Original Article

Friction welding of a nickel free high nitrogen steel: influence of forge force on microstructure, mechanical properties and pitting corrosion resistance

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ABSTRACT

In the present work, nickel free high nitrogen austenitic stainless steel specimens were joined by continuous drive friction welding process by varying the amount of forge (upsetting) force and keeping other friction welding parameters such as friction force, burn-off, upset time and speed of rotation as constant at appropriate levels. The joint characterization studies include microstructural examination and evaluation of mechanical (micro-hardness, impact toughness and tensile) and pitting corrosion behaviour. The integrity of the joint, as determined by the optical microscopy was very high and no crack and area of incomplete bonding were observed. Welds exhibited poor Charpy impact toughness than the parent material. Toughness for friction weld specimens decreased with increase in forge force. The tensile properties of all the welds were almost the same (irrespective of the value of the applied forge force) and inferior to those of the parent material. The joints failed in the weld region for all the weld specimens. Weldments exhibited lower pitting corrosion resistance than the parent material and the corrosion resistance of the weld specimens was found to decrease with increase in forge force.

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1. Introduction

The Ni saving high nitrogen steels (HNSs) are of four types: martensitic, ferritic, Cr–Mn–(Ni–N)/N austenitic and duplex. It is noteworthy that Cr–Mn–(Ni–N)/N grade of HNS has tremendous potential for substitution for the widely used type 304 ASS for household products, automobile parts and materials for the chemical and food industries, mainly due to their outstanding combination of strength and ductility, excellent work-hardening capability and corrosion resistance [1,2].

Weldability is an important issue for the wider expansion in the use of HNSs for structural purpose. Fusion welding of HNS has several limitations such as N2 pores and/or precipitation of Cr-nitrides in the heat-affected zone (HAZ), depending on the welding parameters like welding voltage and current, shielding gas composition and level, type of filler metal, type of flux, thermal cycle, etc., employed [3–5]. All those phenomena degrade the local mechanical and corrosion resistance of the weld to a great extent. Desorption of nitrogen has been found to be the governing mechanism for N2 flux transfer in and out of the base material in case of nitrogen containing ASS
Table 1 – Chemical composition of the base material.

<table>
<thead>
<tr>
<th>Elements</th>
<th>C</th>
<th>Mn</th>
<th>Cr</th>
<th>S</th>
<th>P</th>
<th>Si</th>
<th>N</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt.%</td>
<td>0.076</td>
<td>19.780</td>
<td>17.960</td>
<td>0.007</td>
<td>0.051</td>
<td>0.340</td>
<td>0.543</td>
<td>Bal.</td>
</tr>
</tbody>
</table>

and thus controlling the final nitrogen content and porosity in the weld, as reported by Toit and Pistorius [6]. Therefore, one may opt for high base material nitrogen content, as suggested by the same researchers. However, it has been reported that the tendency for pore formation during welding goes up with increase in nitrogen content of the base material [7]. Researchers have proposed a solution towards this problem by increasing the solubility of nitrogen with the addition of manganese, especially for high chromium–molybdenum alloys [8]. Again, one potential limitation when welding highly alloyed ASS is hot cracking. As a measure to minimize the hot cracking risk, one needs to choose a filler material with low impurity levels (e.g., S, P) in addition to keeping eye on the least degree of segregation of the major alloying elements and minimization of the level of intermetallic phase in the weld metal [7]. Therefore, fusion welding is a challenging and non-reliable process to apply for the HNS type of material; as, here, extreme careful observation on the influence of welding parameters on joint properties and thus control and optimization of those are required.

