Structure, Texture and Magnetic Properties of Laser-Welded Ultrathin Fe–Co–V Foils

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Abstract This research aims to investigate the effect of pulsed Nd:YAG laser micro-welding on the microstructure, texture and magnetic properties of ultra-thin Fe–Co–7.15 wt% V magnetic alloy. Optical microscopy, scanning electron microscopy and electron backscattered diffraction techniques were used to study the microstructural evolutions. Also, vibrating sample magnetometry was used to characterize the magnetic properties. The results showed that the fractions of low Σ coincidence site lattice boundaries in the fusion zone are higher than those of the base metal. The welded samples experience a significant decrease in their magnetic properties. It was found that the formation of new fiber texture in the fusion zone and grain coarsening are the most important factors affecting the magnetic properties.

KEY WORDS: Magnetic properties; Ultra-thin magnetic foils; Texture analysis; EBSD study; Nd:YAG laser micro-welding

1 Introduction

Fe–Co alloys are the materials of choice for applications where high magnetic saturation is a prime design parameter, particularly in aerospace industries and other smart

application. In addition, they have a high Curie temperature, good permeability and good strength [1–5]. These alloys are frequently used in the direct current (DC) motors as rotors or stators in the form of strip. So, it is vital for these alloys to have satisfied ductility. However, the lack of ductility restricts employment of binary Fe–Co alloys in commercial applications [6, 7]. Different heat treatment processes can be used to achieve a wide range of magnetic and mechanical properties in Fe–Co alloys. These alloys experience various phase transformations during heating/cooling cycle. At high temperatures the face-centered-cubic (FCC) γ phase is stable, transforming into base-centered-cubic (bcc) α phase upon cooling. Fe–Co alloys containing 30–70 wt% Co undergo an ordering transformation from disordered BCC structure to ordered Cs–Cl structure below a critical temperature (T_c), 730 °C, which significantly alters their magnetic and mechanical properties [8]. Recrystallization of heavily cold-worked disordered Fe–Co alloys takes place rapidly at ~ 800 °C. At temperatures below the T_c , partial recrystallization, recovery and ordering phenomena may occur [9, 10]. Increase of temperature to higher temperatures leads to the

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grain growth which affects magnetic and mechanical properties [11].

Previous studies show that vanadium addition highly promotes the ductility of Fe–Co alloys so that it is quite possible to produce very thin foil of these alloys [12–15]. On the other hand, depending on the amount of vanadium, the coercive force (H_c) of these alloys varies in the range of 4.0–24.0 kA/m and their residual induction (B_r) in the range of 0.6–1.25 T [15]. Attractive combination of magnetic and mechanical properties in Fe–Co–V alloys makes these alloys unique candidates for producing active parts of rotors in synchronous hysteresis motors (SHMs) [15]. Due to the various phase transformations and structural changes during heating of these alloys, it seems that changes in the magnetic and mechanical properties during welding are inevitable. Moreover, in order to deploy Fe–Co–V alloys in the rotor of SHMs, it is essential to fabricate them in the form of very thin foils which causes more complexity in the welding of them. This requirement has been raised with the goal of obtaining increasingly denser and smaller structures and devices and to minimize eddy current losses [16]. Reduction of losses along with reduction of weight is essential for energy efficiency in the transportation industry [17, 18]. Laser welding is a viable, highly attractive method due to its numerous advantages including elimination of flux; narrow width and deep penetration; high mechanical strength; great resistance to vibration and shock; and minimum degradation of heat sensitive components during assembly [19–23]. These characteristics make laser welding process a suitable alternative for joining ultra-thin foils of Fe–Co–V alloys. Despite its importance, however, laser welding of these alloys has not received enough attention. Comprehensive studies are still required.

This paper is mainly aimed to point out the interplay between weld structure and magnetic properties in a laser-welded Fe–Co–7.15 wt% V alloy. In this context, texture evolution and its relationship with magnetic properties are explored. The authors believe that the obtained results in this work would help with exploiting full advantages of this alloy.

2 Materials and Experimental Procedure

2.1 Materials

In the present study, ultra-thin foils of a Fe–Co–7.15 wt% V alloy with thickness of 0.05 mm were provided in cold-rolled condition.

