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Procedia Engineering 64 (2013) 456 – 466

Procedia Engineering

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International Conference On DESIGN AND MANUFACTURING, IConDM 2013

Comment [S1]: Elsevier to update with volume and page numbers.

Influence of microstructural changes on impact toughness of weldment and base metal of duplex stainless steel AISI 2205 for low temperature applications

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Abstract

Austenite (γ) reformation and microstructural changes are the major anxieties in duplex stainless steel welding and aging process. Insufficient stabilization of austenite phases and intermetallic formations due to unfavourable thermal cycle leads to drastic reduction in the ductility and toughness of duplex stainless steel in particularly at low temperature. In this work, an attempt has been made to analyze the microstructure in the DSS weld, heat affected zone and base metal with respect to their impact toughness. DSS weld joints were fabricated using gas tungsten arc welding process with controlled welding parameters. Ferrite austenite ratio in the weld zone, heat affected zone and base metal was assessed by quantitative metallographic image analysis. The impact test results were correlated with the fractured surface and the microstructure of the tested specimens. The effect of heat treatment on the microstructural changes in the weld and base metal were also investigated with respect to impact toughness. Austenite phases were nucleated in the high temperature heat affected zone during heat treatment of weldment at 1050°C for 1 hour and it leads to enhancement in the impact toughness of the DSS weldment. But, drastic reduction in the impact toughness was observed in the base metal after heat treatment at 850°C and 1050°C due to the formation of sigma phase at 850°C and the coarser ferrite and austenite grains and partially dissolved sigma phase in the microstructure of 1050°C heat treated samples.

Selection and peer-review under responsibility of the organizing and review committee of IConDM 2013

Keywords: Duplex stainless steel; post weld heat treatment; ferrite; austenite; impact toughness

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Selection and peer-review under responsibility of the organizing and review committee of IConDM 2013
doi:10.1016/j.proeng.2013.09.119

1. Introduction

Duplex Stainless Steel (DSS) exhibits good toughness than the ferritic stainless steel grades within the temperature range of - 40°C to 300°C. It is better alternative for austenitic grades (316L and 304L) for high strength applications. The base metal microstructure of DSS usually consist approximately 50:50 ratio of ferrite and austenite phases in its matrix. At very low temperature like – 40ºC, ferrite phase becomes brittle and leads to loss in ductility. Greater decrease in impact toughness may happen in the base and weld metals of DSS when the environmental temperature is decreased [1-3]. And for the service temperature of above 300ºC, DSS nucleates intermetallic phases such as Cr_2N, sigma and chi phases in its grain boundaries which lead to embrittlement and causes severe reduction in impact toughness. Fusion welding is an important fabrication process that DSS should undergo to put this material in service against high corrosive and high strength applications such as offshore concrete structures, oil and gas pipe lines, chemical tankers in ships, desalination plants and ocean mining machinery etc. Rapid thermal cycle during welding of DSS leads to significant reduction in the impact toughness of DSS weld when compared with its base metal. Cooling rate, microstructural changes, residual stress formation and the segregation of alloying elements plays a major role in the impact toughness of DSS weld. Lower arc energy welding processes (LBW, EBW) imposes lesser amount of austenite precipitation due to faster cooling rate and higher arc energy welding processes (GTAW, GMAW, SMAW, FCAW) promotes larger amount of austenite phases in the weldment due to slower cooling rate. In these high arc energy welding processes, GTAW provides excellent efficiency in the weld joint [4 - 7]. Intermetallic formation is the major issue in DSS due to highly alloyed content in its chemical composition. Even very less amount of sigma formation in the DSS leads to severe reduction in impact toughness. Researchers proved that even 1% of sigma phase formation in DSS is more sufficient to cause embrittlement. Coarse grained ferrite structure forms near the fusion line also the reason for reduction in impact toughness of DSS weld [8 - 10]. Various researches carried out on the DSS weld shows that the austenite stabilizing elements like nickel and nitrogen leads to increase in impact toughness. Because, addition of nickel and nitrogen stabilizes more amount of reformed austenite phases also ferrite to austenite transformation commences at higher temperature [11, 12]. In general heat treatment at a lower temperature range from 400 to 1000 °C is not recommended for DSS due to detrimental effects caused by the intermetallic phases. But treating the material in the temperature of above 1000°C may leads to some improvement in the weld microstructure of DSS. In this work attempt has been made to analyze the microstructure of the DSS weldment, HAZ and base metal before and after the heat treatment and correlated with its impact properties. Also the effect of sigma phase nucleation on impact toughness of DSS was investigated.

