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Title: Use of low transformation temperature alloys to reduce distortion in austenitic CrNi steels and insight into corrosive behavior

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Abstract

Austenitic CrNi steels are increasingly being used in modern rail vehicle construction due to their high specific strength and good corrosion resistance. However, significant distortion of the welds can occur, particularly in long or flat sections of the outer shell of the railcar. It has already been demonstrated on several occasions that welding distortion can be effectively reduced by introducing a low transformation temperature (LTT) effect. Nevertheless, a decrease in the alloy content of the weld may lead to diminished corrosion resistance. To evaluate the corrosion behavior, various LTT alloys were welded with a laser beam using an in-situ alloying process and then tested for intergranular corrosion and in a salt spray test. Finally, corrosive behavior is compared with the resulting welding distortion, and the highest distortion reduction and the lowest corrosive attack were achieved with alloys with a Cr content between 15.7 % and 16.4 %.

Keywords

Laser beam welding, low transformation temperature (LTT) effect, corrosive behavior

1 Introduction

Modern rail cars are increasingly being built using austenitic, corrosion-resistant chromium-nickel steels. Steel grades such as 1.4318/AISI 301LN are favoured for their excellent mechanical properties and corrosion resistance. Due to the low carbon content (<0.03 wt.%) and increased nitrogen content (~0.15 wt.%), this material exhibits no significant signs of intergranular corrosion (IGC), even following sensitization [1]. However, a disadvantage of welding austenitic steels compared to mild steels is that they have a higher coefficient of thermal expansion ($\alpha_{\text{austenite}} \sim 2\alpha_{\text{ferrite}}$) and lower thermal conductivity ($\lambda_{\text{ferrite}} \sim 3.3\lambda_{\text{austenite}}$) [2]. Consequently, higher shrinkage stresses build up during welding, which are relieved through plastic deformation once the temperature-dependent yield strength is exceeded. After cooling, this results in increased welding distortion [3]. The low-transformation-temperature (LTT) effect can be used to control residual stress and welding distortion. In this approach, adjusting the chemical composition induces

a martensitic phase transformation at a low transformation temperature within the weld seam. The shrinkage stresses (tensile stresses) that build up during cooling are superimposed by transformation stresses (compressive stresses), which result from the restricted volume expansion of the austenite-martensite phase transformation [4]. Thus, both the resulting residual stresses and the weld distortion are effectively reduced [5, 6]. To enable an LTT effect in austenitic steel, the alloy content must be reduced using low-alloy filler wires. This is accompanied by an increase in C and Mn content, as well as a reduction in Cr, Ni and N, which alters corrosion resistance. In addition to being susceptible to intergranular corrosion (IGC), corrosion-resistant steels in chloride-containing environments may also develop pitting or crevice corrosion due to the local destruction of the passive layer [7]. Since the LTT effect is usually studied in combination with mild steel, there are no studies on the corrosion behavior of the alloys, but these are essential for industrial applications involving the LTT effect in austenitic CrNi steels.

2 Aim of the Investigation

The aim of the investigations presented here is to provide an initial insight into the corrosion behaviour of LTT alloys in 1.4318, and to compare this with their distortion reduction. For this purpose, various LTT alloys were produced using three different unalloyed and low-alloyed filler wires, as well as reference alloys using a matching filler wire. The distortion reduction of these alloys is compared, while their corrosion behavior regarding pitting corrosion, crevice corrosion and intergranular corrosion is investigated.

3 Materials and Experimental Details

In the welding trials, T-joints made of austenitic CrNi steel 1.4318 (X2CrNiN18 7/ ASTM 301LN), were welded, **Table 1**. The welded joints were executed as double-sided full-penetration welds both for the distortion and corrosion trails. The welding filler designated G19 9 (EN ISO 14343-A: G 19 9 L Si) is utilized as the standard welding filler for the base material, and the reference joints are welded from this. To produce alloys with an LTT effect, it is necessary to reduce the alloy content in the weld seam. For this purpose, three filler wires are used, an unalloyed G4Si1 (EN ISO 14341-A: G 4Si1) and two wire with low alloy content ER90S-B3 (DIN EN ISO 21952-B W 2C1M) as well as ER80S-B8 (DIN EN ISO 21952-A G CrMo9). A Trumpf TruDisk 16002 with a wavelength of 1030 nm, an fiber diameter of 300 μm and optics from Kugler with an aspect ratio of 1:3 was utilized, resulting in a focus diameter of 900 μm . Since the filler wires had different diameters, the wire feed was adjusted based on the filling volume $FV = v_w * \pi d^2 / 4$ instead of the wire feed speed v_w . Six parameter variations for each filler wire were welded for the distortion trails and three for the corrosion trails, **Table 2**. Apart from the varying filling volumes, only the laser power was varied between the distortion trails and corrosion trails, while all other parameters were kept constant. The T-joints consist of a 2 mm flange plate and a 3 mm web plate, for the distortion trails each with a length of 250 mm and a width of 150 mm. The plates were connected with three tack welds on each side before welding, **Figure 1** a), c) and d).

