Special issue: Welding

The welding consequences of replacing austenitic with duplex stainless steel

Fracture toughness of welded commercial lean duplex stainless steels

Abbreviations

BM  Base material
CPT  Critical pitting temperature
CMOD  Crack mouth opening displacement
CTOD  Crack tip opening displacement
EBW  Electron beam welding
FCAW  Flux core arc welding
GMAW, see also MAG  Gas metal arc welding
GTAW, see also TIG  Gas tungsten arc welding
HAZ  Heat affected zone
LBW  Laser beam welding
MAG, see also GMAW  Metal active gas
MIG  Metal inert gas
MMA, see also SMAW  Manual metal arc
PAW  Plasma arc welding
PWHT  Post weld heat treatment
SAW  Submerged arc welding
SMAW, see also MMA  Shielded metal arc welding
TIG, see also GTAW  Tungsten inert gas
W/D  Width/depth
WPS  Welding procedure specification

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The welding consequences of replacing austenitic with duplex stainless steel

Björn Holmberg, Outokumpu Stainless, Sweden
Mats Liljas, Outokumpu Stainless, Sweden
Fredrik Hägg, Avesta Welding, Sweden

Abstract

A change of construction material will always have consequences for the fabrication. This is also the case when replacing an austenitic stainless steel with a duplex stainless steel. The duplex stainless steels have higher strength, which might affect forming and machining in the workshop. However, the largest impact of such a change would be on the welding procedures. The weldability of duplex stainless steels is good but their welding characteristics deviate in many cases from those of austenitic grades. To be successful in the change of material, involved personnel must be informed about and qualified for the new material. New welding procedures must also qualified. It is recommended that the qualification of the proposed and performed welding procedures shall be in accordance with the demands for the real application. By fulfilling all these requirements, unexpected deviations and failures can be avoided. This paper will give typical field experiences on the differences between welding austenitic and duplex stainless steels. By using the guidelines in EN ISO 3834-2, welding procedure specifications, system for non-destructive testing and post weld handling can be controlled.

Keywords: Duplex stainless steel, welding, field experience, weld metal properties, quality system

1. Introduction

The modern duplex stainless steels represent an excellent combination of good corrosion resistance and high strength. Another very important driving force to change from austenitic stainless steels to the modern duplex stainless steels has, in recent years, been the strongly increased raw material prices e.g. nickel and molybdenum. Due to the chemical composition of the steel, the products can be produced and supplied at a far more stable price level compared to the austenitic grades.

The intention with this paper is to give some fabricator guidelines when shifting from standard austenitic to duplex stainless steels. The paper will show typical examples from the field where the change of material has caused unexpected problems and finally how a fabricator can avoid unexpected phenomena by using the new standardized method “Quality requirements for fusion welding of metallic materials” EN ISO 3834-2.

The chemical composition and the typical yield strength of hot rolled coils of different modern duplex steels are given in Table 1. Note that the strength values are also dependent on the product form: cold rolled coils normally have higher strength levels.
2. Guidelines to fabricate in duplex steels

Some general guidelines for the fabrication of duplex stainless steels are given below. The guidelines are mainly described in relation to austenitic stainless steels, which still are the dominating steel type.

2.1 Cold forming and machining

Duplex steels have higher yield strengths compared to austenitic steels, which implies that the forming tools must be designed for greater working forces. The springback effect is also higher. Due to somewhat lower fracture elongation compared to austenitic steels, a very sharp bend radius should be avoided. Duplex steels are also in general more difficult to machine than conventional austenitic steels. LDX 2101® is however an exception with excellent machinability [1].

2.2 Joint preparation

All common joint preparation methods used for stainless steels can be applied for duplex steels. If thermal cutting is used, remaining oxides shall be removed before welding for both types of steel.

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### Table 1

**Typical composition and yield strength values of some duplex grades (hot rolled plate).**

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>ASTM</th>
<th>EN</th>
<th>Chemical composition (weight per cent)</th>
<th>Yield Strength (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>C max</td>
<td>N</td>
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<tr>
<td>LDX 2101®</td>
<td>S32101</td>
<td>1.4162</td>
<td>0.03</td>
<td>0.22</td>
</tr>
<tr>
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<td>1.4362</td>
<td>0.02</td>
<td>0.10</td>
</tr>
<tr>
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<td>1.4462</td>
<td>0.02</td>
<td>0.17</td>
</tr>
<tr>
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<td>S32750</td>
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<td>0.27</td>
</tr>
<tr>
<td>4401</td>
<td>316</td>
<td>1.4401</td>
<td>0.04</td>
<td>0.04</td>
</tr>
</tbody>
</table>

**LDX 2101® trademark owned by Outokumpu Stainless**

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### Table 2

**Chemical composition of Avesta Welding’s matching duplex filler.**

<table>
<thead>
<tr>
<th>Avesta Designation</th>
<th>EN</th>
<th>AWS</th>
<th>C</th>
<th>N</th>
<th>Cr</th>
<th>Ni</th>
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<td>EN 1600</td>
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<td>24.5</td>
<td>9.0</td>
<td>&lt; 0.3</td>
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<td>9.5</td>
<td>3.0</td>
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<td>25 9 4 N L R</td>
<td>E2594</td>
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<tr>
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<td>–</td>
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<td>0.14</td>
<td>24.0</td>
<td>9.0</td>
<td>0.6</td>
</tr>
<tr>
<td>2205</td>
<td>22 9 3 N L</td>
<td>E2209</td>
<td>0.03</td>
<td>0.13</td>
<td>22.5</td>
<td>9.0</td>
<td>3.2</td>
</tr>
</tbody>
</table>

* MIG, TIG and SAW wire
2.3 Joint types
To facilitate good penetration and avoid slag inclusions and hot cracking, the joint angle shall be about 10° wider than when welding standard austenitics. The land shall generally be smaller, due to somewhat lower penetration. To obtain optimal mechanical and corrosion properties, filler should in most cases be used. In such cases the welding could be carried out with a gap between the plates.

