Effect of CO₂ Laser Welding on the Shape-Memory and Corrosion Characteristics of TiNi Alloys

Y.T. HSU, Y.R. WANG, S.K. WU, and C. CHEN

A CO₂ laser has been employed to join binary Ti₅₀Ni₅₀ and Ti_{49.5}Ni_{50.5} shape-memory alloys (SMAs), with an emphasis on the shape-memory and corrosion characteristics. Experimental results showed that a slightly lowered martensite start (M_s) temperature and no deterioration in shape-memory character of both alloys were found after laser welding. The welded Ti₅₀Ni₅₀, with an increased amount of B2 phase in the weld metal (WM), had higher strength and considerably lower elongation than the base metal (BM). Potentiodynamic tests revealed the satisfactory performance of laser-welded Ti₅₀Ni₅₀ in 1.5 M H₂SO₄ and 1.5 M HNO₃ solutions. However, the WM exhibited a significantly higher corrosion rate and a less stable passivity than the BM in artificial saliva. On the other hand, the pseudoelastic behavior of the laser weld was investigated only for the Ti_{49.5}Ni_{50.5} alloy, to facilitate tension cycling at room temperature. The cyclic deformation of Ti_{49.5}Ni_{50.5} indicated that the stress required to form stress-induced martensite (σ_m) and the permanent residual strain (ϵ_p) were higher after welding at a given number of cycles (N), which were certainly related to the more inhomogeneous nature of the WM.

I. INTRODUCTION

SHAPE memory alloys (SMAs) are recognized as functional materials and have many industrial applications based on their shape-memory effect (SME) and superelastic or pseudoelastic (PE) properties.^[1,2,3] The TiNi alloy system, which exhibits thermoelastic martensitic transformation, is known as one of the most important SMAs. For engineering applications, the weldability might have great influence on the usage of a material and should be taken into account in the evaluation process. In case of welding SMAs, an additional requirement is to maintain the shape-memory characteristics of the weld. Many research projects have been conducted on joining, in order to develop appropriate welding processes for TiNi alloys. For instance, Nishikawa *et al.* (1982) utilized resistance butt welding to join a TiNi alloy and indicated that the welded specimen can reach 80 pct of the tensile strength of the base metal (BM) and exhibit a SME.^[4] Beyer *et al.* (1986) studied the microstructures of a resistance butt-welded Ti-50.3 at. pct Ni alloy and found that Ti₂Ni precipitates can affect the performance of the SME and weaken the strength of grain boundaries in the weld metal (WM).^[5] Hirose *et al.* (1990) joined TiNi alloys with a CO₂ laser and observed the existence of cellular dendrites in the WM and Ti₂NiO_x oxides in between the dendritic arms.^[6] Furthermore, good PE properties were also obtained after cyclic tensile tests with a 4 pct strain amplitude, in which a residual strain of about 0.2 pct was reported after 50 cycles. Schloßmacher *et al.* (1994) joined Ti-49.3 at. pct Ni with a Nd-YAG (neodymium-yttrium aluminum garnet) laser and revealed that the differential scanning calorimetry (DSC) and tensile curves of the welded specimens are almost the

same as those in the BM.^[7] They also pointed out that the decreased fracture strain of the tensile curve in the welded specimen was due to the Ti₂Ni precipitated at the grain boundaries. Recently, the same authors found that the PE behavior of a Ti-51.5 at. pct Ni alloy can be maintained after laser welding.^[8]