2.2 Welding Process

Foils with dimension of 0.05 mm × 1,000 mm × 25 mm are tightly rolled up around the holder. The end of foil is

Table 1 Test conditions used in this study for samples A and B

Sample	Power (W)	Pulse duration (ms)	Travel speed (mm/min)	Focus position (mm)
A	54	2.625	87.5	−1.625
B	68	2.625	87.5	−0.875

Fig. 1 SE-SEM micrograph of the as-received Fe–Co–7.15 wt% V alloy

fixed by glue to prevent any gap between the foils and minimize weld distortion. The holder begins to rotate and simultaneously the laser beam irradiates the center of specimen surfaces. A pulsed Nd:YAG laser with a maximum mean power of 800 W was used for welding the magnetic foils. Argon shielding gas was provided at a flow rate of 15 L/min to prevent the effects of moisture and contaminations. In order to study the effect of heat input on the welds, two samples were welded following the conditions listed in Table 1.

2.3 Metallurgical Analysis

Microstructure of the welded area was analyzed using optical microscope (OM; Olympus BX50) and scanning electron microscope (SEM; Philips XL30). X-ray diffraction (XRD) measurements were carried out using a Bruker 2D system, with CrK_α radiation to determine the texture and phase identification of the welded samples.

Hitachi scanning electron microscope (SEM) equipped with energy dispersive X-ray spectroscopy system (EDS) and electron backscattered diffractometer (EBSD) was

Fig. 2 OM micrographs of weld bead geometry for sample A **a**, sample B **b**, macromorphologies of the surfaces of sample A **c**, sample B **d**, SEM micrographs of the near the HAZ of sample A **e**, sample B **f**, FZ center for sample A **g**, sample B **h**

used in structure analysis. The HKL CHANNEL5 software was used to perform EBSD data visualization and post processing.

The room temperature magnetic hysteresis ($B-H$) loops of the welded samples were measured by vibrating sample magnetometer (VSM, Meghnatis Kavir Kashan Co.).

3 Results and Discussion

3.1 Microstructural Evolutions

3.1.1 OM and SEM Observation

The as-rolled base metal (BM) was annealed at 550 °C for 2 h in a vacuum furnace and then cooled in the furnace. During the heat treatment, ordering transformation and recrystallization occur [24]. Figure 1 shows a typical SEM

micrograph of the BM after heat treatment. It is shown that the microstructure consists of very fine recrystallized grains.

Figure 2 shows the OM and SEM micrographs of the weld beads geometry and microstructure of FZ and near the heat-affected zone (HAZ) of the welded samples. It is evident in Fig. 2a, b that the weld cross section of sample B is larger than that of sample A due to the increase in heat input. Figure 2c, d shows the macromorphologies of the weld surfaces for the samples A and B, respectively. The middle of the laser welding zones can be considered as a demarcation line, and the crystal solidified macrostructures grow with different orientations from either side of the demarcation line, which is “fish bone” like.

During welding, the joint region experiences significant thermal gradient from the FZ to the BM; this temporal distribution of temperature produces a non-uniform microstructure. The solidification microstructure of the FZ

Fig. 3 Microstructural features of the base metal: **a** EBSD map; **b** misorientation distribution

Table 2 Fraction (%) of the special CSL boundaries after laser welding

Position	$\Sigma 1$	$\Sigma 3$	$\Sigma 9$	$\Sigma 27$	$\Sigma(1-29)$
Base metal	77	1.16	0.45	0.14	80.94
Interface of HAZ and FZ	76	2.3	0.73	0.17	84.27
Center of FZ	63	7	0.023	0	77.25
Near the HAZ (inside the FZ)	75	5.84	0	0	86.64

depends on these parameters: temperature gradient (G), growth rate (R), undercooling (ΔT), and alloy composition. The ratio G/R determines the mode of solidification [25]. Planar and cellular structures are formed instead of dendritic structure if G/R ratio is high enough. In Fig. 2e, f, it is readily observed that grains initially grow in planar form and then the cellular grains grow epitaxially from the interface following the easy-growth direction of the parent Fe–Co–V grains [25]. However, intense convection occurred within the welding pool is unlikely to allow the formation of very large directionally oriented cellular grains [25]. Near the weld centerline, fine equiaxed dendrites nucleate and grow, blocking the columnar dendrites (Fig. 2g, h). The thermal gradient (G) decreases from the fusion boundary to the weld centerline, while the solidification rate (R) increases. The reason can be described by [25]:

$$R = S \cos \alpha, \quad (1)$$

where S is the welding speed and α is the angle between the welding direction and normal to fusion boundary. Consequently, there is a decrease in the ratio of G/R from the fusion boundary toward the weld centerline, which is associated with a decrease in undercooling. In other words,