2. Material and experimental procedures

The chemical composition of DSS AISI 2205 and its filler metal ER 2209 used in this experiment are shown in Table. 1. GTA welding was carried out on DSS plate of dimensions 150×140×8 mm, with a bevel angle of 60° in the faint surfaces of the plates. The polarity used in welding was the Direct Current Electrode Negative (DCEN). ER 2209 filler wire with a diameter of 2.5 mm was used in welding. Radiography Testing (RT) was carried out after welding and it shows that the joints are free from weld defects. The microstructural analysis in the weldment and HAZ were carried out using light optical microscopy. The samples were electrolytic etched using 10% NaOH solution. After welding Vickers hardness test was performed in the polished samples prepared from the weld. Hardness values were measured with 25g loading in three locations and average values were taken for each weld, HAZ and base metal.

Table 1. Chemical composition of base material (AISI 2205) and filler material (ER 2209)

	C	Mn	Si	S	P	Cr	Ni	Mo	Cu	N	Ti	V	Co	Nu	W	Fe
AISI 2205	0.027	1.463	0.42	0.01	0.02	22.8	5.5	3.3	0.1	0.18	0.004	0.06	0.010	0.02	0.04	65.9
ER 2209	0.009	1.50	0.38	0.0005	0.018	22.89	8.66	3.03	-	0.15	-	-	-	-	-	63.36

The welding parameters used in this work are listed in Table 2. Austenite is known to be stable in the temperature range from 1000ºC to 1200ºC. Therefore, Post Weld Heat Treatment (PWHT) was carried out at 1050ºC for the duration of 1 hour to stabilize the austenite phases in the weldment. Also in base metal, heat treatment was carried out at 850°C and 1050°C. Thyristor controlled programmable furnace was used for heat treatment. It took 50 minutes and 1 hour to reach the temperature of 850 and 1050°C respectively. To study the impact behaviour, the samples were prepared with the dimensions of 7.5 × 10 × 55 mm from the heat treated and non-heat treated weld and base metal samples. Impact test was carried out at - 40ºC using impact tester IT 30 as per ASTM standards. Test samples were brought to – 40°C using dry ice. The fractured surface of the impact specimens were captured using high resolution camera and compared with the observed microstructure.

Table 2. Welding parameters

Current (I) (amps)	125
Voltage (V) (volts)	11.6
No of passes	3
Average Welding Speed (U) (mm/sec)	0.523
Arc Energy (Q) (kJ/mm)	2.77
Heat input (kJ/mm) (60% of Arc energy for GTAW)	1.66
Interpass Temperature	150°C to 200°C
Allowable arc energy for Duplex Stainless Steel	0.5 to 2.5 kJ/mm
Shielding gas	99.9% pure argon gas

3. Results and discussion

3.1. Microstructural observations

The base metal microstructure reveals that the austenite (γ) phases are embedded in the ferrite (α) matrix as shown in Fig. 1. The austenite ferrite ratio measured in DSS base metal using quantitative metallographic image analysis shows approximately 50:50. The microstructure obtained from the DSS weld gives entirely different grain structure in weld and HAZ, when compared with the base metal. The evolution of the microstructure in the DSS weld zone has been taken place in three stages after welding. First the microstructure nucleates as allotriomorphs at the ferrite grain boundaries. Due to multipass welding, the weldment subjected to reheating, can result in widmanstätten side plates (needle like structured grains) that grow into the ferrite grains from the grain boundary allotriomorphs, and also as intragranular precipitates inside the ferrite grains. All the three forms of austenite phase i.e. grain boundary allotriomorphs, widmanstätten structure of austenite and intragranular austenite particles were observed in the weldment. And very less amount of intragranular austenite particles was observed. Because intragranular austenite particles require more driving force to nucleate inside the grains. There are no precipitations of intermetallic phases such as Cr_2N formation in weldment and HAZ was observed in this analysis. The microstructure of the weldment shows excessive formation secondary austenite phases in the form of widmanstätten structure in the root of the weld due to reheating the weldment during multipass welding as shown in Fig.2 and 3. In some locations of weld root more than 80% of austenite phases were observed. During welding, the zone near to fusion line approaches the melting point nearly to the temperature of 1450ºC, and becomes fully ferritic on heating. During cooling cycle, reformation of austenite phases was not sufficient in this zone to satisfy the duplex criteria due to rapid cooling achieved. Only the grain boundary allotriomorphs was observed in this zone as shown in Fig.4.This region is known as HTHAZ or overheating zone. HTHAZ gives ferrite levels in the range of 75 to 80%. Higher amount of ferrite precipitation leads to an excessive hardness and embrittlement in particularly at low temperature. Fusion line of the weld is shown in Fig. 5 which shows both weldment and HTHAZ are having with different ferrite austenite ratio. The austenite percentage in the LTHAZ was increased