2.4 Tacking
Compared to tacks in austenitic grades, the tacks shall be longer and they should have a shorter spacing. Tacking should be done with fillers that match the steel grade. If the rules for tacking austenitic grades are followed, this could lead to cracking due to the higher residual stress in duplex welds.

2.5 Weldability in general
The modern duplex steels with a well-balanced chemical composition, produce a HAZ with limited grain growth. A typical microstructure of a welded duplex material is seen in Figure 1. To reach a good ferrite/austenite balance in the weld metal, designed duplex filler metal shall be used. In Table 2, the chemical composition of Avesta Welding’s designed matching filler are given.

The duplex steels can be welded with most methods used for welding austenitic stainless steels. However, welding without filler (autogenous welding), as is common with resistance welding, laser- and EB-welding, might influence the microstructure negatively and thus affect properties of the weldments.

The effect of ferrite content is often discussed in detail. In most cases the effect of ferrite contents, in the range 20–75%, on corrosion and strength cannot be measured. When welding without filler, ferrite levels in the weld metal above 90% can sometimes be obtained. This can cause a reduction in elongation/strength and might give problems when the weldment is strongly deformed. The ferrite/austenite balance can also be influenced by cooling rate, the choice of filler, degree of fusion of the parent material and, to some extent, the composition of the shielding gas.

The welding speed in fully automatic production lines is, due to the composition of the steels, somewhat lower than when welding 300 series. The welder might observe a decreased penetration and a lower fluidity. To improve penetration and fluidity, argon shielding gas with addition of helium and hydrogen can be used. The use of nitrogen addition into the shielding gas (1–2%) will improve pitting corrosion resistance, ductility and strength (GTAW or PAW). Purging gases containing nitrogen are also beneficial for the corrosion resistance. The shielding gas when FCAW is normally 80% Ar + 20% CO₂, but pure CO₂ can also be used.

The use of SMAW gives the welder flexibility and possibility to weld in all positions.
and without gas shielding arrangement. This method is therefore often used on sites where it is difficult to automatize.

**2.6 Post weld cleaning**

As for other stainless steels, duplex steels shall normally be cleaned after welding to restore full corrosion resistance. If the construction shall be exposed to an environment containing halogenides and the medium itself will not remove the weld tint, the post weld cleaning operation is very important.

**2.7 Post weld heat treatment (PWHT)**

This operation may be difficult for welded duplex products. The reason is that some duplex steels give a faster formation of, for example, sigma phase during PWHT compared to standard austenitic steels. In complex constructions, where the thickness varies between different parts, the sections will be exposed to different thermal cycles, which could result in varying properties. In special cases, duplex welded constructions can be stress relieved at 550–580°C. If possible to perform, full annealing at about 1000–1100°C will give the best restoration of the material. In complex structures it might be necessary to support the structure to avoid large deformations caused by the high temperature.

**2.8 Typical weld defects**

Hot cracking in duplex steels is a rare phenomenon. Solidification cracking can however appear if there is a too narrow joint, too high welding speed and strong restraint conditions. Porosity can be a result of nitrogen gas formed from the weld metal or fused parent metal, moisture or other gas-forming substances. Thin beads give the gas bubbles the possibility to escape from the weld pool [2]. To avoid porosity due to the high content of nitrogen in the weld metal or parent metal the beads shall not be too thick and the welding speed not too high. Welding in under-up position (PE) also increases the tendency to porosity. GMAW normally gives more pores compared to other welding methods. The amount for acceptance/rejection is given in the quality level standard EN 25817.

**2.9 Welding LDX 2101® / EN 1.4162**

This steel has a very good reformation of austenite during welding. This implies that good results can be obtained even for autogenous and resistance welding of thin material. When welding with fillers, the matching Avesta LDX 2101® or 2205 can be used. The filler 2205 (22Cr9Ni3MoN) cannot be used in situations where the environment is strongly oxidizing; formation of intermetallic phases such as sigma phase is sluggish, but nitride precipitates may form in the HAZ and in autogenous weld metals [3]. By using a suitable heat input, acceptable weldment properties can be obtained. When welding with SAW, large degree of fusion of the parent material shall be avoided. The use of a basic flux gives a weld metal with better ductility. When using GMAW, the operator should use a modern pulse equipment and preferably three component shielding gas Ar + 30% He + 1.5–3% CO2 which has been found to give best weldability. The heat input and interpass temperature are, as for many other duplex stainless steels, 0.5–1.5 kJ·mm⁻¹ and < 150°C respectively.

**2.10 Welding 2304 / EN 1.4362**

The best weldment properties can be obtained by using designed filler material such as Avesta 2304 or 2205. The relatively low nitrogen content makes it possible to weld with high speed and with good penetration. The weldability is very good, at least as good as for the duplex steel 2205. The low molybdenum content is the main reason why it is not sensitive to deleterious precipitates during welding, but traces of nitrides may be found in the HAZ and in the weld metal.