Almost all the shortcomings of the fusion weld HNS joint can be wiped out by employing a solid state route. Among the solid state welding processes, friction welding and friction stir welding (FSW) are the two most popular ones. In case of FSW, lots of understanding needs to be there on the selection of proper tool material and only plate shaped materials can be joined together [9,10]. On the other hand, friction welding can be helpful for joining materials with varieties of shapes such as rod to rod, rod to plate etc., wherein, at least one component should be axi-symmetric. Among the friction welding processes, rotary friction welding is by far the most common form which accounts for most of the machines and their accessories in today’s industry [11]. This has two principle varieties, based on the availability of energy to weld – (i) direct drive or continuous drive friction welding, sometimes called conventional friction welding uses continuous input supplied by motor driven workpiece, (ii) inertia friction welding, sometimes called flywheel friction welding, uses energy stored in a flywheel. Continuous drive friction welding is a well established solid state joining process which can be used to join a wide range of conventional steels as well as more metallurgically challenging material systems such as superalloys and dissimilar material combinations [11,12].

However, to the best of our knowledge, there are very limited studies on friction welding of HNS [13,14], and that too, on low nitrogen containing HNS. In case of friction welding, upsetting (forge) force has been found to be one of the most important parameters among others such as friction force, friction time (or, burn-off), upsetting time and rotational speed of the moving component [12]. It is also appealing to note that there is very limited report on the corrosion studies of HNS friction welds. In view of the foregoing, in the present study, specimens were joined by varying amount of forge force, while keeping the other friction welding parameters such as burn-off, friction force, upsetting time and rotational speed as constant at appropriate levels and by using a continuous drive friction welding set up. The joint characterization studies include microstructural examination and evaluation of mechanical (micro-hardness, impact toughness and tensile) and pitting corrosion behaviour.

2. Experimental details

2.1. Materials

The base materials employed in this study are high nitrogen steels (rods of 15 mm diameter and 60 mm length) of which chemical composition and mechanical properties are given in Tables 1 and 2 respectively. The materials were available as plates in hot rolled (at 1150°C) condition. The microstructure of the base material is shown in Fig. 1. It consists of equiaxed austenite grains.

2.2. Welding details

A 150 kN capacity continuous drive friction welding machine was employed for the welding experiments. This is a stepless variable speed machine with rotational speed range of 0–2400 rpm. The main variable parameters are friction force, forge force and burn-off which are generally considered to be the variables controlling the quality in friction welding. Trial runs were conducted by changing one of the process variables and remaining others as constant. The working range of forge
force was explored by inspecting cross section for the presence of defect(s). Weld parameter regime was based on defect free welds. In this study, forge force ranged from 20 kN to 60 kN and friction force at 20 kN, burn-off at 5 mm and rotational speed at 1600 rpm were kept constant. Here, the length loss during friction stage is taken as the burn-off. The weld samples are designated (throughout this whole report) on the basis of the three welding parameters used to process them, namely friction force, forge force and burn-off. For example, designation X-Y-Z for a sample reveals the values for the used friction force (X), forge force (Y) and burn-off length (Z) respectively.

2.3. **Metallography**

Friction weld joint cross sections were subjected to standard metallographic sample preparation technique to examine the microstructure under LEITZ optical microscope. Aquazid (75 vol.% HCl + 25 vol.% HNO₃) was employed as etchant to reveal the microstructure of the base as well as weld samples in the etched condition. The specimens after corrosion test were examined under LEITZ optical microscope and LEO scanning electron microscope (equipped with OXFORD EDS analysis instrument) to obtain the idea on morphology and composition of the corroded products.

2.4. **Micro-hardness measurement**

Micro-hardness survey was carried out along and in transverse direction to the weld (bond) line. The ASTM E384-11e1 standard test method was followed during the test. The measurement was carried out with a Matsuzawa Digital Micro Hardness Tester (DMHT) with the standard Vickers indenter and 500 g load.

2.5. **Charpy V-notch impact test**

Impact test was carried out for the samples prepared from the welded joints with adoption of the ASTM E2248-12 standard test method. Specimens were sectioned from the weldment with specimen axis transverse to the weld joint and with notch location at the weld centre, as shown in Fig. 2(a). The notch was cut from the top surface. The location and orientation of the test sample in the bond zone is shown in Fig. 2(b). The tests were carried out in a Tinius Olson (TO) machine at room temperature.