10%, when compared with base metal. The measured value of austenite ferrite ratio in LTHAZ is around 60:40. There was no observation of intermetallic sigma (σ) phases in LTHAZ which is shown in Fig. 6. During welding, LTHAZ attains the temperature range of 800 to 1100ºC, which may stabilize sigma phase in the microstructure. But due to short period of exposure time, this temperature does not have any significant effect regarding sigma phase nucleation.

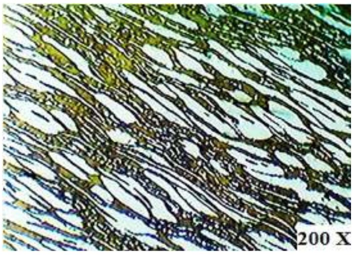

Fig. 1. Base metal

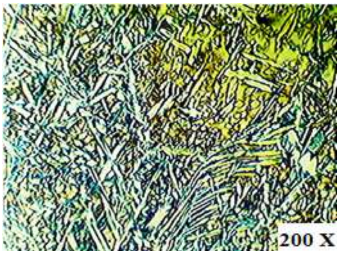

Fig. 2. Fusion zone location 1

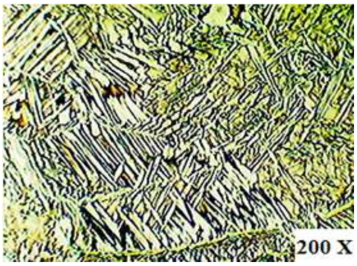

Fig. 3 Fusion zone location 2

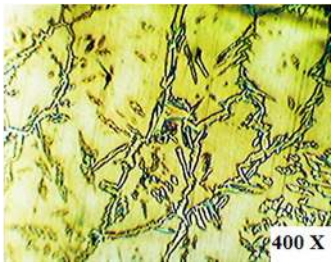

Fig. 4 HTHAZ (Over heating zone)

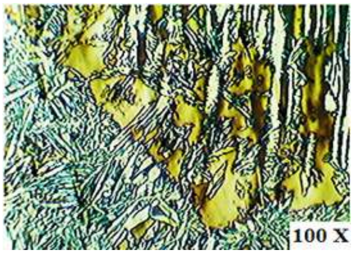

Fig. 5 Fusion line

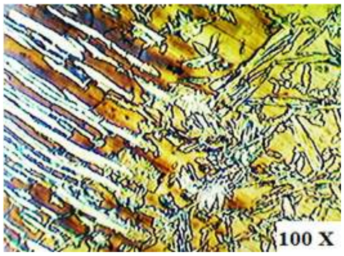

Fig. 6 LTHAZ (partially annealed zone)

3.2. Micro hardness analysis

The measured hardness values at different locations of the polished sample are shown in Fig. 7. In base metal hardness values are greater in ferrite phase than the austenite phase. This is due to high chromium and molybdenum content in the ferrite phase. The hardness of the duplex weldment is higher than that of the base material and HAZ due to strain induced heating and cooling cycle and also due to changes in microstructure such as secondary austenite formation. The strain induced hardening is caused by the compression of the weld region during solidification. It was observed that in some of the locations, the hardness of the austenite phases was higher than the ferrite phase in the weldment. This is due to the formation of secondary austenite phases which usually has high chromium and molybdenum content due to multipass welding which leads to the growth of austenite phases. And HTHAZ gives more hardness due to coarser ferrite grains. The measurement of micro hardness in the austenite phase of HTHAZ is not possible due to very thin grain boundary austenite phases. There was no significant variation in the hardness between LTHAZ and base metal was observed.