From the aforementioned literature survey, it can be seen that laser welding is a promising method for joining TiNi alloys. However, in most of the tensile tests on laser-welded TiNi alloys, conventional tensile specimens were used with a narrow WM in the central part of the specimen.^[6] Under this situation, the tensile axis of the welded specimens was normal to the welding direction. Obviously, the WM only occupied a small fraction (<10 pct) of the gage length, and, therefore, the stress-strain curves of such specimens were difficult to compare with the unwelded ones. To overcome this shortcoming, tensile specimens with a high volume fraction (~66 pct) of the WM within the gage length were prepared for this study. This could be easily achieved by making the welding direction parallel to the loading axis of tensile specimens. Apart from that, the corrosion resistance of the TiNi alloy in various environments is of great importance for practical applications, yet little is known about this subject,^[9–12] especially for the welded structures. The purpose of this study was to investigate not only the SME but also the corrosion characteristics of TiNi SMAs after welding. In order to evaluate the corrosion behavior of laser-welded Ti₅₀Ni₅₀ specimens, potentiodynamic tests in acid and chloride solutions were performed. These results were also compared with those of unwelded specimens for better judgement regarding the use of welded Ti₅₀Ni₅₀ in a specific application.

II. EXPERIMENTAL PROCEDURE

The TiNi SMAs with chemical compositions (at. pct) of Ti₅₀Ni₅₀ and Ti_{49.5}Ni_{50.5} were used in the experiment. The latter alloy was only used to study the PE behavior of laser-welded specimens, due to its austenite finish temperature

Y.T. HSU and Y.R. WANG, formerly Graduate Students, Institute of Materials Science and Engineering, National Taiwan University, are now Process Engineers with the Taiwan Semiconductor Manufacturing Company, Hsinchu, Taiwan 300, Republic of China. S.K. WU and C. CHEN, Professors, are with the Institute of Materials Science and Engineering, National Taiwan University, Taipei, Taiwan 106, Republic of China.

Manuscript submitted November 30, 1998.

Table I. Laser Welding Parameters Used in the Experiment

Laser power	2200 to 2500 W
Travel speed	1500 mm/min
Focal lens	ZnSe
Focal length	190 mm
Focal point	0.5 mm underfocus (0.5 mm below the surface)
Plasma assist gas flow rate	25 ℓpm He
Backing gas flow rate	20 ℓpm Ar

(A_f) being slightly lower than room temperature. High-purity elements of Ti and Ni were melted and cast in a vacuum arc remelting furnace to obtain ingots of about 100 g. The weight losses were measured to be less than 0.01 pct after melting. Ingots were then hot rolled with intermediate annealing at 850 °C into strips of roughly 2 mm in thickness. All strips were subjected to a final annealing of 800 °C/1 h to eliminate residual stresses from rolling.

For the laser welding experiments, a commercially available laser (Rofin-Sinar 850) was employed. This laser generates a continuous-wave CO₂ laser beam at a 10.6 μm wavelength, with an output power of up to 5 kW. Bead-on-plate welds were made without filler-metal additions, by proper selection of processing parameters, to obtain full-penetration welds. Table I lists the laser welding variables utilized in the investigation. The average bead width on the top and bottom surfaces was 2 mm. After welding, no postweld heat treatments were made on these welds. The welded specimens were then characterized by means of metallographic examinations, microhardness measurements, and X-ray diffraction (XRD) with Cu K_α radiation.

The changes in martensitic transformation temperatures of Ti₅₀Ni₅₀ and Ti_{49.5}Ni_{50.5} after welding were measured using DSC. Samples of the BM and the WM were cut from laser welds to obtain a mass of 15 mg for the DSC measurement, which was carried out by using a DUPONT* 2000 thermal

*DUPONT is a trademark of E.I. DuPont de Nemours, Wilmington, DE.

analyzer with a controlled cooling/heating rate of 10 °C/min. Heats of transformation (ΔH) were automatically calculated from the areas under DSC peaks using the software package provided by DUPONT. The shape memory (shape recovery) was examined by a bending test with 1 × 1 × 40 mm³ specimens. The specimens were deformed to an angle of θ_i in liquid nitrogen and then heated to various temperatures up to 125 °C. The initially deformed angle θ_i was found to recover to θ_f , which was temperature dependent. The shape recovery is defined as $(\theta_i - \theta_f)/\theta_i$, and the detailed measurement of the shape-recovery characteristics has been presented elsewhere.^[13]