2.6. **Tensile test**

The tensile test specimens were sectioned from the weldment with specimen axis transverse to the weld joint and weld line at the centre of the gauge length using ASTM E8 code. Tests were carried out in an INSTRON (model no – 5500R) testing machine with 25 mm gauge length and 1 mm min⁻¹ cross head speed.

2.7. **Corrosion test**

The weld joints and the base material were tested for pitting corrosion resistance in an electrolyte of (0.5 M H₂SO₄ + 0.5 M NaCl). The electrochemical measurements were made using a potentiometer. Steady state potential was recorded for 10 minutes after exposure of the specimen into the electrolyte. The potential was raised anodically using scanning potentiostat at a scan rate 2 mV/s. The potential at which the current increases abruptly after the passive region was taken as pitting potential (Eₚᵣ). 

3. **Results and discussion**

3.1. **Macro and microstructure**

3.1.1. **Macrostructure**

3.1.1.1. Visual examinations. The macro images for friction weld joints, presented in Fig. 3, show that the flash diameter is increasing with increase in forge force. It indicates that the extent of plastic working of the material increases as the forge force to the weld is increased.

3.1.1.2. Weld quality. In all the welds examined, the integrity of the welds, as determined by optical microscopic and radiographic examination was very high. No cracks, areas of incomplete bonding or other flaws were observed.

3.1.2. **Microstructure**

Optical micrographs, shown in Fig. 3 reveal that the weld widths are narrower at higher forge force values. Fig. 4 gives an account on the width of the weld at centre and mid-centre regions along the bond line. Here, one can notice that at low forge force values, the width at the centre of the weld zone was higher than that at the mid-centre. However, the trend

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Fig. 2 – (a) Charpy ‘V’ notch impact specimen and (b) location and orientation of Charpy ‘V’ notch in the weld (bond) zone.
is reversed at 40 kN and higher loads. This narrower central region than the periphery with increase in forge force value is a result of more and more squeezing out of the material from the workpiece(s) [11,12]. Fig. 5 presents the micrographs at and in the close vicinity of the bond line for 20 kN, 40 kN and 60 kN forge forces. It shows the change in the shape of the deformation (flow) line with change in forge force value. It has been observed that the flow lines for 40 kN forge force value has the maximum visibility, as also clearly revealed here.

Microstructural evaluation of friction-welded joints revealed four distinct zones in transverse direction to the bond line, as shown in Fig. 5. Those regions are named as transformed, recrystallized and fully plasticized deformed zone (FPDZ), deformed zone (DZ), partially deformed zone (PDZ) and parent material (PM), marked as R-I to R-IV respectively. The etching contrast in Fig. 5 can be taken as criterion for recognition of different regions. The FPDZ is distinguished from all other regions by its whitish appearance. This is because of the fact that the region is not etched at all due to presence of very fine dynamically recrystallized grains.

The DZ appears as a thin blackish layer beside FPDZ. This is because of the fact that it experiences the highest degree of deformation (working) among all the regions along the length. The PDZ and PM gradually appear as lighter ones as one move from the DZ towards the parent material, with the same reasoning as applied to the DZ. In this case, the region consisting of FPDZ, DZ and PDZ may be termed as heataffected zone (HAZ). The width of each separate region and thus that of the HAZ seems to be controlled by the forge force. It should be noted that the boundary lines for DZ with FPDZ and PDZ are not very well defined. Thus, there may arise confusion on the area of span of DZ and also sometimes on its existence. Interestingly, the DZ does not seem to exist in the case of the weld structure for 40 kN forge force value. From visibility point of view, there are well defined boundaries for the FPDZ only. It was observed that grain size of friction welds was much smaller than that of the base material. This is because of the fact that the grains are spun and dynamically recrystallized repeatedly, thus resulting in very fine grains, especially along and in the close vicinity of the FPDZ. The most microstructural changes occur in the FPDZ and DZ. It is very difficult to analyse the microstructural variation in DZ and PDZ with variation of the forge force. This is due to the difficulty in analysing the effect of the strain rate (ε) and temperature attended (Tpeak) on the stress state in DZ and PDZ. This, in turn, is because of the continuous change of position for those regions and thus experiencing the different temperature (T = 0 to a peak temperature, Tpeak) and strain rate regimes at each instant (while traversing towards the bond line and opposite to the direction of frictional heat flow). The authors would like to pay a special attention on this issue in the upcoming studies.