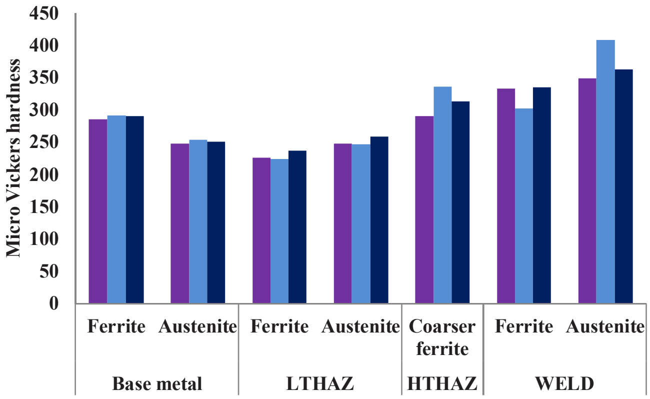

Fig. 7. Vickers Micro hardness test (HV 0.25) in DSS weld, HAZ and Base metal

3.3. Charpy impact test

The impact test results show that the base metal samples of DSS gives excellent toughness by absorbing average of 297 Joules at room temperature. Also it was found that there is no significant reduction in the toughness of the base metal at – 40ºC which is nearly in the average of 288 Joules. The absorbed impact values at room temperature and – 40ºC are shown in Fig. 8. This implies that the presence of austenite phases in the base metal provides excellent impact energy under perfectly alloyed conditions though the ferrite phase gets embrittlement in low temperature. There was a metal flow in the base metal due to its ductile nature during the toughness test as shown in Fig. 9 (a). This kind of metal flow behavior was not observed in the tested weld samples, which reveals almost a brittle fracture as shown in Fig. 9 (b). There was significant reduction in the toughness of the weld zone when compared with the base metal. The absorbed toughness values are shown in Fig. 10 (a). Nearly 40% of its toughness gets reduced in the weld zone. Formation of the constitutional elements like harder secondary austenite phases leads to ductile brittle transition at low temperature. Also at – 40ºC, ferrite structure in a DSS weld fully behaves like a brittle structure. Coarser ferrite grains near the fusion boundary and the formation of residual stresses during welding also the reason for getting low toughness in the tested weld samples. Also it was observed that the weldment of DSS gives uneven values of toughness in the tested samples. The fractured surface of the weldment shows large dimpled brittle fracture as shown in Fig. 10 (b). This is due to the formation of different forms of microstructure and also due to uneven segregation of alloying elements during solid state transformation. This kind of uneven toughness at different locations was rectified after PWHT by holding the material at 1050ºC for the duration of 1 hour followed by water quenching which leads to segregation of substitutional elements into proper phases. It was found out that this solution treatment causes noticeable improvement in toughness at – 40°C in the weldment. This is not only due to the increase in austenite phases in addition due to the release of weld induced residual stresses and dissolving of embrittling phases such as carbides and Cr_2N etc. Muthupandi et al (2003) also stated the improvement in impact toughness of duplex weldment due to solution annealing at a temp of 1050ºC for 30 min after welding. Similar range of toughness values were absorbed in all the tested specimens after PWHT which is shown in Fig. 11. (a). The fractured weld surface after PWHT reveals ductile fractured surface as shown in Fig.11 (b). In this work, an observed location of microstructure does not show any intermetallic phase precipitations. However, very small amount of secondary precipitations like Cr_2N can be possible in the weldment that can also be dissolved during PWHT. Also after heat treatment austenite phases were precipitated in the weld and HTHAZ as shown in Fig. 12 (a) and (b) which had coarser ferrite grains before the heat treatment.