GTAW can be carried out with pure argon as shielding gas. Autogenous welding shall be carried out with nitrogen addition to the shielding gas to increase strength. Gases for GMAW can be of the same type as described for LDX 2101®. To increase weld metal ductility in SAW metals, the use of a basic flux is advisable. The low nitrogen content in the steel implies that the resistance welding procedure should be optimised to increase the austenite
content in the weldment and strength/ductility/corrosion resistance in the weld metal/HAZ. When welding with other methods, the heat input interval can be somewhat wider than that used for LDX 2101® and 2507 (0.5-2.5 kJ·mm⁻¹). The interpass temperature should, as for most other duplex stainless steels, be below 150°C.

2.11 Welding 2205 / EN 1.4462
This steel is today’s most widely used duplex steel. The weldability is good and well documented. Due to extensive experience with the steel, highly efficient welding methods and procedures have been used with good results. The best weldment properties will be obtained if designed filler such as Avesta 2205 is used. Autogenous welding with GTA- and plasma should be carried out with addition of nitrogen to the shielding/purging gas to reach good weld metal properties. Resistance welding gives a very fast cooling rate and high ferrite content. Values of 80–90% are typical for 2205. One way to reduce this is to use a double pulse technique. The heat input to weld 2205 can typically be between 0.5 to 3.0 kJ·mm⁻¹.

2.12 Welding 2507 / EN 1.4410
This steel is mostly used when the construction is going to be exposed in very corrosive environments, combined with high stress levels. However, the high alloy content makes this steel more sensitive to welding compared to the other duplex steels presented above. As for most other duplex steels it is an advantage to use designed filler during welding (Avesta 2507/P100). The maximum recommended heat input is lower than for other duplex steels to minimize precipitates in the weld. The minimum level can be somewhat lower (0.2–1.5 kJ·mm⁻¹). The interpass temperature shall be below 100°C.

3. Typical weld metal properties
In most cases the weld metal strength is higher than the annealed parent material. The ductility/fracture-elongation is always lower. Duplex weld metals also give lower fracture elongation compared to austenitic welds, and lower impact strength. However, fracture mechanics investigations of duplex welded joints show however very good results.

Corrosion resistance is often strongly dependent on post weld cleaning, especially if there is a risk for localized corrosion. Any kind of geometrical deviation such as a crevice may also reduce fatigue strength or initiate corrosion attack. Welds are normally sensitive in this respect.

Duplex stainless steels and welds are, due to a high content of ferrite, much more sensitive to low temperatures compared to standard austenitics when it comes to impact strength requirements. Some typical weldment properties for different parent materials are given in Table 3. As a general rule, impact properties for slag-bearing methods

<table>
<thead>
<tr>
<th>Steel Grade</th>
<th>ASTM</th>
<th>EN</th>
<th>Welding Method</th>
<th>CPT (°C)</th>
<th>Weldment Mechanical Properties</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Rm (MPa)</td>
</tr>
<tr>
<td>LDX 2101®</td>
<td>S32101</td>
<td>1.4162</td>
<td>GMAW FCAW</td>
<td>&gt;6²</td>
<td>725</td>
</tr>
<tr>
<td>2304</td>
<td>S32304</td>
<td>1.4362</td>
<td>GMAW FCAW</td>
<td>&gt;10¹</td>
<td>640</td>
</tr>
<tr>
<td>2205</td>
<td>S32205</td>
<td>1.4462</td>
<td>GMAW FCAW</td>
<td>&gt;20¹</td>
<td>780</td>
</tr>
<tr>
<td>2507</td>
<td>S32750</td>
<td>1.4410</td>
<td>GMAW FCAW</td>
<td>35¹, 37¹</td>
<td>850</td>
</tr>
<tr>
<td>4401</td>
<td>316</td>
<td>1.4401</td>
<td>GMAW FCAW</td>
<td>&gt;8²</td>
<td>600</td>
</tr>
</tbody>
</table>

1) ASTM G 48-E  2) ASTM G 150
(SMAW, FCAW, SAW) are somewhat lower than for gas shielded methods (GMAW, GTAW, PAW, laser). This is a result of the different content of oxides in the weld metal. Further weld metal properties are also given in the literature [2–4].

The corrosion properties are illustrated by the critical pitting corrosion temperature (CPT) in Table 3, the values are given for thoroughly cleaned samples.

### 4. Field experience

Despite all the recommendations given to the fabricators we still see examples where the result has not been what the fabricators expected. Below, some typical cases, which occurred in reality on site, are shown.

#### 4.1 Hot cracking in SAW due a too narrow joint and too low width/depth ratio

- **Steel grade:** 2205
- **Plate thickness:** 15 mm
- **Welding method:** SAW
- **Joint angle:** 60°
- **W/D:** 0.6

**Defect:** Solidification cracking

**Cause:** A too narrow joint and too low W/D ratio. A too high welding speed could also be a reason.

**Solution:** The joint angle should be 80–90°. The land should be smaller to allow a higher voltage giving a higher W/D. A dragging angle (10–15°) also increases penetration. Reduce welding speed.

#### 4.2 Hot cracking in FCAW with a too narrow joint

- **Steel grade:** 2205
- **Plate thickness:** 25 mm
- **Welding method:** FCAW (against ceramic backing)
- **Joint design:** V-joint, 45°
- **Welding position:** PA

**Defect:** Solidification cracking

**Cause:** A too high welding speed will give a thin bead. The joint angle is also too narrow.

**Solution:** Decrease the welding speed. The recommended joint angle for this WPS is 60–70°. Sharp edges and wide joints tend to give flat and wide welds with great risk of solidification cracking.