The specimens for tensile testing were conducted on a MTS machine at room temperature, with a strain rate of 1 × 10⁻⁴ s⁻¹. The loading axis was aligned parallel to the welding direction in the tensile specimens, and the strain was measured by an extensometer with a gage length of 12.5 mm. Figure 1 shows the dimension of the tensile specimens, in which the WM consists of nearly two-thirds the volume fraction of the material within the gage length. This configuration is similar to a lamellar composite, with the WM being sandwiched by two strips of heat-affected

Fig. 1—Schematic diagram showing the dimension of laser-welded tensile specimens.

zones (HAZs). Hence, the tensile results of laser welds represent the mechanical behavior of both the WM and the HAZ, which are the two distinct regions associated with welding. In addition, cyclic loading-unloading experiments, with a maximum strain of 3 pct and up to 50 cycles, were performed on a Ti_{49.5}Ni_{50.5} SMA at 25 °C. The selection of such an alloy facilitated the comparison of PE properties between unwelded and welded specimens at room temperature.

The potentiodynamic behavior of Ti₅₀Ni₅₀ specimens with and without welding was studied in nitrogen-purged 1.5 M H₂SO₄ (25 °C), 1.5 M HNO₃ (25 °C), and artificial saliva (37 °C) solutions. The composition of artificial saliva (each gram) contained 0.844 mg NaCl, 1.2 mg KCl, 0.146 mg CaCl₂, 0.052 mg MgCl₂ · 6H₂O, 0.34 mg K₂HPO₄, 60 mg C₆H₁₄O₆, 2 mg C₈H₈O₃, and 3.5 mg ROCH₂CH₂OH. The electrochemical cell was a conventional three-electrode cell consisting of a working electrode, a saturated calomel reference electrode (SCE), and a platinum counter electrode. The specimens, which were connected to the working electrode, were polished with up to 1000-grit SiC abrasive paper. The exposed surface area of specimens to the solution was 10 × 10 mm² for the BM and 10 × 2 mm² for the WM. The potentiodynamic curves were determined with a Princeton Applied Research 273 potentiostat and associated computer software for corrosion measurements. A scan rate of 300 mV/min was used in the range of -250 ~ +1900 mV (SCE) for the H₂SO₄ and HNO₃ solutions and in the range of -1200 ~ +1900 mV (SCE) for the artificial saliva.

III. RESULTS AND DISCUSSION

A. Metallography and Hardness

Figure 2(a) is a typical surface macrostructure of the laser-welded specimen, in which columnar grains growing toward the weld centerline are obvious. Since the liquid/solid boundary of the trailing portion of a tear-drop weld pool is straight, the grains are aligned in a direction close to the normal of the weld centerline for welding at high speeds. Figure 2(b) shows the optical photograph near the fusion boundary of the weld, with evidence of epitaxial growth, *i.e.*, grain growth initiating from the HAZ at the fusion boundary and proceeding toward the weld centerline. It should be pointed out that no grain growth in the HAZ was found, owing to the low heat input (≤ 100 J/mm) of the process and the use of a fully annealed (strain-free) material prior to laser welding. The microhardness profile across the fusion zone of the laser weld is displayed in Figure 3. Although a difference in grain

Fig. 2—Metallographs showing (a) the surface macrostructure of laser-welded specimens and (b) the microstructure near the fusion boundary.

Fig. 3—Microhardness profile across the fusion zone of a laser weld.

structures between the fusion zone and the HAZ (or the BM) existed, no apparent changes in hardness were disclosed. The hardness (H_v) of the fusion zone (or the WM) was approximately 235 Hv, resembling that in the BM or the HAZ of laser-welded specimens. The results of metallographic examinations and hardness measurements suggested that the HAZ could be considered the same as the BM in the present study.

B. The DSC and Shape-Recovery Tests

Figure 4 shows the DSC results of a $\text{Ti}_{50}\text{Ni}_{50}$ SMA with and without laser welding. The BM exhibits a one-step

Fig. 4—DSC results of (a) the base metal and (b) the weld metal of $\text{Ti}_{50}\text{Ni}_{50}$ SMA.