Table 1: Chemical composition of base material and welding consumables measured by optical emission spectrometry (OES).

Materials	Diameter d [mm]	Chemical composition [wt.%]							
		Cr	Ni	Mo	Mn	Si	Fe	N	C
1.4318	-	17.70	6.39	0.073	1.19	0.42	73.6	0.155	0.034
G19 9	1.0	19.70	10.18	0.16	1.72	0.73	67.0	0.052	0.025
G4Si1	1.0	0.047	0.021	0.006	1.69	0.93	97.0	0.004	0.081
ER90S-B3	0.9	2.44	0.029	1.00	0.54	0.53	95.0	0.006	0.109
ER80S-B8	0.9	8.76	0.13	1.02	0.50	0.54	88.4	0.003	0.070

During the welding process, the development of welding distortion was measured in situ using two linear variable differential transformer (LVDT) sensors. Both LVDT sensors, LVDT1 and LVDT2, are located in the center of the sheets in the direction of welding and at a distance of ± 50 mm from the center of the sheet, perpendicular to the weld, **Figure 1** a) and c). After welding, the samples were unclamped and the resulting welding distortion was determined. Cross sections were then taken from the seams, which were used to measure the seam geometry, chemical composition and hardness.

Table 2: Welding parameter for distortion and corrosion trails

Trail	Wire feed speed 1.0 mm v_w [mm/min]	Wire feed speed 0.9 mm v_w [mm/min]	Filling volume [mm ² x m/min] $FV = \frac{\pi d^2}{4} * v_F$	Laser power [W]	Welding angle (lateral) [°]	Welding speed [m/min]
Distortion	1.32	1.63	1.04	2240	30	1.5
	1.50	1.85	1.18			
	1.68	2.07	1.32			
	1.86	2.30	1.46			
	2.04	2.52	1.60			
	2.22	2.74	1.74			
Corro- sion	1.32	1.63	1.04	2560		
	1.68	2.07	1.32			
	2.22	2.74	1.74			

Further welding parameters:

Shielding gas flow Ar 4.6	[l/min]	20	Collimation length	[mm]	200
Fiber diameter	[μm]	300	Focus diameter	[μm]	900
Focus length	[mm]	600	Focus position	[mm]	0

For the corrosion tests, a 50 mm × 50 mm × 500 mm specimen was welded for each alloy and then cut into 70 mm long sections using a precision cutting machine. Two tests were conducted to evaluate corrosion behavior. The first was a salt spray test (NSS) in accordance with

DIN EN ISO 9227 [8], which was used to assess susceptibility to pitting corrosion. The second test, conducted in accordance with EN ISO 3651-2 [9], assessed susceptibility to intergranular corrosion using Method A (Strauss test). For the salt spray test, three typical surface conditions – as welded, brushed, and pickled – were examined. Brushing was performed using an austenitic steel brush until all heat tint was removed. Acid pickling was performed by immersion in a solution of 15 % nitric acid (HNO_3), 5 % hydrofluoric acid (HF), and water for 60 min [10]. The surface was then cleaned by rinsing with water. The salt spray test runs for 790 h in a corrosion test chamber. At a constant temperature of 35 °C, a 5 % NaCl solution with a pH between 6.5 and 7.2 is continuously sprayed. The volume of spray is defined by an average collection rate of 1.5 ml/h on a horizontal collection surface area of 80 cm². [8] In the Strauss test for intergranular corrosion, the welded specimens are examined without any further post-treatment. The standard specifies testing butt joints on sheet metal or pipes as test specimens, which are bent over a mandrel after the corrosion tests. For the alloys examined here, however, the test is performed on T-joints, eliminating the subsequent bending stress. For the test solution, 100 g of copper (II) sulfate pentahydrate ($\text{CuSO}_4 \cdot 5\text{H}_2\text{O}$) is dissolved in 700 ml of distilled water. Then 184 g (100 ml) of sulfuric acid ($\rho_{20} = 1.84 \text{ g/ml}$) is added, and the volume is made up to 1,000 ml with distilled water. The samples are embedded in electrolytic copper in the form of chips at the bottom of the flask and boiled in the test solution for 24 h. Subsequently, the samples were examined for intergranular corrosion using scanning electron microscope (SEM). [9]