#### 4.3 Slag inclusions in FCAW - with a too narrow joint

- **Steel grade:** 2205
- **Plate thickness:** \( t_1, t_2 = 15 \) mm
- **Welding method:** FCAW (against round backing)
- **Joint design:** Half V-joint
  - \( \alpha = 30° \)
  - \( D = 2 \) mm
  - \( C = 3 \) mm
- **Welding position:** PB

**Defect:** Slag inclusion

**Cause:** Too narrow joint, too small root gap and a too low current.

**Solution:** The recommended bevel angle (\( \alpha \)) shall be at least 50–55° and the minimum root gap (D) 3 mm.
4.4 Porosity formation in GMAW with too high welding speed

- Steel grade: LDX 2101®
- Plate thickness: 5 mm
- Welding method: GMAW (welded from both sides)
- Shielding gas: 100% Ar
- Welding speed: 450 mm·min⁻¹
- Joint design: V-joint
  - \( \alpha = 60^\circ \)
  - \( C = 2 \text{ mm} \)
  - \( D = 2 \text{ mm} \)

- Defect: Porosity

- Cause: Too high welding speed, not optimal shielding gas, somewhat thick bead.

- Solution: Decrease welding speed and decrease land (C). A better gas would be a three component gas such as Ar + 30% He + 2–3% CO₂.

4.5 Spatter due to direct copying of the WPS from 1.4401 when changing to LDX 2101®

- Steel grade: LDX 2101®
- Sheet thickness: 2 mm
- Welding method: GMAW
- Shielding gas: Argon + 30% He +1.8% N₂ + 0.03% NO
- Metal transfer: Short arc
- Joint design: Butt and fillet welds

- Defect: Spatter and bad arc stability.

- Cause: The use of a short arc metal transfer and an improper shielding gas.

- Solution: Use pulsed arc welding and a three component shielding gas, for example Ar + 30% He + 2–3% CO₂. Change/optimise the welding parameters after the new material and filler.

5. How to avoid unexpected defects when changing parent material

Unexpected phenomena in a material, which is new for the fabricator, can have many explanations. One could be the desire to produce in the most economical way. The fabricator's decisions are often based on available equipment, historical experience, knowledge, construction and the skill of the operators. For all these reasons the fabricator might not be able to or might not want to follow all recommendations and guidelines given by the steel producer and/or the filler material supplier. In many cases the decision by the fabricator to choose a welding procedure deviating slightly from the material supplier recommendations will be successful and will not cause any unacceptable phenomena. However, sometimes, effects can appear that cause appreciable economical loss.
One way to overcome this problem is to follow the standard procedure EN ISO 3834-2 that describes quality requirements for fusion welding. The standard gives guidelines to handle requirements before production starts, under production and what type of final testing shall be carried out before release of the welded construction. It also requires that involved personnel (welding engineers, welders, non-destructive testing personnel) has been trained for the job. Other standards can also be of interest for the producer to follow. The standard EN 287-1 describes how to qualify welders. By using EN ISO 15614-1 a suitable welding procedure that reflects the actual fabrication can be qualified. To verify that the actual welded construction is fit for purpose, the construction could finally be inspected with some kind of non-destructive testing. If for example this way of thinking is used, a fabricator will eliminate most of the unexpected problems that might appear when changing parent material from standard austenitic steel to a duplex material.

6. Conclusions

When changing an essential parameter such as parent material type in production, it is very important that all involved personnel is informed about the decision. The next step for the managers is to study consequences in prefabrication, equipment, and training of personnel, effects on full-scale production and finally qualification of the fabrication procedures including welding. The general fabrication routines used for austenitic grades can in principle be followed for handling duplex stainless steels, but some main deviations described in this paper must be observed. Examples in which fabricators have not used proper instructions, with failure as a result, illustrate the need to adapt to new materials. A simple way of implementing quality consciousness for fusion welding at a fabricators site is to follow the intentions in EN ISO 3834-2. By this method unexpected phenomena can be avoided when changing parent material.

References


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Fracture toughness of welded commercial lean duplex stainless steels

Henrik Sieurin, Royal Institute of Technology, Sweden (now at Det Norske Veritas, Norway)
Elin M. Westin, Outokumpu Stainless, Sweden
Mats Liljas, Outokumpu Stainless, Sweden
Rolf Sandström, Royal Institute of Technology, Sweden

Abstract
Duplex stainless steels are successfully used in pressure vessel applications due to high mechanical strength combined with superior corrosion resistance. There has been a debate concerning the toughness level of welds in duplex stainless steels. For this reason, fracture and impact toughness data were generated for 30 mm thick parent material and welds of two commercial duplex stainless steels; LDX 2101® (EN 1.4162, UNS S32101) and 2304 (EN 1.4362, UNS S32304). Welds were produced with flux core arc welding (FCAW) and resulting microstructures and mechanical properties are presented. These show that both steels exhibit high fracture toughness in base metal and welds, close to what was earlier obtained for the more common grade 2205 (EN 1.4462, UNS S31803). Fracture toughness data were evaluated with the master curve approach giving a reference temperature characterizing the onset of cleavage cracking. The reference temperatures were below -100°C for the parent materials and around -100°C for the welds. Established relations were used to correlate impact and fracture toughness transition temperatures. The results verify that these steels have satisfactory fracture toughness properties and that this can be predicted from impact toughness data.