Microstructures at the centre and the periphery along the bond line for weld bead configurations (20-20-5) and (20-60-5) are shown in Fig. 6(a) and (b) respectively. The structure at centre and periphery gets refined with increase in forge force in the present study. Microstructural features presented in Fig. 6(a) and (b) show coarser grains at the peripheral region as compared to those at the central region. We know that linear

Fig. 3 - Macrographs for the weld joints for different forge force values (the arrow marked images are optical microscopic images).

Fig. 4 - Variation of the weld width at central and mid-central regions of the bond line for different forge force values.
velocity ($v$) of any point within the cross-section of the joining parts (say, rods, as are here) is linked with the rod radius, $r$ and angular velocity of the moving rod, $\omega$ by $v = \omega r = 2\pi n r$, where $n =$ rotational speed (in rpm). Thus, the $v$ value remains zero at the centre of axes for both the specimens (called as dead-zone) at any point of time during friction welding process. It goes up while going away from the central point and towards the periphery. Therefore, more and more frictional rubbing and so the heat generation would be there in the peripheral regions as compared to those near to central zone, as one move away from the dead zone. This increment in temperature in the frictional stage increases the grain sizes of the peripheral regions as compared to those in central ones. Now, upon influence of the increased forge force and thus the strain rate, grain sizes for both the central and peripheral regions become finer. However, the magnitude of the grain refinement as a result of a specified increment in forge force is higher in case of peripheral grains as compared to that for grains at central region. This is clearly evident in Fig. 6(a) and (b). The major reason behind this happening is the availability of the coarser and softer grains in the peripheral regions as against somewhat finer and hard ones at the central part. It is clear from the discussion so far that the effect of change of forge force in case of friction welding is expected to be proved by the prominent visibility in the changed grain structure at the peripheral regions.

The XRD results, presented in Fig. 7 show the peaks for austenite ($\gamma$-Fe) in base as well as all the weld specimens. This reveals the high austenitic stability of the base material, even with high strain rate hot deformation in the present study. This is also well supported by the joint microstructure presented here and by the literature on the effect of nitrogen in austenite stabilization [1].

### 3.2. Mechanical properties

#### 3.2.1. Micro-hardness survey

The typical micro-hardness profiles along the bond line with least and highest forge forces are shown in Fig. 8. There is a

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**Fig. 5** - Deformation (flow) lines and bending of the material near to the bond line for different forge force values.

**Fig. 6** - Microstructure near to and away from the dead zone along the weld line for (a) least (20 kN) forge force and (b) highest (60 kN) forge force.
slight variation in hardness values for the different regions with change in forge force value. The highest hardness values for all the regions were obtained for the joint with highest forge force value. The variation of hardness values along the bond line shows a decreasing trend when moved from centre to periphery, for a particular forge force value. This is because of the existence of the coarser grains as one move from centre to the periphery, as already reported in the literature [15].

3.2.2. Impact toughness
Charpy impact energy data for the base material as well as for friction welds for varying forge force are shown in Fig. 9. An examination of the results indicates that the charpy energy decreases with increasing forge force values. This may be proved by a close observation of the crack path experienced by the specimens during impact loading, as shown in Fig. 10 and supported well by report by Rajashekar et al. [16]. The crack paths for different samples were observed by cutting the sample perpendicular to the notch at the centre. Maximum crack path length was noticed in case of base material. It was found that crack is highly tortuous in that case as compared to those in as welded conditions. This may be further proved by observing the fractographs for the base material and for welds with highest forge force value, as shown in Fig. 11. Here, one can observe very clear evidence of the reduced toughness in case of welded samples; though, the base material fractures in fairly ductile mode.

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**Fig. 7** – XRD peaks for (a) base material and (b) weld sample with highest (60 kN) forge force value.

