This may be because the post-process etching time was significantly shorter in duration than the original etching treatment. This shorter etching time was used to avoid over etching the retained original CE surfaces, which could lead to hydrogen embrittlement [32]. A slight increase in the oxide peak observed by Raman spectroscopy (see Figure 6) reveals that is was a valid concern, because increasing the crystallinity of Ti can lead to brittleness [33]. The shorter etching time may also have resulted in the decrease of the corrosion potential from  $-194 \pm 32$  mV vs SCE to  $-319 \pm 27$  mV vs SCE. In the regions where the detrimental Ni oxides were removed, a new Ti oxide was created. With the shorter etching time, the oxidation may have been incomplete in these areas, causing the surface to be more reactive than the original CE surface.

Additionally, an increase in the surface roughness of the LPCE surface compared to the CE surface increases the number of initiation sites for corrosion to occur. Even after the post-process chemical etch, this increased roughness could contribute to the observed disparity in breakdown potentials of the original CE surface and the post-processed CE surface. The 3D topographical profiles for post-process surface treatments are shown in Figure 10.



**Figure 10: Surface topography maps of CE, LP, LPMP, and LPCE samples.**

Both the LPMP and CE surfaces had comparable breakdown and repassivation potentials. A previous study by Pequegnat et al. [18] showed that the oxide thickness of the mechanically polished surface is about half the thickness of the chemically etched surface. However, the oxide layer of a MP sample continues to develop *in situ* [27] preventing earlier failure due to a thinner oxide. It is important to note that the mechanically polished surface does not undergo any preferential corrosion due to the differences in the bulk compositions. This result aligns with the work of Ruhlig et al. [34] who found no significant difference in corrosion performance across the surface of a polished cross-section of a NiTi weld.

Both the mechanically polishing and chemically etching post-process surface treatments were able to restore the corrosion performance of the NiTi SMA to pre-processed capabilities. This shows that any detrimental effects due to laser processing of NiTi SMA are superficial and can be easily removed; however, the optimization of the post-process treatment methods is still required in order to achieve even better corrosion performance.

## 5.0 Conclusions

In this study, the evolution of the surface reactivity, crystallinity, and overall oxide stability were characterized after laser processing. It was also demonstrated how post-process surface treatments can remove the surface oxide crystallinity that forms during laser processing of NiTi SMA. The most significant findings of this study included:

1. An increase in surface reactivity was observed in the HAZ region by SECM. It was determined to be a result of increased crystallinity with possible contributions from Ni plume deposits following the heating/cooling cycles that occur during laser processing.
2. The degree of oxide crystallinity increased when the amount of laser pulses increased. The additional heat input from multiple pulses allowed for more crystalline oxide growth.
3. The post-process surface treatments employed were effective at restoring the corrosion performance of laser-process NiTi SMA to pre-processed capabilities. Raman spectroscopy verified the removal of surface oxide crystallinity by both mechanical polishing and chemical etching of the surface of LP samples.

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## Highlights

- Surface mapping identified heat affected zone (HAZ) as location of highest reactivity after laser processing
- Evolution of oxide crystallinity in HAZ with increased laser energy was tracked
- Post-process acid etching and mechanical polishing were successfully employed to restore corrosion performance
- Reduction in corrosion performance was determined to be on the surface only, not within the bulk of the HAZ