Keywords: Duplex stainless steel, welding, fracture toughness, KJC, master curve

1. Introduction
Many papers have shown the favourable result of composing a material of two or more phases, and ferritic-austenitic or duplex stainless steel is one excellent example. Duplex steels contain approximately equal proportions of ferrite and austenite combining good properties of both phases. High general corrosion resistance and high toughness at low temperatures are obtained from the austenite, while ferrite contributes to improved strength and resistance to stress corrosion cracking. The high strength compared to austenitic grades can be used to reduce the wall thickness, which in turn lowers the material cost, welding time, transportation weight, ecological impact; consumable and energy consumption. However, to be able to fully utilize the strength benefits, proper design rules must be available. In general, the European code is based on both minimum yield strength and tensile strength. Unfortunately the safety factor for the tensile strength seems unnecessarily large, which is somewhat restrictive for duplex stainless steel.

For pressure vessels there are additional requirements on fracture toughness to avoid brittle failure. As duplex steels exhibit less ductile behaviour than austenitic steels, they cannot be treated in the same manner. Due to the lack of an adequate standard, duplex steels are treated as ferritic steels, which show a clear ductile-to-brittle transition curve at a certain temperature range. Welding can increase the transition temperature resulting in a local reduction of low temperature ductility. The toughness of duplex welds can be
reduced by a high ferrite content [1]. An equal balance of austenite and ferrite is retained by use of proper welding parameters and suitable filler metal, often over-alloyed with nickel [2]. The ferrite fraction consequently affects the toughness of duplex stainless steel, but the experience from fracture toughness studies is limited.

To utilize the high strength when designing pressure vessels in duplex stainless steels it is essential to verify high toughness to avoid risk of failures. The fracture toughness of the duplex grade 2205 has been studied by Dhooge and Deleu [3–6], Wiesner [7], and Sieurin and Sandström [8]. Crack tip opening displacement (CTOD) has verified high fracture toughness for both parent material and weld metal from different welding methods. Sieurin and Sandström [8] correlated the fracture toughness to the impact toughness by reference temperatures derived by Wallin [9]. Wallin’s master curve is the basis for the American testing and analysis standard ASTM E1921-97 [10]. It determines the fracture toughness in the brittle-to-ductile transition range, where the reference temperature, \( T_0 \), characterizes onset of cleavage cracking at elastic and/or elastic plus plastic instabilities. Ericsson et al. [11] showed that the transition temperature increased with increasing specimen thickness. Data from the work of Sieurin and Sandström [8] has been used to include high strength duplex stainless steel in the revision of the pressure vessel standard EN 13445. This is discussed in another work [12].

The variety in the duplex portfolio is large, which is why it is of importance to evaluate the fracture behaviour in other duplex grades than 2205. As the turbulence in nickel price has had a large effect on the material costs lately, the interest has increased in exchanging austenitic grades with low-alloyed duplex stainless steels where it is possible to utilize the high strength. The duplex stainless steel 2304 has the potential to replace 316; while 304 often can be substituted with the more recently developed LDX 2101®. In the latter steel, nickel, that has a documented positive effect on low temperature toughness, has partly been replaced by manganese and nitrogen. This emphasizes the significance of assessing the fracture toughness properties. The objective of the present work was to confirm that the duplex stainless steels LDX 2101® and 2304 show sufficient fracture toughness in both base and weld metal, and also that the impact toughness can be used for estimating the fracture toughness.

2. Experimental

The lean duplex materials were 30 mm thick LDX 2101® and 2304, solution annealed at 1100°C and 1050°C, respectively, followed by water quenching. The compositions of the used base metal and consumables are found in Table 1. The filler metals have higher nickel contents for improved austenite formation and to increase the toughness.

The weldments were produced in double-V joints by using shielded metal arc welding (SMAW) for the root bead and flux core arc welding (FCAW) for subsequent passes in the longitudinal direction. The welding parameters are given in Table 2.

Cross-sections of welds were polished to mirror finish with SiO₂ in the last step. The phase fractions were revealed by etching in modified Beraha II (60 ml H₂O + 30 ml HCl + 1 g K₂S₂O₅) and the ferrite content was assessed by image analysis. The hardness was determined with Vickers microhardness at a load of 0.5 kg. Tensile tests across the welds