Table II. Shape Memory Recovery of $\text{Ti}_{50}\text{Ni}_{50}$ Alloy with and without Laser Welding

Temperature	Shape Recovery (Pct)	
	The Base Metal	The Weld Metal
40 °C	0 pct	0 pct
75 °C	37 pct	40 pct
95 °C	95 pct	93 pct
125 °C	97 pct	98 pct

$\text{B2} \leftrightarrow \text{B19'}$ martensitic transformation with a martensite start (M_s) temperature of 38 °C (Figure 4(a)), while the WM shows an M_s temperature of 32 °C (Figure 4(b)). By comparison of Figures 4(a) and (b), it is found that transformation peaks and M_s temperature are all smaller in the WM. This might come from the fact that the interstitial atoms, in particular, oxygen, were introduced into the WM during welding. The content of oxygen in the BM and the WM was found to be in the neighborhood of 300 and 500 ppm, respectively. For the shape-recovery test, the results of $\text{Ti}_{50}\text{Ni}_{50}$ specimens with and without laser welding are given in Table II. Obviously, the WM of laser-welded specimens showed no deterioration in shape recovery. This feature, together with a minor change in M_s temperature, confirmed that the laser welding is a suitable process for joining TiNi SMAs.

(a)

(b)

Fig. 5—XRD results of (a) the base metal and (b) the weld metal of $\text{Ti}_{50}\text{Ni}_{50}$ specimens.

C. The XRD Analysis

Figure 5 presents the XRD results of $\text{Ti}_{50}\text{Ni}_{50}$ specimens before and after laser welding. It also reveals the existence of B2 and B19' peaks in these specimens at room temperature. However, the relative amount of B2 in the BM (Figure 5(a)) is less than that in the WM (Figure 5(b)). This coincides with the DSC results, as indicated in Figure 4, in which the BM has more B2 that has been transformed to B19' than does the WM at room temperature. It is important to note that an abnormally high intensity of B2_{200} is observed in Figure 5(b), which is different from JCDPS diffraction data for nickel titanium, which has the highest intensity of B2_{110} .^[14] This effect is due to the development of a solidification texture associated with the welding process. The surface of the specimen analyzed by X-rays (Figure 5(b)) was perpendicular to the growth direction of the crystals during solidification. It is known that the $\langle 100 \rangle$ directions are the preferred growth directions for cubic metals.^[15] The $\text{Ti}_{50}\text{Ni}_{50}$ SMA is a B2 (ordered bcc) structure, and, thus, it is quite reasonable to have $\langle 100 \rangle_{\text{B2}}$ preferred growth orientations, which cause the abnormally high intensity of B2_{200} in Figure 5(b).

Fig. 6—Typical tensile stress-strain curves at room temperature of $\text{Ti}_{50}\text{Ni}_{50}$ SMA with and without laser welding.

D. Tensile Test

Figure 6 shows the engineering stress-strain curves at room temperature for unwelded and laser-welded specimens. As shown in Figure 4, the tensile specimens are comprised of B2 and B19' mixed phases at room temperature. Nevertheless, the welded specimen had more B2 phase than the unwelded specimen at room temperature. It has been reported that the B2 phase has a higher strength than the B19' phase.^[16] Therefore, a higher tensile strength of the laser weld than that of the BM is expected in the stress-strain curves. Figure 6 also reveals that the fracture strain of the BM (15 pct) is higher than the recoverable strain (7 to 8 pct) of the shape-memory recovery in a $\text{Ti}_{50}\text{Ni}_{50}$ SMA, but the fracture strain of the laser weld is considerably lowered (7 pct). The deterioration of the fracture strain in a laser-welded $\text{Ti}_{50}\text{Ni}_{50}$ SMA could be attributed to (1) the segregation of solute and impurity elements during solidification; (2) the coarse-grained and dendritic structures in the WM; and (3) the increase in B2 phase of specimens after welding. Figure 7 shows tensile-fractured fractographs of the unwelded and laser-welded specimens. Distinct fracture modes are observed in these specimens, *i.e.*, a predominant dimple fracture (Figure 7(a)) for the unwelded specimen and a brittle appearance (Figure 7(b)) for the welded specimen in the WM region. It is noted that the HAZ (which is basically the same as the BM in this study, as discussed in Section III-A) in the welded specimen has a similar fracture appearance, as shown in Figure 7(a). In addition, the weld centerline, which is a narrow region with the most severe segregation in the welding, still can be seen in Figure 7(b). Although the laser-welded specimen consisted of only 66 pct WM in the gage section, the tensile properties of the entire WM specimen could be estimated. As shown in Figure 1, the WM is sandwiched by two strips of BM in the gage section of tensile specimens, resembling a lamellar composite under isostrain conditions of loading. The tensile strength of the composite (σ_{com}) can be calculated by the rule of mixtures:

Fig. 7—Tensile-fractured appearance of (a) unwelded and (b) laser-welded $\text{Ti}_{50}\text{Ni}_{50}$ specimens.

$$\sigma_{\text{com}} = 0.66 \sigma_{\text{WM}} + 0.34 \sigma_{\text{BM}}$$

where σ_{WM} and σ_{BM} are the tensile strengths of the WM and BM, respectively. Taking $\sigma_{\text{BM}} = 650$ MPa (point A) and $\sigma_{\text{com}} = 740$ MPa (point B) from the stress-strain curves in Figure 6, a tensile strength of the WM (σ_{WM}) of 786 MPa could be obtained accordingly. However, a tensile elongation of the WM in the neighborhood of 7 pct, similar to that of the composite, would be expected. This was supported by the fractographic observation of laser-welded specimens, in which the WM exhibited a brittle fracture mode while the BM revealed a ductile fracture mode. Apparently, the fracture strain of the composite (or the laser weld) was limited by the tensile ductility of the WM. Since the BM had a better ductility, the fracture strain of the WM was reached prior to that of the BM in the tensile test of laser-welded specimens.

E. Cyclic Loading-Unloading Test

The DSC results of a $\text{Ti}_{49.5}\text{Ni}_{50.5}$ SMA indicated that the M_s temperatures were approximately -20°C and -25°C for the BM and the WM, respectively. The shape-recovery test at 25°C also disclosed no deterioration of this alloy after welding (96 pct of the WM vs 98 pct of the BM). Therefore, the cyclic deformation of the alloy at room temperature with 3 pct strain amplitude could exhibit the PE behavior. Figure 8 displays such a test of $\text{Ti}_{49.5}\text{Ni}_{50.5}$ specimens before and after welding for N (number of cycles) = 1, 5, 10, and 50, in which the permanent (or irreversible) residual strain (ϵ_p) is observable at higher cycles. Several

Fig. 8—Stress-strain curves of the cyclic deformation with $N = 1, 5, 10$, and 50 for (a) unwelded and (b) laser-welded $\text{Ti}_{49.5}\text{Ni}_{50.5}$ specimens at room temperature.

Fig. 9—Schematic illustration of σ_m , E_1 , and E_2 in a typical loading-unloading curve.

PE parameters of the stress-strain curve, such as σ_m , E_1 , and E_2 , are defined and illustrated in Figure 9 to facilitate comparison. The term σ_m is used to describe the stress