The impact toughness test results for the heat treated base metal samples give very low amount of impact energy. Grain size was increased in the DSS base metal during the heat treatment at 1050°C as shown in Fig. 13 (b). Two cooling conditions were adopted after heat treatment i.e. air cooling and water quenching. Water quenched base metal samples gave slightly higher impact energy than the air cooled one due to the absence of repeated thermal cycle during cooling. There was an observation of intermetallic sigma phase in the DSS base metal after heat treatment in air cooled sample which is shown in Fig. 13. (a). The average value of impact energy from the heat treated base metal samples and fractured surfaces are shown in Fig. 14 and 15. Heejoon Hwang and Yongsoo Park [13] reported that the intermetallics precipitated in the microstructure were not dissolved into matrix completely during the treatment temperature of 1050ºC. The grain size of both ferrite and austenite phases increased and there by number of the grains reduced which may leads to reduction in impact toughness. The same observation was found in the present analysis.

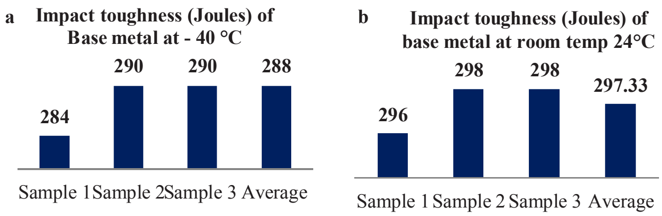

Fig. 8. Impact toughness of base metal (a) At – 40°C; (b) At room temperature

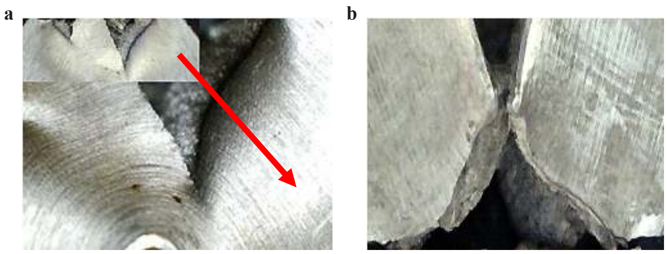

Fig. 9. (a) Metal flow in the DSS base metal; (b) Cleavage fracture in DSS weld

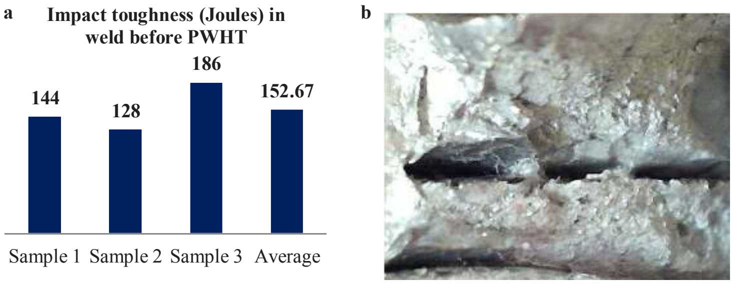

Fig. 10. (a) Impact toughness of weldment before PWHT; (b) Fractured surface

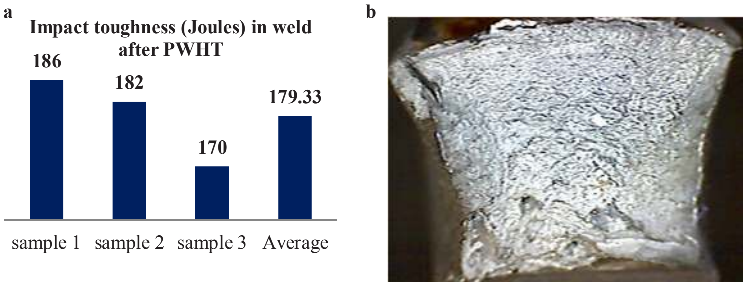

Fig. 11. (a) Impact toughness of weldment after PWHT; (b) Fractured surface

