

EFFECT OF PRE-POST TIG WELDING HEAT TREATMENT ON CAST NI SUPERALLOY

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ABSTRACT

The effect of a pre (before) and post (after) heating welding treatment on the microstructure and mechanical properties of the scrap blades made of cast INC738LC superalloy is the main goal of the present investigation. The filler used in TIG welding was a INC 625 solution hardened superalloy as the proposed solution for hot cracking of the INC738LC cast superalloy in literature. The TIG welding was processed with respect to the constantly optimized parameters (current, voltage, speed, gas flux rate and number of passes) to make a mechanical properties comparison between the as received superalloy and the welded superalloy with heat treated specimens. The characterization techniques employed in this study are hardness measurements, tensile tests, optical microscopy and scanning electron microscopy. We found that the proposed preheating improves the TIG welding of the INC 738 LC superalloy specimens and the post welding heat treatment enhances its mechanical properties.

KEYWORDS: Ni-base superalloy, Microstructure, Heat treatment, Welding, Precipitation.

1. INTRODUCTION

Nickel-based superalloys are typically used in the combustor and turbine sections where elevated temperatures are maintained during operation [1]. Due to the poor machinability of Ni-superalloys, one of the important manufacturing approaches is vacuum investment casting [2]. Before being used in service, the as-cast components are often subjected to complex heat treatments designed to establish a controlled size and distribution of precipitates [1]. They must achieve the best combination of corrosion resistance and creep strength [3] as a result of their submission to a standard heat treatment [4].

The standard heat treatment of the INC 738 LC Ni superalloy consists of a solution heat treatment for 2 hours at 1120°C air cooling+ aging treatment at 845°C for 24 h air cooling, which will produce bimodal γ' (Ni₃Al) precipitates [5] consisting of primary cubical γ' precipitates and spherical secondary γ' precipitates. Their sizes are cited to be 40 nm to 70 nm for the fine precipitate, and 450 nm to 700 nm for the coarse precipitate[6]. The hardness of the cast INC 738 LC Ni superalloy rises to a maximum value (Hv \approx 485) at 827°C [7].

It is recognized that during long term severe service operation, blades made of INC738LC Ni

superalloy undergo a series of time, temperature and stress dependent microstructural changes as porosity, micro shrinkages and other inhomogeneities in casting parts have been reported to limit their high temperature mechanical properties [8], such as creep strength and resistance to cracking [9].

Many researchers stated that the degradation is due to the formation of secondary phases [8], i.e.; the decomposition of γ' precipitates (loss in volume fraction) by the appearance of the β phase [10], [11] or often overheating and subsequently corrosion [5]. Other researchers postulated that the failure is due to the decomposition of the carbides and borides which are the typical grain boundaries strengtheners [12].

The refurbishment of the pricey components of the gas turbine blades is preferred to replacing them, which is due to the high manufacturing cost of the hot parts of the damaged gas turbine blades used for power generation [13], especially those of first stage turbines made of the cast INC738LC Ni superalloy.

One of the economical industrial processed commonly used to repair the damaged blades of a gas turbine is the fusion welding TIG-Tungsten Inert Gas arc welding or GTAW [14]. TIG welding of the INC738LC Ni-superalloy blades without optimization of the welding parameters [15] can produce hot cracks or the grain-boundary liquation cracking due to the

boron segregation [13]. Many studies report that the solution to the hot cracking mechanism during the welding of the Ni superalloy and the post-welding heat treatment is avoided by suitable and practical preheating [16, 8]. It was also mentioned that a preheating at 1055°C can make the material ductile, with a hardness of about 340 HV, avoiding the formation of cracks after TIG welding [7], [8], while others propose the use of a hardened superalloy as INC 625 [17]. The thermal cycles during the welding of the Ni-base superalloy can decrease the mechanical properties of the welded joint [18]. Finally, after the TIG welding, the joined parts of the INC 738 LC Ni superalloy were subjected to post-welding heat treatment, its ultimate goal being to obtain a Ni superalloy with important mechanical properties [19] and to improve material service life [20].

The analysis of current literature shows that there is little research related to the mechanical properties of the cast Ni-base superalloy joints, using optimized parameters of TIG welding process and combined thermal regime (preheating and post-weld heat treatments).