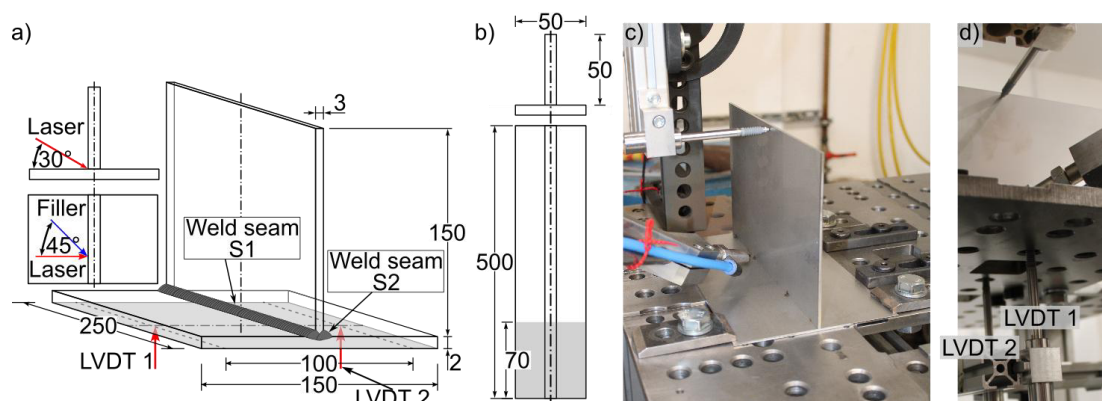


Figure 1: Component geometry of the distortion samples (a), sample geometry for corrosion testing (b), setup with clamping fixtures and LVDT positions (c and d).

4 Results and Discussion

Initially, cross-sections were used to examine the weld quality, hardness values and alloy content. Cross-sections of the distortion trails show that, as the filling volume increases, the weld head area increases and the penetration depth decreases, **Figure 2** a) to f). With the laser power of 2240 W used, a constant weld cross-sectional area of $6.96 \pm 0.21 \text{ mm}^2$ is achieved. However, increasing the filling volume alters the ratio of molten filler wires to base material, represented by the seam ratio parameter, equation (1) and **Figure 2** j). Consequently, the seam ratio increases with an increase in filling volume, resulting in a higher proportion of filler wires in the weld and a weld with a more pronounced seam head. Furthermore, when the filling volume FV exceeds $1.46 \text{ mm}^2 \times \text{m/min}$, the seam transition angle between the flange plate and the weld becomes very small (less than 90 °)

and there is almost no overlap between the two welds, **Figure 2 d) to f)**. To achieve a flatter transition angle and greater protection against an unwelded root gap, the laser power was increased to 2560 W for the corrosion tests, **Figure 2 g) to i)**. Increased laser power results in an increase in the seam cross-sectional area to $7.52 \pm 0.5 \text{ mm}^2$, and with the same filling volume this leads to a reduced seam ratio compared to the distortion trails.

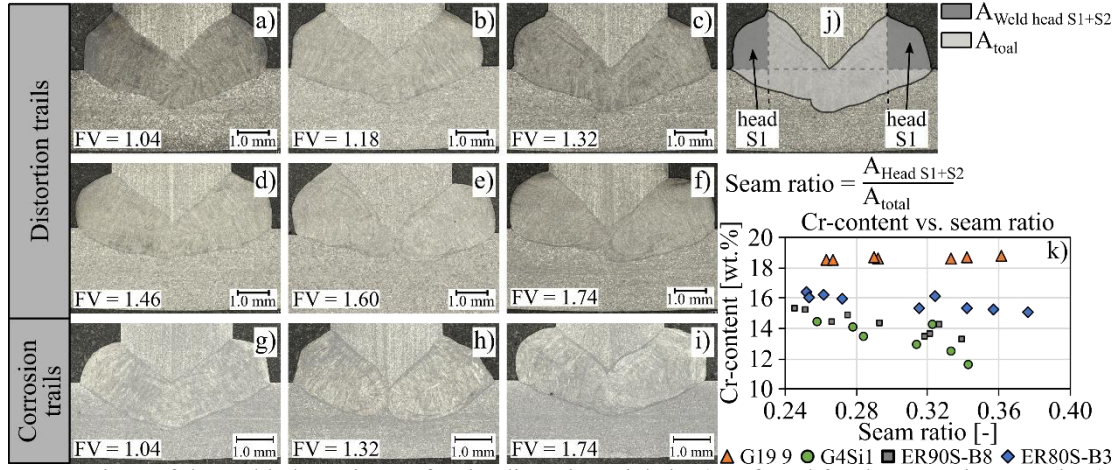


Figure 2: Cross-sections of the welded specimens for the distortion trials in a) to f) and for the corrosion tests in g) to i). Definition of the seam ratio in j), the relationship between alloy content and seam ratio, using chromium as an example in k).