Fig. 4 Microstructural features of the interface of HAZ and FZ

the undercooling reaches its minimum magnitude near the weld centerline, which is the reason for the equiaxed dendrite formation. Change in solidification mode in the fusion zone (FZ) of other welded materials such as Ni-based alloys and steels have been reported by Zhang *et al.* [20] and Ventrellaa *et al.* [23]. It is well known that the formation of new fine grains can originate from various sources in welding: (1) the growth of preexisting nuclei

Fig. 5 Microstructural features of the center of FZ in sample B: **a** EBSD map; **b** misorientation distribution

created by dendrite fragmentation, or by grain detachment; (2) heterogeneous nucleation at second-phase particles present in the melt, or impurities in the weld pool; and (3) surface nucleation by exposing the weld pool to a stream of cooling gas or by instantaneous reduction or removal of the heat input [25]. Considering the absence of any impurity and second phase in the FZ, it can be said that heterogeneous nucleation is a less probable mechanism. So, other mechanisms, particularly surface nucleation (due to the presence of cold shielding gas and pulsed laser), are the predominant mechanisms for the nucleation of fine equiaxed grains in the center of FZ.

3.1.2 EBSD Observation

EBSD was used to study the grain boundary evolution across the welded samples. The results reveal the presence of a strong texture in the BM. In addition, it is observed that the initial microstructure is in a recrystallized state and consists of very fine equiaxed grains with an average grain size of about 482 nm. As shown in Fig. 3, most of the grain boundaries are low angle in character (LAB fraction

is $\sim 77\%$) and only sporadic high angle grain boundaries (HABs) are clustered within some grains.

Misorientation distribution is characterized by a sharp low-angle peak and some clustering of the misorientation axes near 60° . It is well established [26–29] that low Σ CSL boundaries (usually $\Sigma \leq 29$), especially $\Sigma 3^n$ ($n = 1, 2, 3$) boundaries possess special chemical, mechanical, electronic, kinetic and energy-related properties. Furthermore, the energy of the coherent twin boundaries ($\Sigma 3$) is much smaller than that of other boundaries [30]. Table 2 lists the fraction of the special CSL boundaries in the various regions. In the BM, the fraction of CSL boundaries is only 3.94%, which is composed of 1.16% of $\Sigma 3$, 0.45% of $\Sigma 9$ and 0.14% of $\Sigma 27$.

Figure 4 presents the EBSD map taken from the interface of heat-affected zone (HAZ) and FZ. This map clearly shows elongated grains in the FZ. It is deduced that recrystallization has been completed in the HAZ and accordingly the fine equiaxed grains appear due to the temperature rise in the HAZ. The presented area in Fig. 4 consists of 8.37% CSL boundaries which is higher than that

Fig. 6 Microstructural features of near the HAZ in sample A: **a** EBSD map; **b** misorientation distribution

of BM. This value composed of 2.3% $\Sigma 3$, 0.73% $\Sigma 9$ and 0.17% $\Sigma 27$. Completion of recrystallization in the HAZ is the most possible reason of increase in the fraction of CSL boundaries.

EBSD map from the center of the FZ of sample B and the corresponding distribution of misorientation angles are presented in Fig. 5. The center of FZ (Fig. 5a) is characterized by the extension of HABs. In this region, the fraction of LABs has been decreased to 63%. This means that this region experiences higher temperature, leading to the extended formation of HABs. Center of the FZ has been consisted of very fine equiaxed grains with an average grain size of 1.8 μm . It is also noticed that the fraction of CSL boundaries has been increased up to 14.25%, and this fraction includes a larger amount of twin boundaries ($\sim 7\%$).