required for stress-induced martensite (SIM) and is determined by the intersection of extrapolation lines of the elastic and PE portions of the stress-strain curve on loading. The term E_1 represents the dissipated energy per unit volume in one complete cycle, while E_2 is the stored energy per unit volume on loading and available for release upon unloading. As a result, the efficiency of energy storage (η) can be expressed as $E_2/(E_2 + E_1)$, according to a previous study.^[17] Figure 10(a) is the plot of σ_m vs N , revealing that the σ_m value of the WM is higher than that of the BM. This is in agreement with the predication of the Clausius–Clapeyron equation,^[18] *i.e.*, the more the temperature deviates from M_S , the more σ_m is needed for SIM. Figure 10(b) is the plot of ε_p vs N , in which ε_p increases rapidly at the beginning as N increases and then remains nearly constant after 20 cycles. The magnitude of ε_p was approximately 0.09 pct for the BM and 0.19 pct for the WM after 50 cycles. The increased ε_p of the WM might come from the fact that the more inhomogeneous characteristics, in terms of grain structures, solidification substructures, segregation, *etc.*, inherently adhere to the WM. The higher ε_p (the less recoverable strain) would also induce more work hardening, which is supported by the evidence that the slope of the PE portion becomes steeper as N increases (Figure 8). Figure 10(c) shows that the efficiency of energy storage increases with increasing N for both the BM and the WM at low cycles and reaches a steady state at high cycles. This tendency is similar to the ε_p vs N curves, as shown in Figure 10(b). It has been demonstrated that the cold-rolled (work-hardened) TiNi SMA can improve its η value.^[17] Therefore, the improved η value at low cycles could be related to the effect of cyclic (work) hardening, which became saturated as N further increased. Clearly, the smaller the E_1 and the larger the E_2 values of the WM, as compared to those of the BM at all cycles (Figure 8), are responsible for the improved η value of the WM. However, the exact mechanism that causes the η value of the WM to be higher than that of the BM after cycling still needs to be uncovered.

F. Potentiodynamic Behavior

Potentiodynamic curves for the BM and the WM of a Ti₅₀Ni₅₀ SMA exposed to 1.5 M H₂SO₄ and HNO₃ solutions at 25 °C are presented in Figures 11(a) and (b), respectively. All curves in the figures display no active-to-passive transition peaks, indicating that the surface layers of both welded and unwelded specimens were passivated spontaneously in these solutions. The BM and the WM had nearly the same passive region in either H₂SO₄ or HNO₃; however, the passive current density of the Ti₅₀Ni₅₀ SMA tended to increase after welding. For instance, the passive current density increased from 25 $\mu\text{A}/\text{cm}^2$ in the BM to 42 $\mu\text{A}/\text{cm}^2$ in the WM in H₂SO₄ and from 19 $\mu\text{A}/\text{cm}^2$ in the BM to 33 $\mu\text{A}/\text{cm}^2$ in the WM in HNO₃. Thus, the polarization behavior of the WM in these solutions was characterized by a small increase (less than 20 $\mu\text{A}/\text{cm}^2$) in current density relative to the BM, particularly at potentials of the stable passive range. It is apparent that the inhomogeneous nature of the WM might cause a less stable passive layer, as can be concluded from a slightly higher passive current density in the solution. The potentiodynamic curves in the 1.5 M H₂SO₄ and HNO₃ solutions revealed that the welded specimens did not have significant differences in electrochemical behavior

(a)

(b)

(c)

Fig. 10—Unwelded and laser-welded Ti_{49.5}Ni_{50.5} specimens after 3 pct cyclic deformation at 25 °C for (a) σ_m , (b) ε_p , and (c) η vs N plots.

Fig. 11—Potentiodynamic curves of $\text{Ti}_{50}\text{Ni}_{50}$ specimens with and without laser welding in the solutions of (a) 1.5 M H_2SO_4 at 25 °C, (b) 1.5 M HNO_3 at 25 °C, and (c) artificial saliva at 37 °C.

as compared to the unwelded $\text{Ti}_{50}\text{Ni}_{50}$ SMA. However, a distinct polarization behavior between the WM and the BM in artificial saliva at 37 °C is observed, as shown in Figure 11(c), in which the corrosion potential (E_{corr}) of the WM was shifted toward a more noble potential of -58 mV, in contrast to that of -495 mV of the BM. The passivity of the TiNi SMA has been reported to be unstable in chloride-containing media and is also affected by the presence of an intermetallic phase, which provides initiation sites for the breakdown of passive films.^[19] Apparently, the WM has a stronger tendency to form such intermetallic particles in the

solidified structure. As a result, welding not only reduced the passive range but also enhanced the anodic dissolution of the $\text{Ti}_{50}\text{Ni}_{50}$ SMA by increasing the passive current density from $22.3 \mu\text{A}/\text{cm}^2$ in the BM to $169 \mu\text{A}/\text{cm}^2$ in the WM in artificial saliva.