The alloy content of the seams was determined using energy-dispersive X-ray spectroscopy (EDS), **Figure 2 k)** and **Table 3**. For the reference series using G19 9 as the filler wire, the alloy content increases as the seam ratio increases, since the wire has a higher alloy content than the base material. For the three wires intended to generate the LTT effect in the weld, however, an increase in the seam ratio leads to a reduction in the alloy content. The Cr content ranged from $11.63 \pm 1.21 \text{ wt. %}$ (G4Si1, FV 1.74, seam ratio 0.34) to $16.24 \pm 0.57 \text{ wt. %}$ (ER80S-B8, FV 1.04, seam ratio 0.262).

$$\text{Seam ratio} [-] = \frac{A_{\text{Weld head S1}} + A_{\text{Weld head S2}}}{A_{\text{total weld seam}}} \quad (1)$$

$$M_s [^{\circ}\text{C}] = 561 - 474 * \%C - 33 * \%Mn - 17 * (\%Ni + \%Cr) - 21 * \%Mo \quad (2)$$

$$PREN = \%Cr + 3.3 * \%Mo + 16 * \%N \quad (3)$$

The LTT effect is enabled by an increase in the martensite start temperature (M_s), which is caused by a reduced alloy content. To account for the effects of other important alloying elements on the LTT effect, such as C, Mo and Mn, the M_s temperature can be calculated using the empirical equation (2) from [11]. A comparison of the hardness values shows that the reference series with G19 9 is not hardened, regardless of the filling volume used and the resulting alloy content. In the welds, hardness is slightly reduced to $217 \pm 11 \text{ HV1}$, while the base material has a hardness of $239 \pm 13 \text{ HV1}$. For welds exhibiting the LTT effect, hardness values range from 337 HV1 to 414 HV1, **Figure 3 a)**. The highest hardness values occur at an M_s temperature of approximately 180 °C, decreasing as the M_s temperature drops. Based on the achieved hardness values, an LTT effect can be expected in all alloys containing G4Si1, ER90S-B3 and ER80S-B8.