2. EXPERIMENTAL STUDY

The Ni-base super-alloy used in this study was a INC738LC trade mark, with low carbon content. The sample was provided in an as-cast state, from the blade of a first stage gas turbine in the Electric Power Plant of M'sila, Algeria. The analysed blade had been in service for about 64,000 h. Table 1 shows the nominal composition of the received base material.

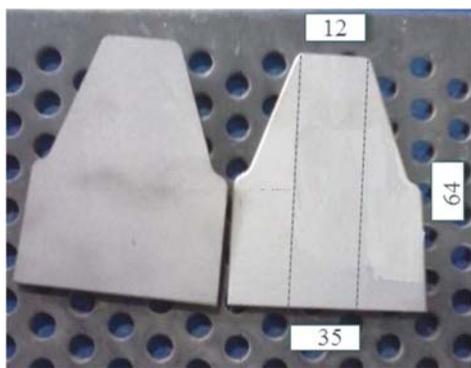


Fig. 1. The 2 mm thick plates cut from the blade root of the Ni superalloy INC738LC

Table 1. Nominal composition of the Ni based superalloy material used [15]

Element	Wt [%]	Element	Wt [%]
Al	3.65	Fe	0.23
Nb	0.86	Co	8.20
Mo	1.74	Ta	1.39
Ti	3.08	W	2.95
Cr	15.40	Ni	Bal.

The shape of the INC738LC superalloy plates is shown in figure 1. The samples were cut from a gas turbine blade (MS9001-1st stage), using a wire-discharge cutting machine (EDM"AF 35", laboratory of the Maintenance of Industrial Equipment in the MIE company, Algeria) [21].

2.1. Microhardness Measurement

The microhardness of the as received INC738LC base material has been measured using a microhardness apparatus (LECO M-400-A hardness tester) and an indentation force value of 5 N for 15 seconds. The mean value of microhardness, obtained in 10 different measuring points, was 398.17 HV. The application of preheating is considered mandatory.

2.2. Preheating

Before welding the plates were subjected to the proposed preheating, as seen in figure 2. It consists of a (re) solution treatment followed by primary precipitation treatment (at 1055°C/4 hrs/WQ “water quench”) which may lead to the regeneration of carbides and borides, providing some ductility to the superalloy. This preheating was inspired from the heat treatment applied by Danis & al. [22] and Xu & al. [8]. The proposed preheating cycle is an improvement in terms of time efficiency as compared to the time used for the preheating cycles of the U.M.T “University of Manitoba Heat treatment” by decreasing to 66.67 %, to 77% for the NUMT “New U.M.T” and to 62.5 % for the FUMT “Furnace U.M.T” [23], thus saving the amount of the heat energy needed for preheating. Before the TIG welding process, plates were chamfered and a deep V groove of 35° angle was machined.

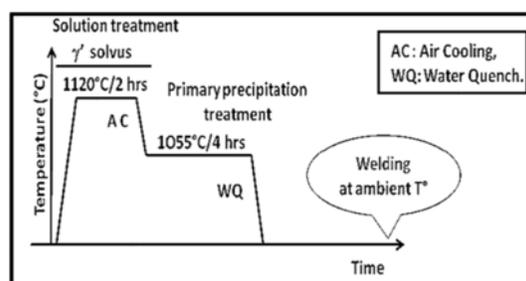


Fig. 2. The proposed diagram of preheating



Fig. 3. Plates chamfered

2.3. Post-welding Heat Treatment

After welding, post weld heat treatment (PWHT) is essential to improve the homogeneous microstructure and mechanical properties in the weld zone [20]. To do so, the welded parts were subsequently subjected to post-welding heat treatment as shown in figure 5, as defined by Wangyao & al. [19].

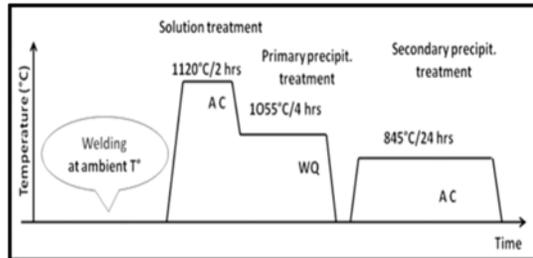


Fig. 5. The diagram of post-welding heat treatment

2.4. Welding Process

For the TIG welding process a Miller Syncrowave 350 L apparatus (XMT 450 CC/CV) was used. The values of the optimized welding parameters (that were kept quite constant) are shown in table 2. Before welding, the joining surfaces were slightly treated with sandpaper P240 and cleaned with acetone. The welded joint obtained is shown in figure 4.