**Fig. 8** – Micro-hardness variation along the bond line.

**Fig. 9** – Charpy impact toughness values for base material and weld joints for different applied forge forces.
3.2.3. Tensile properties
The friction weld joints possess poorer strength than the base material, as is evident through observation of Fig. 12 and Table 3. There was not much variation between the tensile strength of the joints, processed with different forge forces. The joints fail at the weld zone, as revealed in macrographs shown in Fig. 13. Fractography in Fig. 14 reveals more ductile mode of fracture in case of base material than that for the weld joints.

3.2.4. Pitting corrosion behaviour
Typical polarization curves for base material and weld joints for least and highest forge force values are presented in Fig. 15. Pitting potential ($E_{\text{pit}}$) was taken as the criterion for comparison of pitting corrosion resistance. Less positive $E_{\text{pit}}$ values imply lower resistance to pitting, and vice versa. Fig. 16 shows the pitting potentials for parent material and for weld joints

<table>
<thead>
<tr>
<th>Material/sample type (X-Y-Z)</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>% El</th>
<th>% RA</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base material</td>
<td>940</td>
<td>1160</td>
<td>39</td>
<td>69</td>
</tr>
<tr>
<td>Sample 20-20-5</td>
<td>730</td>
<td>987</td>
<td>21</td>
<td>47</td>
</tr>
<tr>
<td>Sample 20-30-5</td>
<td>765</td>
<td>1005</td>
<td>23</td>
<td>41</td>
</tr>
<tr>
<td>Sample 20-40-5</td>
<td>757</td>
<td>988</td>
<td>23</td>
<td>43</td>
</tr>
<tr>
<td>Sample 20-50-5</td>
<td>835</td>
<td>1033</td>
<td>23</td>
<td>47</td>
</tr>
<tr>
<td>Sample 20-60-5</td>
<td>783</td>
<td>1036</td>
<td>30</td>
<td>43</td>
</tr>
</tbody>
</table>

X, friction force; Y, forge force; Z, burn-off length.

Fig. 10 – Macrographs, crack path for the base material and friction weld joints after failure under impact loading.

Fig. 11 – Fractographs for the base material and friction weld joints after failure under impact loading.

Fig. 12 – Ultimate tensile strength values for base material and weld joints for different applied forge forces.
Fig. 13 – Macrographs for the base material and friction weld joint for highest forge force value after failure under tensile loading.

Fig. 14 – Fractographs for the base material and friction weld joint for highest forge force value after failure under tensile loading.

for various forge force values. Pitting resistance of friction weldments was found to be much lower than that of the parent material, as also reported in the literature in case of AISI 304 austenitic stainless steel [17]. In general, localized corrosion resistance of welds is affected by the microstructural changes and, in the present study, the same is clearly evident from the preferential pitting along the weld (deformation) lines for all the welded samples, as shown in Fig. 17. It is noteworthy

Fig. 15 – Typical polarization curves for the base material and friction weld joints for least (20-20-5) and highest (20-60-5) forge force values.

Fig. 16 – Pitting potential (E_{pit}) values for the base material and friction weld joints for different forge force values.
that the severity of preferential corrosion attack (visually, the thickening of the lines by the presence of corroded products) is higher with increase in forge force. The reason for this is the presence of more and more worked (strained) structure with higher forge force values. It indicates that corrosion resistance of the weld joint decreased with increase in the forge force. This is further supported by the SEM images, as presented in Fig. 18 and EDS results on corroded surfaces, given in Table 4. It should be noted that the formation of Fe-, Cr- and Mn-rich oxides at the weld zone is enhanced with increase in forge force. Barbucci et al. [18] has shown that the higher passive currents and increased propensity to pitting corrosion of the work hardened samples may be explained by the formation of much more defective oxides during its anodic oxidation, with easy paths that enhanced sulphate ingress. The growth of such oxides was related to the formation of defects in the grains and more defective interfaces in the bulk material, resulting from the build up of internal stresses during cold rolling. Therefore, similar type of things may be expected in our present study on friction welding type of metal working.