Table 3: Chemical composition of the alloys measured using EDS

Trail	Wire	Filling volume [mm ² x m/min]	Chemical composition [wt.%]						
			Cr	Ni	Mo	Mn	Si	C*	N*
Distortion	G19 9	1.04	18.49 ± 0.37	7.12 ± 0.38	0.44 ± 0.21	1.34 ± 0.25	0.54 ± 0.1	0.032*	0.128*
	G19 9	1.18	18.49 ± 0.8	7.1 ± 0.8	0.43 ± 0.39	1.32 ± 0.51	0.54 ± 0.21	0.032*	0.127*
	G19 9	1.32	18.58 ± 0.56	7.21 ± 0.62	0.34 ± 0.26	1.35 ± 0.36	0.55 ± 0.15	0.031*	0.125*
	G19 9	1.46	18.6 ± 0.8	7.24 ± 0.86	0.43 ± 0.39	1.39 ± 0.52	0.54 ± 0.21	0.031*	0.121*
	G19 9	1.60	18.7 ± 0.77	7.41 ± 0.81	0.47 ± 0.38	1.38 ± 0.47	0.55 ± 0.2	0.031*	0.12*
	G19 9	1.74	18.78 ± 0.72	7.45 ± 0.7	0.43 ± 0.33	1.39 ± 0.44	0.56 ± 0.19	0.031*	0.118*
	G4Si1	1.04	14.41 ± 0.77	4.99 ± 0.81	0.43 ± 0.38	1.19 ± 0.52	0.55 ± 0.23	0.051*	0.12*
	G4Si1	1.18	14.06 ± 1.13	4.97 ± 0.79	0.45 ± 0.4	1.22 ± 0.5	0.57 ± 0.23	0.055*	0.112*
	G4Si1	1.32	13.42 ± 0.96	5.03 ± 0.81	0.43 ± 0.38	1.23 ± 0.51	0.54 ± 0.22	0.055*	0.112*
	G4Si1	1.46	12.89 ± 1.11	4.78 ± 0.81	0.44 ± 0.4	1.25 ± 0.5	0.6 ± 0.24	0.056*	0.109*
	G4Si1	1.60	12.52 ± 1.19	4.18 ± 0.89	0.43 ± 0.37	1.2 ± 0.45	0.62 ± 0.22	0.06*	0.102*
	G4Si1	1.74	11.63 ± 1.21	4.05 ± 0.5	0.45 ± 0.25	1.21 ± 0.29	0.62 ± 0.12	0.059*	0.104*
	ER90S-B3	1.04	15.19 ± 1.02	5.05 ± 0.68	0.43 ± 0.3	1.25 ± 0.41	0.57 ± 0.18	0.046*	0.116*
	ER90S-B3	1.18	14.77 ± 0.8	4.95 ± 0.76	0.42 ± 0.38	1.26 ± 0.5	0.56 ± 0.21	0.047*	0.113*
	ER90S-B3	1.32	14.24 ± 1.15	4.76 ± 0.57	0.34 ± 0.27	1.27 ± 0.33	0.57 ± 0.14	0.047*	0.112*
	ER90S-B3	1.46	14.21 ± 1.17	4.56 ± 0.75	0.42 ± 0.38	1.29 ± 0.44	0.6 ± 0.21	0.049*	0.108*
	ER90S-B3	1.60	13.36 ± 0.93	4.38 ± 0.54	0.44 ± 0.25	1.28 ± 0.28	0.61 ± 0.12	0.05*	0.105*
	ER90S-B3	1.74	13.18 ± 0.94	4.09 ± 0.81	0.44 ± 0.4	1.34 ± 0.5	0.64 ± 0.22	0.05*	0.103*
	ER80S-B8	1.04	16.24 ± 0.65	5.05 ± 0.68	0.43 ± 0.3	1.25 ± 0.41	0.57 ± 0.18	0.046*	0.116*
	ER80S-B8	1.18	15.9 ± 0.83	4.95 ± 0.76	0.42 ± 0.38	1.26 ± 0.5	0.56 ± 0.21	0.047*	0.113*
	ER80S-B8	1.32	16.08 ± 0.96	4.76 ± 0.57	0.34 ± 0.27	1.27 ± 0.33	0.57 ± 0.14	0.047*	0.112*
	ER80S-B8	1.46	15.34 ± 0.79	4.56 ± 0.75	0.42 ± 0.38	1.29 ± 0.44	0.6 ± 0.21	0.049*	0.108*
	ER80S-B8	1.60	15.21 ± 1	4.38 ± 0.54	0.44 ± 0.25	1.28 ± 0.28	0.61 ± 0.12	0.05*	0.105*
	ER80S-B8	1.74	15.02 ± 0.91	4.09 ± 0.81	0.44 ± 0.4	1.34 ± 0.5	0.64 ± 0.22	0.05*	0.103*
Corrosion	G19 9	1.32	18.7 ± 0.54	5.19 ± 0.81	0.51 ± 0.43	1.08 ± 0.49	0.48 ± 0.22	0.052*	0.118*
	G4Si1	1.04	15.05 ± 0.88	5 ± 0.7	0.53 ± 0.34	1.07 ± 0.42	0.49 ± 0.18	0.055*	0.114*
	G4Si1	1.32	14.27 ± 0.79	4.81 ± 0.8	0.58 ± 0.43	1.04 ± 0.47	0.5 ± 0.19	0.056*	0.111*
	ER90S-B3	1.04	15.18 ± 0.95	4.8 ± 0.84	0.58 ± 0.42	1.02 ± 0.48	0.51 ± 0.21	0.058*	0.106*
	ER90S-B3	1.32	14.37 ± 0.84	4.52 ± 0.78	0.62 ± 0.47	0.99 ± 0.49	0.51 ± 0.21	0.058*	0.107*
	ER90S-B3	1.74	13.54 ± 0.54	4.38 ± 0.71	0.58 ± 0.44	0.99 ± 0.42	0.5 ± 0.2	0.059*	0.104*
	ER80S-B8	1.04	16.38 ± 0.81	4.8 ± 0.84	0.58 ± 0.42	1.02 ± 0.48	0.51 ± 0.21	0.058*	0.106*
	ER80S-B8	1.32	16.03 ± 0.6	4.52 ± 0.78	0.62 ± 0.47	0.99 ± 0.49	0.51 ± 0.21	0.058*	0.107*
	ER80S-B8	1.74	15.3 ± 0.94	4.38 ± 0.71	0.58 ± 0.44	0.99 ± 0.42	0.5 ± 0.2	0.059*	0.104*

$$*C(N) \text{ content [wt. \%]} = m_{\text{Filler}} * C_{\text{Filler}} + m_{\text{Base}} * C_{\text{Base}}$$

$$m_{\text{Filler}} = \frac{A_{\text{Head},S1} + A_{\text{Head},S2}}{A_{\text{total}}}$$