Table 2. TIG welding parameters

TIG welding parameters	Millersyncrowave 350 L × (XMT 450 CC/CV)
Welding speed [mm/s]	0.4
Argon gas rate flow [l/min]	8
Number of passes	2
Welding current [A]	40
Welding voltage [V]	10
Filler used INC 625 Ø [mm]	1.6
Deep V groove angle [°]	35
Space (distance) between Ni superalloy plates [mm]	2

INC625 filler was selected as suggested by Banerjee & al. [17] and its chemical composition is shown in table 3.

Table 3. Chemical composition of the INC625 filler

Element	Wt [%]	Element	Wt [%]
Al	0.40	Co	1.00
Nb	3.31	Ti	0.13
Mo	8.39	Ta	2.74
Cr	20.95	Ni	Bal.
Fe	0.27		



Fig. 4. The welded specimens

2.5. Machining Tensile Test Specimens

The tensile test specimens were cut from the welded plates after the post-welding heat treatment by use of the numerically controlled wire discharge machine (EDM"AF 35"). Figure 6 shows the shape and the measurements of the tensile experimental specimen. The welded zone is in the middle of the tensile-test specimens.

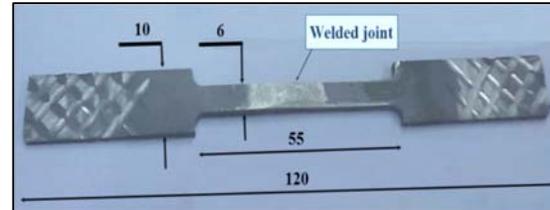


Fig. 6. The tensile test specimen

2.6. Metallurgical and Mechanical Characterizations of Welds

2.6.1. Metallographic Characterization

The metallographic observations of the base metal structure, welded and heat-treated parts were carried out by means of the computer-assisted "Leica Microsystems Belgium BVBA-DMLM" type microscope. The surface examination of the sample was possible by applying (Marble's reagent) for 30 seconds as dwell time. The average volume fraction or population of coarse γ' particles was determined by the surface area measurement by means of the "image J" software.

2.6.2. Mechanical Characterizations

The as received metal and the pre-post welding heat

treated plates were subjected to the mechanical tests (microhardness measurements, tensile tests). The hardness of the pre-welding metal was taken in mean value (10 measurements). The hardness of the welded and post-welded heat-treated joints was quantified with a space of 0.5 mm from the base metal to the welded zones. The three specimens of INC738LC Ni superalloy as received and the three-welding heat treated specimens were also subjected to tensile tests by using a Zwick/Roell Z100 tensile experimental machine at a constant speed of 2 mm/min, as seen in Saib and Boumerzoug [15].

2.6.3. Fracture Faces SEM Observations

The cross-section observations of broken specimens after the tensile tests were performed by Scanning Electron Microscopy (SEM) TESCAN VEGA Easy Probe.

3. RESULTS AND DISCUSSION

3.1. Microstructure Observations

3.1.1. Microstructure of the as-received Metal

The as-received INC738LC superalloy was obtained from a blade, which had been operating in a gas turbine engine at high temperature for several years. Figure 7 shows the base metal microstructure. A dendritic structure is observed, which indicates that the blade was produced by a conventional casting process. It can be observed that the alloy has a multi-phase microstructure composed of the FCC γ austenitic matrix, bi-modal γ' Ni₃(Ti, Al) intermetallic precipitates (primary “coarse” and secondary “fine” ones), γ - γ' eutectic, carbides [20]. The coarse γ' precipitates had a 34 % volume fraction.

3.1.2. Microstructure of the Preheated Ni Superalloy

3.1.2.1. Microstructure of the Ni Superalloy after (re) Solution- Heat Treatment at 1120°C /2h/Air Cooling

Figure 8 shows the γ' precipitates and the M₂₃C₆ carbides. The distinction from the (re) solution treatment (quenched) metal is that it contained a large fraction of fine γ' precipitates and spherical MC carbides (in grain boundary and intergranular) as compared to coarse γ' precipitates. The measured volume fraction of the coarse γ' precipitates decreased to 20%.