<table>
<thead>
<tr>
<th>Filler</th>
<th>Method</th>
<th>t</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>Mn</th>
<th>N</th>
<th>C</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Cu</th>
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<tbody>
<tr>
<td>LDX 2101®</td>
<td>BM</td>
<td>30</td>
<td>21.5</td>
<td>1.60</td>
<td>0.29</td>
<td>4.92</td>
<td>0.23</td>
<td>0.032</td>
<td>0.66</td>
<td>0.025</td>
<td>0.001</td>
<td>0.28</td>
</tr>
<tr>
<td>2304</td>
<td>BM</td>
<td>30</td>
<td>22.7</td>
<td>4.69</td>
<td>0.31</td>
<td>1.44</td>
<td>0.10</td>
<td>0.020</td>
<td>0.50</td>
<td>0.026</td>
<td>0.001</td>
<td>0.25</td>
</tr>
<tr>
<td>LDX 2101®</td>
<td>SMAW</td>
<td>Ø 3.25</td>
<td>24.9</td>
<td>8.97</td>
<td>0.13</td>
<td>0.83</td>
<td>0.13</td>
<td>0.020</td>
<td>0.61</td>
<td>0.019</td>
<td>0.011</td>
<td>0.08</td>
</tr>
<tr>
<td>LDX 2101®</td>
<td>FCAW</td>
<td>Ø 1.2</td>
<td>24.1</td>
<td>7.35</td>
<td>0.21</td>
<td>0.69</td>
<td>0.13</td>
<td>0.045</td>
<td>0.72</td>
<td>0.032</td>
<td>0.012</td>
<td>0.22</td>
</tr>
<tr>
<td>2304</td>
<td>SMAW</td>
<td>Ø 3.25</td>
<td>24.6</td>
<td>9.13</td>
<td>0.13</td>
<td>0.78</td>
<td>0.141</td>
<td>0.025</td>
<td>0.77</td>
<td>0.020</td>
<td>0.025</td>
<td>0.10</td>
</tr>
<tr>
<td>2304</td>
<td>FCAW</td>
<td>Ø 1.2</td>
<td>24.8</td>
<td>8.95</td>
<td>0.12</td>
<td>0.80</td>
<td>0.123</td>
<td>0.021</td>
<td>0.60</td>
<td>0.018</td>
<td>0.012</td>
<td>0.07</td>
</tr>
</tbody>
</table>
Shielding gas used was Ar + 25% CO₂ + 0.03% NO at a flow rate of 20–25 l·min⁻¹

Parameters | LDX 2101® | 2304
---|---|---
Current, A | SMAW | FCAW* | SMAW | FCAW*
28 | 231 – 260 | 110 – 114 | 225 – 240
Voltage, V | 28 | 31 | 28 | 31
Welding speed, mm/s | 3.7 | 4.5 – 6.2 | 3.5 | 4.2 – 6.2
Heat input, kJ/mm | 0.9 | 1.3 – 1.6 | 0.9 | 1.2 – 1.7

*SMAW and FCAW parameters for welding of LDX 2101® and 2304

were conducted on specimens of DF13 type in accordance with the European standard EN 10002-1 [13] and ASTM A370 [14]. Bend tests were performed on ground welds to an angle of 180° using a mandrel diameter of three times the plate thickness following ASTM E 290-97a [15]. Impact toughness testing at various temperatures was carried out on full size Charpy-V specimens in base and weld metal according to EN 10045-1 [16].

Fracture toughness was tested in a rig containing a three point loading fixture and a cooling box specially developed for fracture toughness testing, Figure 1a. The box is filled with alcohol and contains chambers for liquid nitrogen, Figure 1b. The alcohol is stirred and can be cooled down to approximately -110°C. More information regarding the test rig and the control system can be found in Sieurin’s thesis [17].

A three-point bend specimen with an attached clip gauge can be found in Figure 1c. The geometry of the tested specimens was 30 x 60 x 400 mm. Notches were machined with transverse-longitudinal orientation with the notch located in the transverse plane in the longitude direction making the cracks grow along the elongated microstructure in the rolling direction. All specimens were full-thickness single-edge-notch bend-bars. The specimens were pre-fatigued at room temperature according to ASTM E 813-89 [18], generating cracks of a length of approximately 1.5 mm. Sidegrooves were machined after the pre-fatigue in order to force straight crack growth during the final test. The depth of each side-groove was 10% of the thickness. The distance between the outer loading points of the three-point bend specimens was 270 mm. The specimens were subjected to an increasing monotonic force at a displacement rate of 0.1·mm·s⁻¹, while the force and crack mouth opening displacement (CMOD) were measured to the point at which either

Fig. 1 Laboratory equipment and three-point bend specimen type used for fracture testing. (a) Component test rig. (b) Cooling box seen from above. (c) Three-point bend specimen with a clip gauge for measuring the CMOD.
brittle crack extension occurred, a maximum force plateau was reached or a discontinuity in the load-displacement record could be observed (a so-called pop-in).

The fracture toughness testing was based on ASTM E1921-97 [10]. This method was originally developed for ferritic steels, but as the ferrite controls the fracture behaviour in duplex stainless steels, the test method has been shown to be applicable also for these steels [8, 19]. The master curve method involves elastic-plastic theories and determination of a reference temperature, which characterizes the onset of cleavage cracking [20]. The J-integral describes the energy release rate for non-linear elastic materials (including duplex stainless steels) and the amount of energy required to initiate and propagate a crack. To determine the J-integral, the specimen is loaded while the CMOD is measured continuously. The area beneath the force-CMOD curve gives the amount of energy required for fracture and the critical J-integral, \( J_C \) [10] is evaluated at a pop-in or where a maximum force plateau is reached. A pop-in in this case was a marked, audible discontinuity in the force-CMOD record with a small temporary drop of the force. The \( J_C \)-value is used to calculate the elastic-plastic stress intensity factor, \( K_{JC} \), characterising failure after more or less extended ductile crack growth. The \( K_{JC} \) values obtained from the test series are used to derive a transition temperature, \( T_0 \), corresponding to a median fracture toughness of 100 MPa·m\(^{1/2}\) for a specimen with a thickness of 25 mm. The \( T_0 \) value is obtained by iteration of Eq. 1,

\[
\begin{align*}
\sum_{i=1}^{n} \frac{\delta_i \exp \{0.019 \left[ T_i - T_0 \right]\}}{11+77 \exp \{0.019 \left[ T_i - T_0 \right]\}} - \sum_{i=1}^{n} \frac{(K_{JCi} - 20)^{0.5} \exp \{0.019 \left[ T_i - T_0 \right]\}}{11+77 \exp \{0.019 \left[ T_i - T_0 \right]\}}^{5/4}
\end{align*}
\]

where \( \delta \) is a censoring parameter, which is 1 for uncensored data and 0 for censored data [10]. All data in the present study was uncensored. The reference temperature is used to establish a transition temperature curve, the master curve, Eq. 2.

\[
K_{JC} = 30 + 70 \exp \{0.019 \left[ T - T_0 \right]\}
\]  

Base and weld metal were tested at temperatures ranging from -30°C to -112°C. Complementary crack tip opening displacement (CTOD) was calculated according to ASTM E 1820-99a [21]. This has previously been described in the work of Sieurin and Sandström [8].