$$m_{\text{Base}} = 1 - m_{\text{Filler}}$$

In the reference series with G19 9, the resulting weld distortion rises from 2.2° to 2.82° as the seam ratio rises, **Figure 2 b**). This is due to the increasing volume of the weld heads, since the shrinking volume acts with maximum leverage in this case. Alloys exhibiting the LTT effect generally display the same behaviour. For G4Si1, distortion increases from a minimum of 1.02° to a maximum of 1.90° , while for ER80S-B8 it increases from a minimum of 0.57° to a maximum of 1.40° . However, compared to G19 9, there is a reduction in distortion depending on the alloy content respectively the Ms temperature. Thus, the smallest reduction in distortion is achieved with G4Si1 at the highest Ms temperature and the largest with ER80S-B8 at the lowest. Alloys containing ER90S-B3 fall between G4Si1 and ER80S-B8 in terms of distortion and alloy content. On average, distortion reduction is 36 % for G4Si1 and 54 % for ER80S-B8 compared to the reference with G19 9.

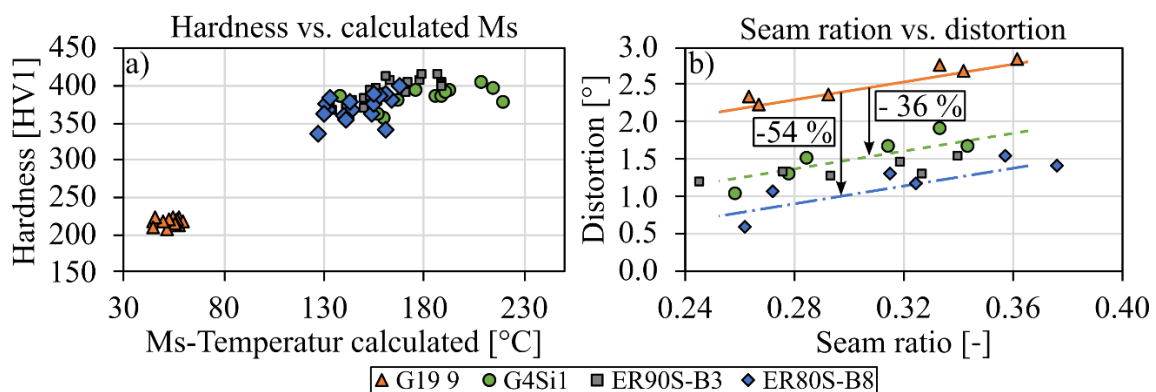


Figure 3: Effect of the achieved Ms-temperature on the hardness of the seams (a), effect of seam ratio on the resulting distortion angle (b).

Both pitting and crevice corrosion are localised forms of corrosion that occur particularly in chloride-containing environments [12]. While their initiation mechanisms differ – pitting corrosion typically results from a local breakdown of the passive film [13], whereas crevice corrosion is driven by restricted mass transport and oxygen depletion within confined geometries [14] – both processes share a common propagation mechanism.

Following the local destruction of the passive layer, a small anodic area forms within the pit or crevice, while the surrounding, still-passivated surface acts as a large cathode, leading to high local current densities and rapid metal dissolution [12]. The process is further accelerated by the accumulation of metal ions, local acidification, and chloride enrichment within the occluded region, which inhibit repassivation and promote continued corrosion [15]. The weld seams are subject to a salt spray test to determine their susceptibility to pitting and crevice corrosion. However, due to the similarity of these two corrosion phenomena, it is not always possible to distinguish between them.

After exposure in the salt spray test, the resulting corrosion attack is evaluated on cross-sections of the specimens. As-welded specimens exhibit the greatest number and depth of cavities, reaching up to 2.2 mm, **Figure 4 a**). Post-treatment of the surface by brushing or pickling reduces the depth and number of cavities. In addition to the deep cavities, uniform corrosion also occurs at the weld seams, as represented by the median. The respective results are 15 μm for as-welded, 13 μm for

brushed and 17 μm for pickled specimens. After 790 h in the salt spray chamber, hardly any corrosion products were visible on the pickled welds compared to the other variants. Therefore, the increased roughness of the pickled surface is likely due to excessive pickling, **Figure 4 b) to g)**.