3.1.2.2. Microstructure of the INC738LC after the Primary Precipitation Heat Treatment (1055°C/4h/WQ)

Figure 9 shows the microstructure of the INC738LC Ni superalloy after isothermal heat treatment at 1055°C, which is called primary precipitation.

The output of the volume fraction of the coarse γ' precipitates increased to 23 %. A segregation between the grains is observed, with the existence of a large fraction of blocky MC carbides (in grain boundary and intergranular). The shape of grain boundaries has a continuous and finer form which is the suitable shape to weld the superalloy without cracks, as mentioned by Thakur [7].

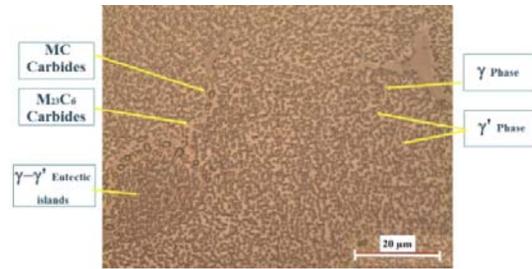


Fig. 7. Microstructure of the as-received metal

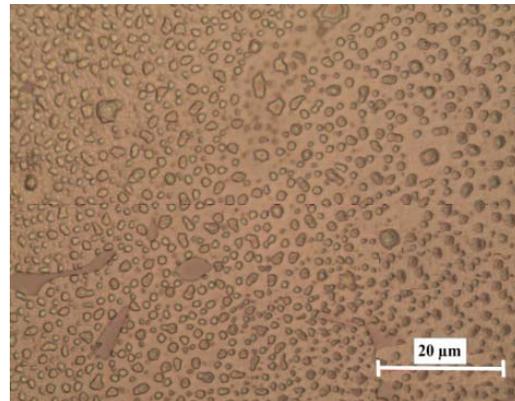


Fig. 8. Microstructure of the INC738LC superalloy after (re) solution-heat treatment at 1120°C/2h/air cooling

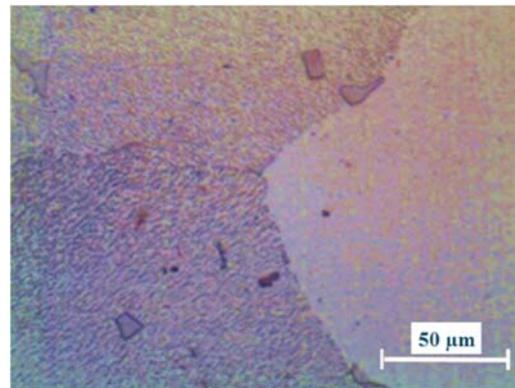


Fig. 9. Microstructure of the INC738LC Ni superalloy after primary precipitation heat treatment

3.1.3. Microstructure of the Welded Ni Superalloy

Figure 10 shows the different zones of the INC738LC Ni superalloy microstructure welded joint. The welded joint microstructure differs significantly from the parent metal [24].

It can be observed a segregation in the microstructure of the base metal between the grains, one with fine γ'

precipitates and another with bimodal γ' precipitates, with 24% as volume fraction of coarse γ' precipitates, in addition the presence of different carbides, borides; these constituents continued to exist in the HAZ at the base metal side. Laves phases are rejected to the interface between FZ and HAZ. In FZ, we detected the γ'' precipitation and the appearance of the δ phase, a small amount of the γ' precipitation, and Laves phase had the bulk appearance of the MC carbides.