The corrosion rate (micrometers/year), determined by Tafel extrapolation methods to obtain the corrosion current density (I_{corr}) from potentiodynamics curves, is given in Table III. A lower I_{corr} value is thought to indicate a better resistance to general corrosion. The corrosion rate is related to I_{corr} by the equation (Faraday's law):

Table III. Corrosion Rate of Unwelded and Laser-Welded Ti₅₀Ni₅₀ Specimens in Various Solutions

Solution	Specimen	I_{corr} ($\mu\text{A}/\text{cm}^2$)	Corrosion Rate ($\mu\text{m}/\text{y}$)*
1.5 M H ₂ SO ₄ at 25 °C	BM	0.6	8.6
	WM	1.0	14.7
1.5 M HNO ₃ at 25 °C	BM	0.6	8.6
	WM	0.6	8.6
Artificial saliva at 37 °C	BM	1.3	17.0
	WM	18.3	274.4

*1 $\mu\text{m}/\text{year} \cong 4$ mpy (milli-inches per year).

Corrosion rate ($\mu\text{m}/\text{year}$)

$$= \frac{3.27(I_{\text{corr}}, \mu\text{A}/\text{cm}^2) \cdot (\text{Equivalent weight, g})}{(\text{Density, g}/\text{cm}^3) \cdot (\text{Surface area, cm}^2)}$$

The calculated equivalent weight and the measured density of the Ti₅₀Ni₅₀ SMA are 26.65 g and 6.47 g/cm³, respectively. The surface area of specimens exposed to the solution was 1 cm² for the BM and 0.2 cm² for the WM, as mentioned in the experiment. As can be seen in Table III, the corrosion rates of the unwelded specimens in H₂SO₄, HNO₃, and artificial saliva solutions are considered to be quite low, in the range of 8 to 17 $\mu\text{m}/\text{year}$. In case of the welded specimens, the corrosion rates were higher than those of the unwelded specimens in these solutions. The increase in corrosion rates of the alloy after welding was relatively small in the 1.5 M H₂SO₄ and HNO₃ solutions, but quite substantial in artificial saliva. The high corrosion rate of the WM (274.4 $\mu\text{m}/\text{year}$), associated with an increased passive current density and a reduced passive range in artificial saliva (Figure 11(c)), strongly suggested that a laser-welded Ti₅₀Ni₅₀ SMA might not be suitable for use in saliva or like environments.

IV. CONCLUSIONS

1. The DSC results indicated that the transformation temperatures of the Ti₅₀Ni₅₀ and Ti_{49.5}Ni_{50.5} SMAs were slightly depressed after welding. The minor change in M_s temperature and no deterioration in the shape-memory character of the WM suggested that laser welding is suitable for joining TiNi SMAs.
2. The XRD results of the Ti₅₀Ni₅₀ SMA confirmed the presence of more B2 phase in the WM than that in the BM at room temperature. An abnormally high intensity of B2₂₀₀ in the WM was also observed, owing to the existence of the $\langle 100 \rangle$ preferred growth directions of B2 during solidification.
3. The laser weld had a higher strength but significantly lower ductility than the unwelded Ti₅₀Ni₅₀. The improved strength of laser welds was associated with the increased amount of B2 in the weld, while the decreased fracture strain was attributed to coarse-grained dendritic structures in the WM.
4. The cyclic deformation of the Ti_{49.5}Ni_{50.5} SMA at room temperature showed that σ_m (the stress required for SIM) and ϵ_p (the permanent residual strain) were higher after welding at a given number of cycles. The higher σ_m and ϵ_p values of the laser weld were certainly related to the more inhomogeneous nature of the WM.

5. Potentiodynamic tests revealed the satisfactory performance of the Ti₅₀Ni₅₀ SMA after welding in 1.5 M H₂SO₄ and HNO₃ solutions. In contrast, the WM exhibited a significantly higher corrosion rate and a less stable passivity (a higher passive current density with a reduced passive range) than the BM in artificial saliva. Accordingly, the use of a welded Ti₅₀Ni₅₀ SMA in saliva or like environments should be more carefully evaluated.