3. Results

3.1 Microstructure and mechanical properties

Figure 2 illustrates the microstructure in weld metal, heat-affected zone (HAZ) and base metal. The etchant colours the ferrite dark, distinguishing it from the bright austenite. The base metal austenite is elongated in the longitudinal direction and the microstructure in LDX 2101® was somewhat coarser than in 2304. The base metal austenite is distributed as laths in the ferrite matrix; while in the weld metal it is mainly present at the grain boundaries and as Widmanstätten plates growing into the grains. The weld metal microstructure was coarser for 2304 than for LDX 2101®. 2304 is more sensitive to grain growth causing larger grains in the HAZ and consequently in the weld metal. Intragranular austenite was formed due to reheating from subsequent weld beads. The amount of chromium nitride precipitates was low. Some minor porosity and slag particles were found in the weld metal.

The mechanical properties in the transverse direction of the parent metals are presented in Table 3. The strength of the materials increased and the ductility decreased with decreasing temperature, which was comparable with that of 2205 [8]. LDX 2101® has somewhat higher strength than 2304, which is primarily due to the higher nitrogen content in the former.

The mechanical properties in the transverse direction of welded materials, the weld metal ferrite content; the results from the Vickers microhardness and the bend test can be found in Table 4.
All welded tensile specimens failed in the base metal. The strength of the weld metal was higher compared to the parent metal with a slightly decreased elongation. The ferrite content after annealing and quenching (as-delivered condition) was 55 ± 4% in the LDX 2101® and 52 ± 2% in 2304. The ferrite fraction was somewhat lower in the weld metal due to extensive austenite formation and some secondary austenite formation. The hardness measurements showed no large differences over the welded cross-sections. LDX 2101® was, however, harder than 2304 and had somewhat higher strength. This could be related to the finer grain size and higher nitrogen content in LDX 2101® weld metal. All samples exposed to bend testing were free of defects.

3.2 Toughness

The impact toughness results for weld and base metal are presented in Table 5. 2304 base metal had higher impact toughness than LDX 2101® base metal as a result of higher nickel content and finer microstructure, while the weld metals showed comparable values due to similar final composition. The large difference between base and weld metal in 2304 can partly be correlated to a larger grain size in the weld metal. The corresponding average impact toughness in base metal (BM) and weld metal (WM) at various temperatures is shown in Table 5.
Temperatures where the impact toughness is 40 J, T40J, and 27 J, T27J, are presented in Table 6. The temperatures were estimated by fitting polynomial curves using a least square method to the data points. As will be discussed later, the T27J value is used for evaluating the fracture toughness reference temperature. The largest difference in T27J was seen between LDX 2101® base and weld metal.

The force-CMOD records for the base and weld metal fracture toughness testing can be found in Figure 3.

![Fig. 3](image.png)

Temperatures for 27J and 40J impact toughness in base metal (BM) and weld metal (WM)

<table>
<thead>
<tr>
<th>Material</th>
<th>Position</th>
<th>T27J (°C)</th>
<th>T40J (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>LDX 2101®</td>
<td>BM</td>
<td>-70</td>
<td>-51</td>
</tr>
<tr>
<td>LDX 2101®</td>
<td>WM</td>
<td>-93</td>
<td>-54</td>
</tr>
<tr>
<td>2304</td>
<td>BM</td>
<td>-106</td>
<td>-102</td>
</tr>
<tr>
<td>2304</td>
<td>WM</td>
<td>-92</td>
<td>-62</td>
</tr>
</tbody>
</table>

After reaching maximum force plateaus in the base metal, the forces decreased slowly with increasing CMOD. However, pop-ins were the critical event for evaluation of the base metals. These appeared close to the maximum force and were difficult to observe due to the limited size. All weld metal fractures were brittle after some plastic deformation and only occasional specimens showed pop-ins. The plastic part of the force-CMOD curves was most pronounced at the highest testing temperatures.
The fracture toughness results of LDX 2101® and 2304 base and weld metal can be found in Figures 4 and 5.

The reference temperatures for LDX 2101® were -112°C and -92°C for the base and weld metal, respectively. The deviation from the reference curve was rather large for the base metal, especially for the specimens tested at the highest temperatures. However, the toughness was high at all tested temperatures and the scatter was less than in previous studies [8]. The weld metal data fit the master curve well and the toughness was satisfactory.

The 2304 base and weld metal showed comparable fracture toughness with reference temperatures of -112°C and -109°C, respectively. Fracture toughness data usually gives a large scatter, but the master curve fit the base metal results well, while the weld metal...
showed larger scatter. The master curve is designed to give conservative values at low temperatures and the precision at higher toughness might be less accurate.

A comparison of the fracture toughness of LDX 2101® [19] and 2304 with 2205 [8] can be found in Figures 6 and 7. The base metal reference temperature was equal for LDX 2101® and 2304 base metal, 31°C higher than for 2205. The weld metal reference temperature was higher for all materials, with the largest difference for the 2205 welds.