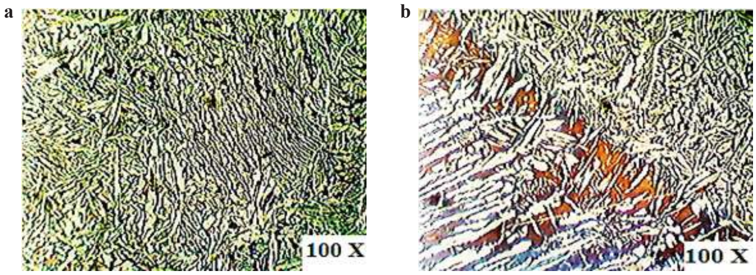

Fig. 12. (a) Microstructure of the weldment; (b) HAZ after PWHT

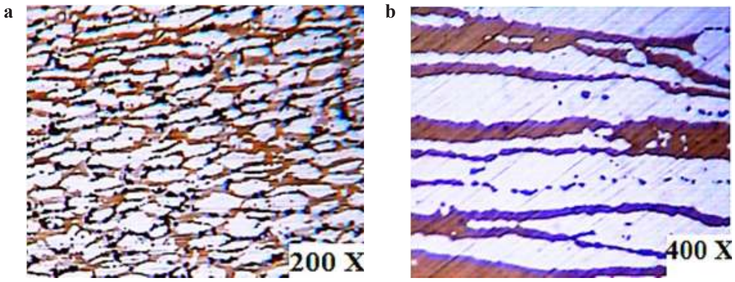

Fig. 13. Microstructure of the base metal after heat treatment at 1050°C
(a) Air cooled specimen (transverse direction); (b) Water quenched specimen (rolling direction)

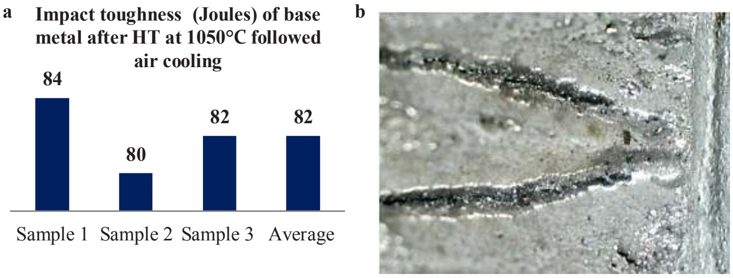

Fig. 14. (a) Impact toughness of base metal after HT at 1050°C followed by air cooling; (b) Fractured surface

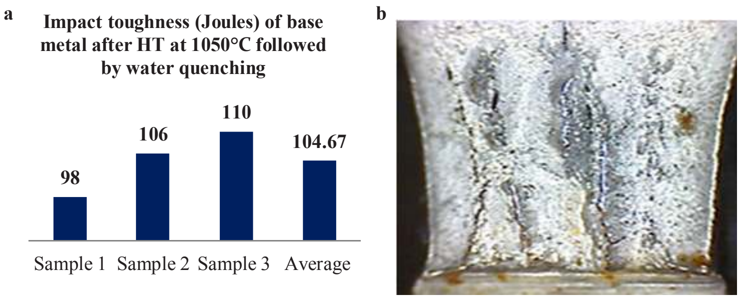

Fig. 15. (a) Impact toughness of base metal after HT at 1050°C followed by water quenching; (b) Fractured surface

3.4 Effect of sigma phase formation

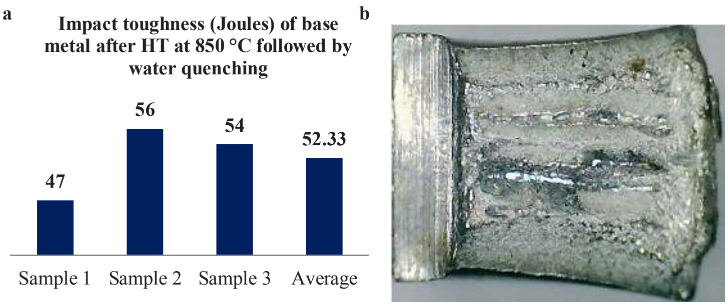

Fig. 16. a) Impact toughness of base metal after HT at 850°C b) Fractured Surface

Sigma phase nucleation causes embrittlement in its microstructure which leads to severe reduction in the impact toughness of DSS base metal. The average toughness value absorbed in the tested sample was 52.33 Joules which is far less than the toughness of the base metal which is free from sigma in the as received condition. The absorbed energy and fractured surface of the sigma contained sample is shown in Fig. 16. (a) and (b). EDX analysis was carried out in the sigma phase and the nearer locations. The measured locations in EDX analysis are indicated in Fig. 17 (a) and (b). The sigma phase contains the excessive accumulation of chromium and molybdenum atoms in it which causes depletion in the surrounding regions which is shown in Fig. 18. The segregation of major alloying elements in the sigma phase and nearer ferrite austenite phases is shown in Table. 3.