In the present study, the microstructural change is attributable to high strain rates and recrystallization. It is likely that both introduction of high strain and dynamic recrystallization can cause microstructural changes in the nugget zone. Relatively higher magnitudes of plastic strains are induced in friction welds due to severe plastic deformation. These plastic strains may also contribute to strain induced martensite formation in the nugget zone. A similar phenomenon is noticed in stainless steels by increasing cold working above 23% [19]. In that case, number of active anodic

<table>
<thead>
<tr>
<th>Elements</th>
<th>wt.%</th>
<th>at.%</th>
</tr>
</thead>
<tbody>
<tr>
<td>(a) For corroded surface of base material</td>
<td></td>
<td></td>
</tr>
<tr>
<td>CK</td>
<td>2.58</td>
<td>10.07</td>
</tr>
<tr>
<td>OK</td>
<td>3.05</td>
<td>8.54</td>
</tr>
<tr>
<td>SiK</td>
<td>0.37</td>
<td>0.62</td>
</tr>
<tr>
<td>CrK</td>
<td>17.97</td>
<td>16.42</td>
</tr>
<tr>
<td>MnK</td>
<td>18.96</td>
<td>16.40</td>
</tr>
<tr>
<td>FeK</td>
<td>57.07</td>
<td>47.95</td>
</tr>
<tr>
<td>Total</td>
<td>100</td>
<td>100</td>
</tr>
<tr>
<td>(b) For corroded surface of 20-20-5 configuration</td>
<td></td>
<td></td>
</tr>
<tr>
<td>CK</td>
<td>4.81</td>
<td>14.79</td>
</tr>
<tr>
<td>SiK</td>
<td>0.21</td>
<td>0.11</td>
</tr>
<tr>
<td>CrK</td>
<td>18.11</td>
<td>17.13</td>
</tr>
<tr>
<td>MnK</td>
<td>18.96</td>
<td>16.97</td>
</tr>
<tr>
<td>FeK</td>
<td>57.91</td>
<td>51.00</td>
</tr>
<tr>
<td>Total</td>
<td>100</td>
<td>100</td>
</tr>
<tr>
<td>(c) For corroded surface of 20-60-5 configuration</td>
<td></td>
<td></td>
</tr>
<tr>
<td>CK</td>
<td>5.29</td>
<td>16.09</td>
</tr>
<tr>
<td>SiK</td>
<td>0.54</td>
<td>0.94</td>
</tr>
<tr>
<td>CrK</td>
<td>19.05</td>
<td>17.27</td>
</tr>
<tr>
<td>MnK</td>
<td>18.60</td>
<td>16.47</td>
</tr>
<tr>
<td>FeK</td>
<td>56.52</td>
<td>49.23</td>
</tr>
<tr>
<td>Total</td>
<td>100</td>
<td>100</td>
</tr>
</tbody>
</table>
sites in the system is increased and thus results in severe localized pitting corrosion.

In addition to the above, the localized attack in stirred (deformed) zone can also be explained as follows. Harmful anions, most notably the Cl\textsuperscript{−} ion, have been shown to cause chemical breakdown of passive oxide films on stainless steels, as reported by Pickering and Frankenthal [20]. In stirred zone, internal stresses, often approaching the yield strength, may be produced [21]. Anions will migrate to stress gradients, which results in localized regions of high anion concentrations and in saline electrolytes, electrolysis reactions from corrosion can produce localized regions. Passivation breakdown in association with lack of spontaneous repassivation in the presence of such electrolytes promotes accelerated localized attack [22]. This is in agreement with the present results on pitting corrosion of friction welds of high nitrogen steel. In view of the above, it is clearly understood that friction welding of HNS causes poor corrosion resistance of nugget zone.

- Welds show poorer tensile strength than the base material and the forge force does not have an effect on the tensile strength of the welds.
- Pitting corrosion resistance of friction weldments is much lower than that of base material. It decreases with an increase in forge force.

**Conflicts of interest**

There is no conflict of interest to declare.

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