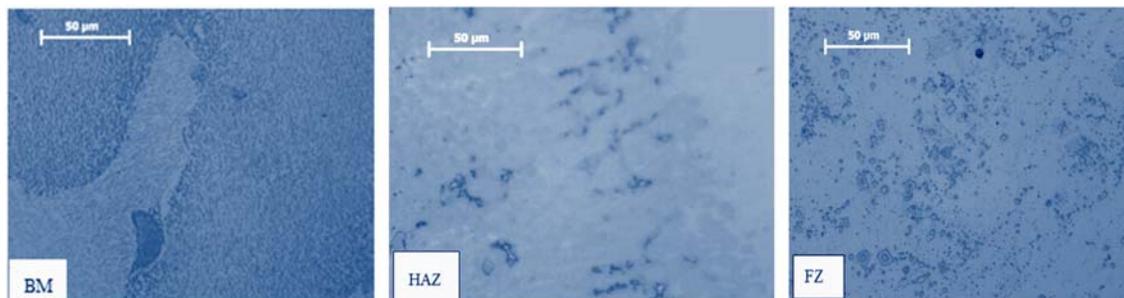


Fig. 10. Microstructures in the (BM, HAZ, FZ) of INC738LC Ni superalloy after welding

3.1.4. Microstructure of the Post-welding Heat Treated Ni Superalloy

Figure 11 shows the different zones of the welded joint microstructure after the post-welding heat treatment of the INC738LC Ni superalloy. We noticed the presence of the coarse γ' precipitates near the base metal, where their volume fraction was 44%. On the FZ side, there are very fine γ' precipitates.

However, the difference in the γ' precipitates size between coarse and very fine is due to the formation of atoms of γ' precipitates, which could diffuse in more amount and/or more distance from the base metal zone to FZ after complete dissolution of all γ' precipitates during higher temperature of (re) solution treatment, and finally re-precipitate in uniform size after the aging heat treatment, as described by Wangyao & al. [19].

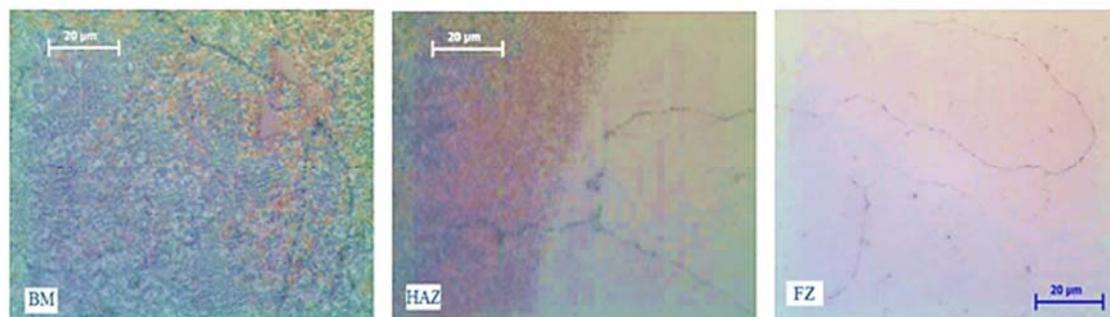


Fig. 11. The microstructures in the (BM, HAZ, FZ) of INC738LC Ni superalloy after post-welding heat treatment

3.2. Mechanical Characterization of INC738LC Ni Superalloy

3.2.1. Microhardness Measurements

3.2.1.1. The as-received Cast INC738LC Ni Superalloy and the filler

The mean microhardness value (10 measurements) of the base metal Ni superalloy INC 738 LC as received, was: 398.17 Hv. On the other hand, the mean

microhardness of the INC625 filler metal was evaluated to have the value of 256.41 Hv.

3.2.1.2. The Preheating of the INC738LC Ni Superalloy

After the (re) solution heat treatment (1120°C/2h/AC), the average microhardness of the base metal became 418.59 Hv. However, the average microhardness of the Ni superalloy after the primary precipitation heat treatment (4 hours at 1055 °C/WQ) reached the value of 351.44 Hv.

The hardness of the metal decreases when the coarser γ' increases a little in volume fraction in comparison to the fine γ' precipitates.