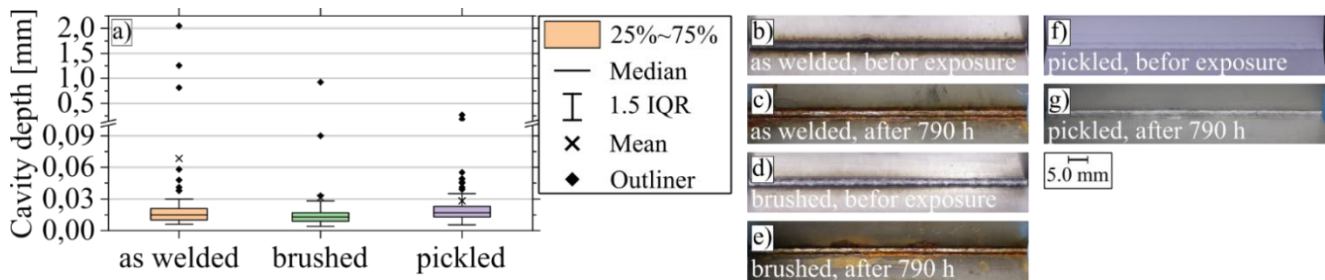


Figure 4: Effect of surface post-treatment on the depth of the resulting cavities in a). Examples of the three surface conditions before and after exposure in the salt spray chamber using the ER90S-B3 alloy with FV 1.04 $\text{mm}^2 \times \text{m/min}$ in b) to g).

The Pitting Resistance Equivalent Number (PREN), calculated using equation (3), is frequently used as a benchmark for a material's resistance to pitting and crevice corrosion [16]. At PREN values of up to 18.27, severe corrosion is observed, with cavity depths of up to 2.2 mm, **Figure 5 e)**. The maximum cavity depth of 2.2 mm was measured on the as-welded specimen with ER90S-B3 and FV1.04, with almost the entire weld being destroyed. As PREN increases, the maximum cavity depth decreases to 0.09 mm for $\text{PREN} > 18.27$, **Figure 5 c), d) and f)**. Additionally, the appearance of the corrosion attack changes. While distinct cavities form at low PREN values, high PREN values tend to result in general surface roughening, with no local, deep corrosion pits present.

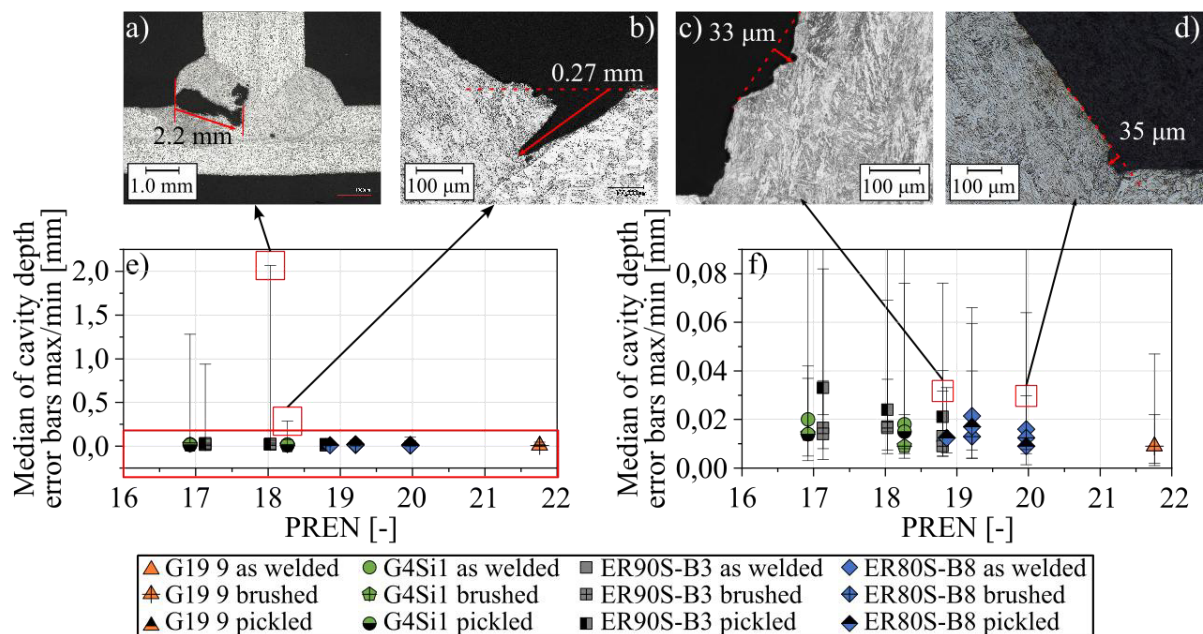


Figure 5: Results after 790 hours of exposure in the salt spray chamber. Cross-sections showing the extent of corrosion in a) through d). Summary of the results based on the median cavity depth, with the maximum and minimum cavity depths used as error bars in e) and f).