The EBSD micrograph from near the HAZ (inside the FZ) of sample A and its associated misorientation distribution histogram are shown in Fig. 6. Elongated grains in this region are obvious with an average grain size of about 3.74 μm . Misorientation angle distribution shows that most of grain boundaries are LAB ($\sim 75\%$). However, there is also a sign of bimodal distribution of grain boundaries in the investigated

region. It should be noted that the fraction of LAB seems to decrease and that of HAB to increase. In this region, the fraction of HAB is about 25% which shows a little increase relative to the BM. This region consists of about 11.64% CSL boundaries which is noticeably higher than that of the BM, 3.52%. Similar result has also been reported by Yang et al. [29] who found that during laser surface melting of 304-stainless steel, the FZ microstructure develops along $\langle 100 \rangle$ direction and large amount of CSL boundaries are formed.

3.2 Texture Evolution

Figure 7 shows the $\phi_2 = 0^\circ$ and 45° sections of ODFs for the as-received sample and FZ of welded samples, respectively. Texture of the as-received sample in Fig. 7a has a dominant α -fiber and a weak γ -fiber. Also, rotated Cube and relatively weak Cube texture component can be observed in the section of $\phi_2 = 0^\circ$. This finding is similar to the results of Deng et al. [31] who found that Fe–Co–2 V alloy after cold rolling and subsequent annealing at 600 $^\circ\text{C}$ consists of a perfect α -fiber, weak γ -fiber and also some rotated Cube and Cube texture components. But after laser

Fig. 7 Textures of specimens revealed by $\phi_2 = 0^\circ$ and 45° , ODF sections of the ideal orientation positions for the main texture components **a**, BM **b**, FZ of sample A **c**, FZ of sample B **d**

Fig. 8 Intensity distributions along α -fiber **a**, γ -fiber **b**

welding, the intensity of α -fiber noticeably decreases, rotated Cube and Cube texture components, however, disappear and γ -fiber becomes stronger. The quantitative development of texture for the as-received sample and FZ of the welded samples can be best described by plotting the orientation densities values along certain lines as shown in Fig. 8. It is noticed in Fig. 8a that the α -fiber, and

especially those texture components like, $\{001\} \langle 110 \rangle$ and $\{112\} \langle 110 \rangle$, become weaker for welded samples. Component of $\{001\} \langle 110 \rangle$ is the rotated cube component in α -fiber, and this component is often present after recrystallization. As shown in Fig. 8b, the intensity of γ -fiber (specially $\{111\} \langle 110 \rangle$ component) becomes stronger in the FZ of the welded samples.

Fig. 9 Inverse Pole Figures for **a** BM, **b** FZ of sample A, **c** FZ of sample B

Figure 9 shows the IPFs for the as-received sample and FZ of the welded samples. It is evident that the intensity of $\langle 001 \rangle$ direction for the as-received sample is higher than the other direction. But, after laser welding, the intensity of $\langle 001 \rangle$ direction decreases and the intensity of $\langle 111 \rangle$ becomes stronger.

3.3 Magnetic Properties of Laser-Welded Samples

The values of saturation induction (B_s) for the as-received sample and welded samples are listed in Table 3. The B_s value for the as-received sample (2.09 T) has been increased after laser welding (2.29 and 2.26 T for sample A and sample B, respectively). Increase in B_s value can be

Table 3 Magnetic properties of the as-received and welded samples

Sample	B_s (T)	$(BH)_{\max}$ (kJ/m ³)	B_r (T)	H_c (Oe)
As-received	2.09	54.4	1.043	150
A	2.29	3.33	0.187	74.1
B	2.26	2.01	0.152	61.57

interpreted by $\gamma \rightarrow \alpha$ phase transformation which occurs during laser welding.

XRD patterns for the as-received sample and FZ of welded samples are shown in Fig. 10. The diffraction pattern of the as-received sample shows some peaks corresponding to γ and α phase. But, it is observed that after laser welding, there are only α peaks, the γ peak has been disappeared. It is well known that the γ phase is a paramagnetic phase. The presence of this paramagnetic phase in the as-received sample leads to a decrease in B_s value [32]. But, after laser welding, the paramagnetic γ phase disappears and consequently the B_s value of the welded samples increases. Elimination of γ phase in the welded samples might be due to the residual stresses induced by laser welding. Rapid cooling during laser welding leads to the formation of residual stress in the FZ. According to the findings of Zakharov *et al.* [15], residual stress-induced deformation leads to the transition of paramagnetic γ phase to ferromagnetic α phase and thus to the raising of B_s value.