3.2.1.3. The welded INC 738 LC Ni Superalloy

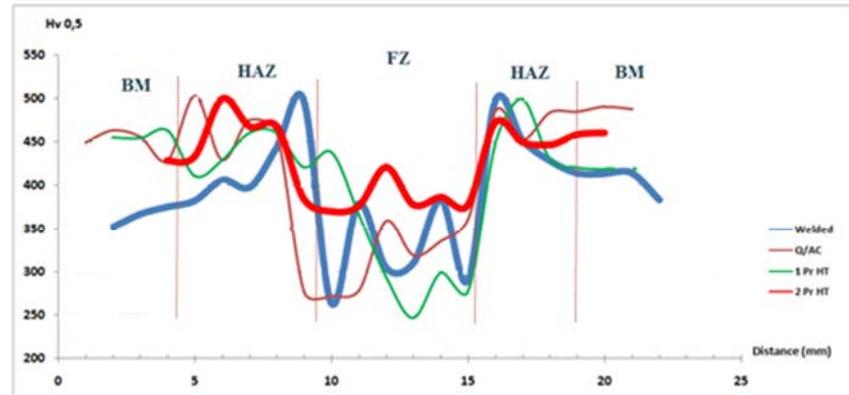
The values of the mean microhardness measured on the different zones of the welded joint are figured by the thick blue line in figure 12. After welding, high hardness values were obtained in the heat affected zone (HAZ), of about 500Hv, while the hardness in the FZ was 370 Hv, and about 350 Hv in the base metal.

3.2.1.4. The Post-welding Heat Treated INC 738 LC Ni Superalloy

The values of the mean microhardness measured in the different zones of the welded bead after the different cycles of post-welding heat treatment are

represented by the thin brown, thin green and thick red lines in figure 12. After welding followed by the (re) solution (quenching) and (re) precipitation treatment, the microhardness of the base metal was 450 Hv, whereas the HAZ approached 500 Hv, and the FZ was equal to 410 Hv. This minor raise in hardness was due to the (re)precipitation of the strengthening phases γ' , γ'' , δ .

It was found that in the FZ, the precipitation of δ , γ'' increased the yield strength. The high hardness values might also be due to the very dense or very fine γ' precipitates in the base metal and in HAZ, together with more γ' precipitates in the weld filler near the HAZ [25]. It was found that these γ' precipitates become coarser due to more dispersion by other elements such as aluminium and/or titanium from the base metal into FZ, followed by the formation of a coarser size of γ' precipitates in this area [26].



(Q/AC: quench air cooling, 1 Pr HT: 1st precipitation heat treatment, 2nd Pr HT: 2nd precipitation heat treatment)

Fig. 12. Micro hardness measurements after welding and post-welding thermal cycles

3.2.2. Tensile Experimental Study

Tensile experiments were carried out at room temperature 20 °C on base metal (as received) specimens and on heat treated welded specimens (figure 14), thus allowing to make a comparison of their mechanical characteristics. The tensile tests results are shown in figure 13 and table 4.

Figure 13 shows the tensile curves where we noticed that the tensile curves of the specimens consist of the following conventional zones: an elastic zone followed by a plastic zone, and finally the rupture. The higher mechanical strength (UTS) is realized in the base metal specimens followed by the heat treated post-welded specimens respectively. The ultimate tensile strength of the heat-treated welded specimens was similar to the as received metal (717.44 MPa \approx 721.21 MPa).

The ultimate tensile strength of the heat-treated welded specimens was higher by (66.38 N/mm² \approx 70

MPa) and the Young's modulus by 42 GPa respectively, as compared to the welded specimens with the optimized TIG welding parameters that had not undergone any post heat treatment, as reported by Saib and Boumerzoug [15].

Table 4. UTS results

Specimens	UTS [MPa]	E [GPa]
Cast reference [8]	765	200
BM (Base metal-as-received)	721.21	191
WOP (welded with optimized parameters) [16]	651.06	134
HT (Welded Heat Treated)	717.44	176

Figure 14 shows that the broken welded heat-treated specimens after tensile tests were fractured in the FZ very close to the HAZ.

3.2.3 Fractography

SEM observations were performed to analyse the nature of the fracture faces for the weld seams after the tensile experiments on the base metal (as received) specimens and the welding specimens with different thermal cycles, to relate the fracture characteristics to structure and properties. The fractured surfaces of the broken specimens of the base metal showed the cleavage fracture [27].