In addition to the salt spray test, susceptibility to intergranular corrosion was investigated using the Strauss test. With the exception of three specimens, the resulting IGC crack length is less than

60 μm , **Figure 6** c), d) and g). As with the salt spray test, the crack length decreases with increasing chromium content. At a chromium content of over 16 wt.%, the maximum crack length is 20 μm , **Figure 6** d) and g). No IGC cracks were detected in the reference sample with 18.7 wt.% Cr. For the three outliers, crack lengths of up to 650 μm were observed, with all cracks occurring at the fusion line, **Figure 6** a), b) and f). These cracks occur in the specimen with higher filling volumes: ER80S-B8-FV1.74 (650 μm), G4Si1-FV1.32 (260 μm) and ER90S-B3-FV1.74 (150 μm). Inhomogeneities in the alloy composition can particularly occur at high filling volumes, which is especially evident at the fusion line where partially molten and solid phases coexist, **Figure 6** e) and h). Consequently, the Cr content can vary significantly in these areas, potentially serving as an initiation point for IGC.

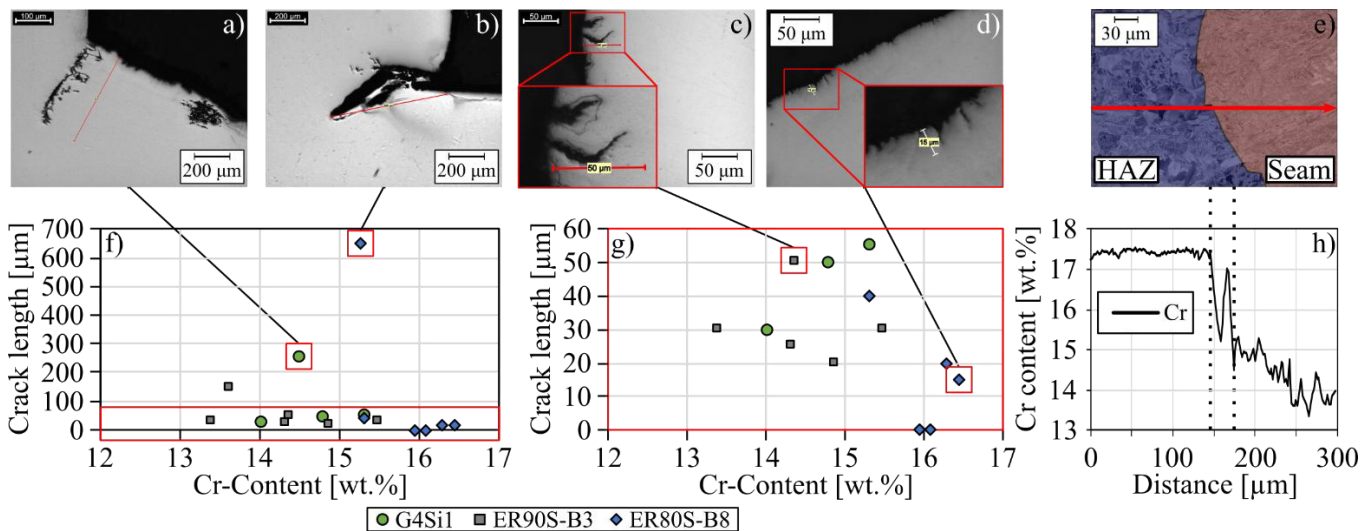


Figure 6: IGC crack length and SEM images in a)–d) and f)–g). Cr content at the fusion line in e) and h)

5 Summary

Both corrosion tests revealed that susceptibility to corrosion decreases as the alloy content (PREN and/or Cr) increases. However, corrosion was still observed in the alloys with the highest alloy content in both the salt spray and Strauss tests. No cavities were detected in the salt spray test starting at a PREN value of 18.27; only general surface roughening with a maximum depth of 90 μm was observed. In the Strauss test, deep intergranular corrosion (IGC) cracks were only found at the weld transition, which is attributable to inhomogeneity in the chemical composition. All other samples exhibited IGC cracks with a depth of less than 60 μm , with the depth of the cracks decreasing as the Cr content increased. Regarding component distortion, it was found that a small filling volume leads to lower distortion values. If an LTT effect is induced in the weld by reducing the alloy content, distortion is always reduced compared to the reference alloy with the same seam ratio. For alloys with the highest alloy content due to a small filling volume ($M_s \sim 130^\circ\text{C}$, Cr ~ 16.4 wt.%), an average distortion reduction of 54 % is achieved. Therefore, it is evident that a high alloy content with a small filling volume is advantageous in terms of corrosion and distortion. Ultimately, a decision must be made on a case-by-case basis as to whether the corrosion phenomena, which still occur even at high alloy contents, outweigh the advantage of reduced distortion. When combined with additional corrosion protection measures, which are frequently

used for finished parts, the reduction in distortion could be more beneficial than slightly lower corrosion resistance.

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Conflict of Interest

The author declares no conflict of interest.

Data Availability Statement

The data that support the findings of this study are available at <http://hdl.handle.net/21.11102/042ce519-24d5-47d5-94c4-4acf1385d14f> upon request.

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