ACKNOWLEDGMENTS

The authors gratefully acknowledge the support of the Republic of China National Science Council (Contract Nos. NSC86-2216-E-002-013 and NSC87-2216-E-002-030)

REFERENCES

1. *Engineering Aspects of Shape Memory Alloys*, T.W. Duerig, K.N. Melton, D. Stockel, and C.M. Wayman, eds., Butterworth-Heinemann Ltd., London, 1990.
2. *SMST-94 Proc. 1st Intl. Conf. on Shape Memory and Superelastic Technologies*, A.R. Pelton, D. Hodgson, and T.W. Duerig eds., MIAS, Monterey, CA, 1994.
3. *SMST-97 Proc. 2nd Intl. Conf. on Shape Memory and Superelastic Technologies*, A.R. Pelton, D. Hodgson, S. Russell, and T.W. Duerig eds., MIAS, Monterey, CA, 1997.
4. M. Nishikawa, H. Tanaka, M. Kohda, T. Naganra, and K. Watanabe: *J. Phys.*, 1982, vol. 43, Suppl. 12, pp. C4: 839-44.
5. J. Beyer, P.A. Besselink, and J.H. Cindenhors: *Proc. Intl. Symp. on Shape Memory Alloys*, Y. Chu, T.Y. Hsu, and T. Ko, eds., Guilin, China, China Academic Publishers, Beijing, 1986, pp. 492-97.
6. A. Hirose, M. Uchihara, T. Araki, K. Honda, and M. Kondoh: *J. Jpn. Inst. Met.*, 1990, vol. 54, pp. 262-69.
7. P. Schloßmacher, T. Haas, and A. Schußler: *SMST-94 Proc. 1st Intl. Conf. on Shape Memory and Superelastic Technologies*, A.R. Pelton, D. Hodgson, and T.W. Duerig eds., MIAS, Monterey, CA, 1994, pp. 85-90.
8. P. Schloßmacher, T. Haas, and A. Schußler: *SMST-97 Proc. 2nd Intl. Conf. on Shape Memory and Superelastic Technologies*, A.R. Pelton, D. Hodgson, S. Russell, and T.W. Duerig eds., MIAS, Monterey, CA, 1994, pp. 137-42.
9. J.-P. Randin: *J. Biomed. Mater. Res.*, 1988, vol. 22, pp. 649-66.
10. G. Rondelli, B. Vicentini, and A. Cigada: *Corr. Sci.*, 1990, vol. 30, pp. 805-12.
11. Y. Oshiki and S. Miyazaki: *Corr. Eng.*, 1991, vol. 40, pp. 1009-25.
12. D. Starosvetsky, O. Khaselev, and J. Yahalom: *Corrosion*, 1998, vol. 54, pp. 524-30.
13. H.C. Lin and S.K. Wu: *Scripta Metall. Mater.*, 1992, vol. 26, pp. 59-62.
14. *JCPDS-ICDD*, 1601 Park Lane, Swarthmore, PA 19081, 1994.
15. R.E. Reed-Hill and R. Abbaschian: *Physical Metallurgy Principles*, 3rd ed., PWS-KENT Publishing Company, Boston, MA, 1992, pp. 444-46.
16. K. Otsuka and K. Shimizu: *Intl. Met. Rev.*, 1986, vol. 31, pp. 93-114.
17. H.C. Lin and S.K. Wu: *Acta Metall. Mater.*, 1994, vol. 42, pp. 1623-30.
18. S. Miyazaki, Y. Ohmi, K. Otsuka, and Y. Suzuki: *J. Phys.*, 1982, vol. 43, Suppl. 12, pp. C4: 255-60.
19. R.S. Dutta, K. Madangopal, H.S. Gadiyar, and S. Banerjee: *Br. Corr. J.*, 1993, vol. 28, (3), pp. 217-21.