Figure 8 shows the CTOD calculations for the LDX 2101® and 2304 base and weld metals. Wiesner [7] has recommended a minimum CTOD value of 0.1 mm as the minimum design limit.
4. Discussion

4.1 Fracture toughness

Impact or fracture toughness testing can be used to evaluate the fracture behaviour of a material. Fracture toughness testing is rather cost intensive. Eq. 3 shows a statistical correlation between fracture toughness and impact toughness that was established by Wallin [22] after testing 141 steels with yield strengths ranging from 300 to 1000 MPa. The fracture toughness reference temperature, $T_0$, can be evaluated here if the impact transition temperature, $T_{27J}$, is obtained by Charpy-V testing.

$$T_0 = T_{27J} - 18$$

$T_0$ is the temperature where $K_{JC}$ is 100 MPa$\sqrt{m}$ and $T_{27J}$ is the transition temperature for an impact energy of 27 J. The impact toughness testing in this work gave a $T_{27J}$ of -70°C to -106°C; hence resulting in calculated $T_0$ values between -88°C and -124°C. The $T_0$ results from the fracture toughness testing were between -92°C and -112°C indicating good correspondence with Eq. 3.

The $T_0$ values of LDX 2101® and 2304 weld metal were in parity with previous results obtained for 2205 weld metal [8], while the base metal of LDX 2101® and 2304 showed lower fracture toughness compared to 2205 base metal [8]. The high nickel content in 2205 contributes to the high fracture toughness. The relatively low fracture toughness for 2205 SAW metal is probably related to somewhat lower weld quality than normal. Thus, the fracture toughness of 2205 was significantly higher for the parent material than for the welds, while the difference was much smaller for LDX 2101® and 2304. However, Dhooge and Deleu [3] found high fracture toughness of both 2205 base and SAW weld metal.

4.2 Fracture surface

The fracture surfaces of base and weld metals can be found in Figure 9. The parent metals showed splits growing perpendicular to the fracture surface along the rolling direction. The splits were located in the short-transverse direction and were only present in the base metal. The number of splits increased with decreasing toughness as previously reported by Nilsson [23]. Both the splits and the main crack appeared to propagate within the ferrite.

---

**Fig. 8** CTOD measurements for LDX 2101® and 2304 base and weld metals. The dashed line indicates the minimum CTOD requirement [7]
The splits reduce the stress intensity at the crack tip and the fracture toughness increases by relaxation of triaxial stresses [24]. The fracture toughness of the base metal was estimated at the maximum force or when pop-ins occurred. The pop-ins could be caused by the splits, which might explain the poor correspondence of the data with the master curve for the base metal. Fracture toughness testing of the somewhat tougher 2205 base metal showed fewer but deeper and wider splits [8]. The larger number and the more even distribution of the splits in LDX 2101® and 2304 base metal might explain the smaller scatter of the $K_{JC}$ results compared to the 2205 base metal.

The fracture surface of the weld metal was a mixture of cleavage fracture and ductile, dimple rupture. Some of the welded specimens failed in the criterion of the straightness of the crack front according to the standard [10]. The cracks in the weld metal propagated along the columnar grains as seen in Figure 11.

**Fig. 9** Fracture surfaces of (a) LDX 2101® base metal tested at -50°C, (b) LDX 2101® weld metal tested at -60°C, (c) 2304 base metal tested at -90°C and (d) 2304 weld metal tested at -74°C

**Fig. 10** Split growing in the ferrite perpendicular to the fracture surface of (a) LDX 2101® and (b) 2304

**Fig. 11** Cross-section perpendicular to the fracture surface showing crack growth along the columnar weld microstructure in (a) LDX 2101® and (b) 2304
4.3 CTOD
Crack tip opening displacement (CTOD) has verified high fracture toughness of 2205 for both parent material and weld metal from different welding methods correlating fairly well with Charpy toughness data and measurable displacements were obtained also for quite low impact toughness [3-8, 25-27]. CTOD calculations of LDX 2101® and 2304 were performed for comparison since no K_{JC} and master curve evaluation for these materials have been found in the literature. Although the CTOD requirement seems to be more conservative than the reference temperature approach in ASTM E 1921-97 [10], a CTOD of 0.1 mm corresponds to low temperatures for LDX 2101® and 2304 base and weld metals. A correlation between CTOD and K_{JC} has been suggested [28] and Sieurin et al. [19] confirmed a similar relation for LDX 2101®.

5. Conclusions
The fracture and impact toughness of the duplex stainless steels LDX 2101® and 2304 were tested. The temperature dependence of the fracture toughness was described by the master curve, which involves determination of a reference temperature, T_{0}, characterizing the onset of cleavage cracking. The duplex grades were subjected to shielded metal arc welding (SMAW) of the initial root bead and flux core arc welding (FCAW) for the subsequent welds. The results were compared with 2205 base and submerged arc weld (SAW) metal. Both base and weld metals showed sufficient fracture toughness for most low temperature applications. The base metal reference temperature was -112°C for LDX 2101® and 2304, compared to -143°C for 2205. The reference temperature of the weld metals was -92°C for LDX 2101® and -109°C for 2304, compared to -101°C for 2205. The present study has shown high fracture toughness for the three commercial duplex grades LDX 2101®, 2304 and 2205, and that the fracture toughness can be correlated to the impact toughness.

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References


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