Another important factor that determines the performance of a permanent magnet is the maximum energy product ($(BH)_{\max}$). $(BH)_{\max}$ of a permanent magnet is given

Fig. 10 XRD patterns for the as-received sample and FZ of welded samples

by the demagnetisation curve of the second quadrant of the B – H hysteresis loop. This extends from the B_r to H_c . In other words, in order to get large $(BH)_{\max}$, the values of B_r and H_c of the magnets should be large. The changes of H_c , B_r and $(BH)_{\max}$ for the as-received sample and welded samples are also listed in Table 3. It is obvious that the values of H_c , B_r and $(BH)_{\max}$ decreased after laser welding. Two factors have been considered as the most important reasons responsible for the decrease in these parameters. The first factor is texture evolution. As discussed in Sect. 3.2, a new texture is produced upon welding. This is important because some magnetic properties such as B_r can be affected by orientation of grains [32]. As discussed above, by laser welding, Cube and rotated Cube texture component have been disappeared, whereas γ -fiber becomes the dominant texture in the FZ. The intensity of α -fiber also decreases after laser welding. It is well established that in the Fe–Co–V alloys containing more than 41% of Co, magnetization is hardest in the $\langle 001 \rangle$ direction [33]. In other words, direction of $\langle 001 \rangle$ is the hard axis of magnetization and directions of $\langle 011 \rangle$ and $\langle 111 \rangle$ are the medium and easy axis of magnetization, respectively. In this respect, it can be said that Cube and rotated Cube texture and α -fiber are preferable for higher B_r value in these alloys since they have hard and medium magnetization directions. This means that in this condition, magnetic moments do not need to rotate toward the external magnetic field and therefore in the absence of magnetic field, $H = 0$, the B_r value is high. On the other hand, the γ -fiber which represents groups of orientations for $\langle 111 \rangle$ parallel to the normal direction is formed in the FZ. Thus, according to the findings of Hall [33], presence of this fiber texture enhances easier magnetization direction. Therefore, the formation of grains in the FZ with an enhanced orientation parallel to the easier magnetization direction leads to the decrease in B_r .

Change of grain orientation can also be deduced from IPF plots in which $\langle 001 \rangle$ fiber texture has been weakened in the FZ of welded samples.

The second factor is grain size which can significantly influence the magnitude of H_c . Surmail [34] reported that the coercivity of most existing FeCo-based alloys is dictated essentially by the grain size, and a unique line can be used to approximate the dependence:

$$H_c = 36.7 + 777.5/d, \quad (2)$$

where H_c is the coercivity, in A/m; and d the grain size in μm .

Dependence of H_c on the grain size implies that grain boundaries serve as pinning sites for the magnetic domain wall. So, the formation of larger grains in the FZ is responsible for the decrease in the H_c value of welded samples. As a results, it is deduced that the decrease in B_r

and H_c leads to the reduction of surrounded area in the second quadrant of B – H loop and hence the $(BH)_{\max}$ value decreases.

4 Conclusions

Ultra-thin Fe–Co–7.15 wt% V magnetic alloys have been welded using Nd:YAG laser. On the basis of the present results, the following conclusions can be reached.

1. It was found that a gradient of microstructure is formed in the FZ, where solidification mode is changed from planar in the adjacent of HAZ to equiaxed dendritic in the center of FZ.
2. Texture of the as-received sample which consists of strong, γ -fiber, rotated cube and weak cube texture components, changes after laser welding. Rotated cube and cube textures disappear after welding and the γ -fiber develops. Presence of γ -fiber weakens the magnetic properties.
3. EBSD study shows that the fraction of low Σ CSL boundaries, especially twin boundaries increase during laser welding.
4. $(BH)_{\max}$ value of welded samples encounter with a considerable decrease. Change of grain orientation from hard axis to easy axis of magnetization and also grains coarsening are the main factors which lead to decrease in the mentioned property.

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