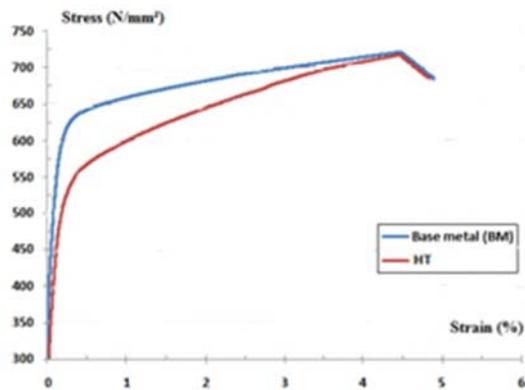


Fig. 13. Tensile curves of the post-welded heat-treated specimens (HT) versus the base metal (as-received)

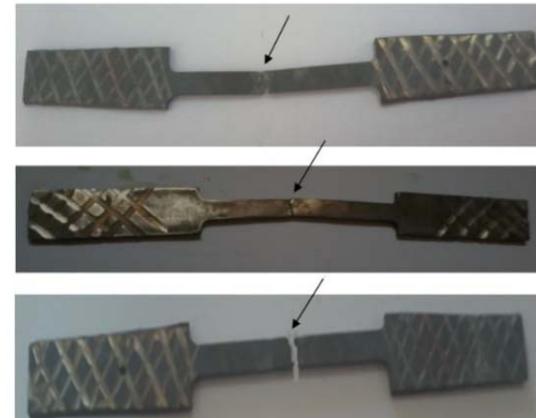


Fig. 14. The broken post-welding heat treated specimens after tensile test

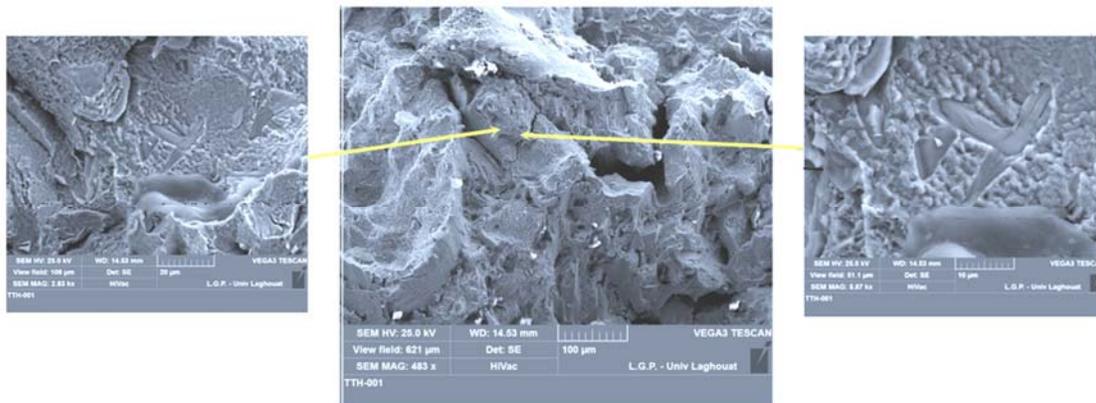


Fig. 15. SEM observations of a fracture faces of the specimen post-welding heat treated Ni superalloy

4. CONCLUSIONS

Through this investigation, the structural and mechanical properties of the welding of the INC 738LC superalloy with pre-post welding heat treatments were explored; the results can be briefly summarized as follows:

- The hardness of the as-received cast INC738 LC Ni superalloy was 398.17 Hv and the coarse γ' precipitates volume fraction was about 34 %.
- The preheating of the cast INC 738 LC Ni superalloy allows its hardness to decrease to a low value of about 351.44; however the volume fraction was 23% as volume fraction of coarse γ'

precipitates.

- During welding; when the cast Ni superalloy (base metal) is exposed to the high temperature of the TIG welding torch, in the HAZ at the base metal, the hardness of the welded Ni superalloy is allowed to change from 359.44 Hv to 400 Hv, but the volume fraction of coarse γ' precipitates in the base metal was 24%. In the FZ, the hardness was 410 Hv.
- The hardness of the post-welded heat-treated cast Ni superalloy (base metal) increases to reach the value of 450 Hv, 44% being the volume fraction of coarse γ' precipitates. In the FZ, the hardness jumps to 410 Hv.

- The ultimate tensile strength and Young's modulus of the heat-treated welded specimens were similar to those of the as-received metal.
- The welding of IN738 LC Ni superalloy free of cracks needs preheating and post heat treatments to improve the mechanical properties and life service of the blades, and this layout may be used to repair the scrap turbine blades.

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