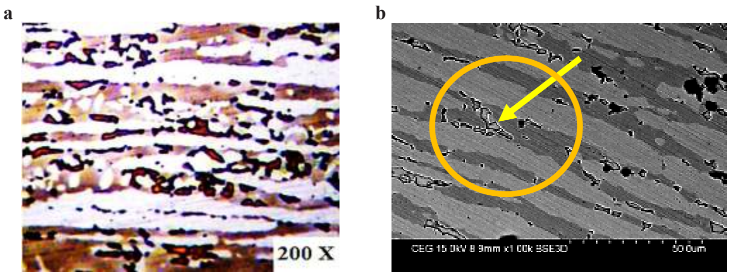

Fig. 17. (a) Base metal of DSS with sigma precipitation; (b) SEM image

Table. 3. The EDAX analysis of sigma phase

Element	Percentage amount in different phases					
	Sigma (σ)		Ferrite (α)		Austenite (γ)	
	Region 1	Region 2	Region 1	Region 2	Region 1	Region 2
Cr	25.61	22.90	23.34	22.71	19.94	20.01
Mo	7.06	8.66	2.20	2.54	1.65	1.88
Mn	1.02	1.26	1.27	1.13	0.90	1.33
Ni	2.58	3.80	3.32	3.18	6.47	6.26
Si	0.48	0.59	0.49	0.26	0.52	0.31
Fe	56.22	55.04	69.38	66.61	68.21	67.17

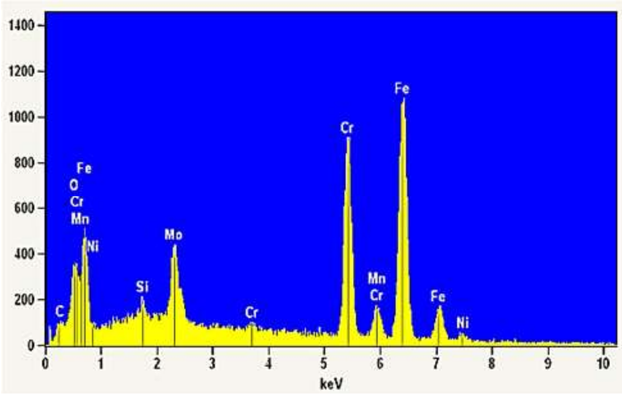

Fig. 18 EDAX analysis

4. Conclusions

In this work an experiment has been carried out on impact toughness of DSS base metal and weldment before and after heat treatment. Based on the observations from the experiment, the following conclusions were arrived.

- Microstructure of DSS weld reveals three different forms of austenite phases i.e. grain boundary allotriomorphs, widmanstätten structure and intragranular austenite particles.

- DSS weldment exhibits excessive hardness than the base metal and HAZ. Also in some of the measured locations in DSS weld, the hardness of the austenite phases is greater than the ferrite phases due to the formation of harder chromium rich secondary austenite phases.

- Base metal of DSS exhibits good toughness even at low temperature i.e. - 40ºC. However the toughness of DSS weld has decreased significantly due to the formation of harder secondary austenite phases, uneven segregation of alloying elements, and formation of coarser ferrite grains near the fusion line.

- Heat treatment of DSS weld at 1050°C causes significant enhancement in the impact toughness by promoting austenite phases in the HTHAZ. However, drastic reduction in the impact energy was absorbed in base metal after heat treatment due to grain growth. Heat treatment at a temperature of 1050°C does not promote austenite phases in the base metal of DSS. Also during the heat treatment, sigma phase was nucleated at 850°C which is not dissolved completely in the solid solution during the heat treatment process at 1050°C.

- Heat treatment at a temperature of 850°C causes the nucleation of sigma phases in the grain boundaries which leads to rigorous reduction in the impact toughness due to embrittlement.

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