J. S. Afr. Inst. Min. Metall., vol. 90, no. 3.
Mar. 1990. pp. 59-65.

Literature review on the influence of weld-heat inputs on the mechanical and corrosion properties of duplex stainless steels

by J.W. FOURIE* and F.P.A. ROBINSON†

SYNOPSIS

Austenitic stainless steels are welded using low-heat inputs. However, owing to differences in the physical metallurgy of austenitic stainless steels and duplex stainless steels, low-heat inputs should be avoided for duplex stainless steels. This paper highlights the differences in solidification mode and transformation characteristics of these two types of alloys. Recent work on the effect of heat input on the properties of weldments of duplex stainless steels is reviewed.

SAMEVATTING

Oustenitiese vlekvrye staal word met gebruik van laehitte-insette gesweis. Maar as gevolg van die verskille in die fisiese metallurgie van oustenitiese vlekvrye staal en dupleksvlekvryestaal moet laehitte-insette vir dupleksvlekvryestaal vermy word. Hierdie referaat bring die verskille in die stollingswyse en transformasie-eienskappe van hierdie twee soorte legerings na vore. Daar word 'n oorsig gegee oor onlangse werk in verband met die uitwerking van die hitte-inset op die eienskappe van sweisstukke van dupleksvlekvryestaal.

Introduction

Austenitic stainless steels are used extensively in corrosive environments. A major drawback of these steels is their susceptibility to chloride-induced stress-corrosion cracking. Although ferritic stainless steels are far more resistant to this type of corrosion damage, they do not have quite the ductility and weldability of the austenitics. By choice of a suitable composition and with the right thermo-mechanical processing, it is possible to produce stainless steels having a two-phase structure consisting of austenite and ferrite. Such stainless steels are referred to as duplex stainless steels, and they offer certain advantages over single-phase grades, including higher yield strengths and resistance to stress-corrosion cracking. Duplex stainless steels are finding increasing use in South Africa, and the properties of one such steel, SAF 2205, have been reviewed by Hoffman[1].

It is very important for stainless steels to be weldable, and this aspect has been the subject of intensive research. For the commonly used austenitic stainless steels, it is desirable to have the material in the solution-annealed condition for optimum corrosion resistance. For this reason, weld thermal cycles are chosen to ensure rapid cooling of the weld metal and the adjacent heat-affected zone (HAZ). This is done to prevent the formation of deleterious phases that could adversely affect the corrosion properties. Heat inputs and interpass temperatures are therefore often restricted to 1,5 kJ/mm and 150°C respectively[2].

Welding recommendations for duplex stainless steels were originally based on experience with austenitic stainless steels. However, recent research has shown that higher heat inputs and slower cooling rates can actually be beneficial to the weldment properties of duplex stainless steels. It is the aim of this paper to highlight the differences in the physical metallurgy of austenitic and duplex stainless steels, and to review recent work conducted on the weldment properties of duplex stainless steels.

Metallurgy

The Duplex Structure

Although duplex stainless steels can consist of a mixture of austenite and martensite, or of ferrite and martensite, most duplex stainless steels consist of ferrite and austenite. Typical compositions of some of the more common duplex stainless steels are presented in Table I. Many other grades of duplex stainless steels exist, but most of these would fall in the same compositional range as shown in Table I. It is clear from this table that a wide range of compositions exist, extending from 3RE60 with 18 per cent chromium to the highly alloyed Zeron 100.

Fig. 1 is a pseudo-binary phase diagram for chromium and nickel with 70 per cent iron[4]. The composition of a typical duplex stainless steel falls in the $\alpha + \gamma$ phase field. For many stainless-steel compositions, the austenite phase is expanded so that the ferrite phase is separated into high- and low-temperature ferrite. This has led to the practice of denoting high-temperature ferrite as delta ferrite, and low-temperature ferrite, which forms by the transformation of austenite, as alpha ferrite. It is evident from Fig. 1 that ferrite exists continuously from solidification to room temperature for duplex stainless steels, and all ferrite is therefore denoted as alpha ferrite. Because the $\alpha/(\alpha + \gamma)$ and $(\alpha + \gamma)/\gamma$ phase boundaries are not vertical, the ratio of ferrite to austenite in a particular grade would depend on the exact composition, as well as the thermo-

* Middelburg Steel & Alloys (Pty) Ltd, P.O. Box 133, Middelburg, 1050 Transvaal.
† Department of Metallurgy and Materials Engineering, University of the Witwatersrand, P.O. Wits, 2050 Transvaal.
 SA ISSN 0038-223X/3.00 + 0.00. Paper received 28th April, 1989.

TABLE I
TYPICAL COMPOSITIONS (IN PERCENTAGES BY MASS) OF SOME COMMONLY USED DUPLEX STAINLESS STEELS[3]

Alloy	Cr	Ni	Mo	Cu	N	C (max)	Other
* CD4MCu	25,0	5,5	2,1	3,2		0,04	Nb 0,2
AISI 329	26,0	5,0	1,5			0,08	
3RE60	18,5	4,7	2,7			0,03	
2304	23,0	4,0			0,1	0,03	
2205	22,0	5,5	3,0		0,14	0,03	
Uranus 50	20,5	7,5	2,5	1,5		0,03	
* Zeron 25	24,4	5,5	2,5		0,12	0,04	
Ferralium 255	26,0	5,5	3,0	1,7	0,17	0,08	
* Fermanal	27,0	8,5	3,1	1,0	0,23	0,08	
* Zeron 100	25,0	6,5	3,5	1,0	0,25	0,03	W 1,0

* Used for castings

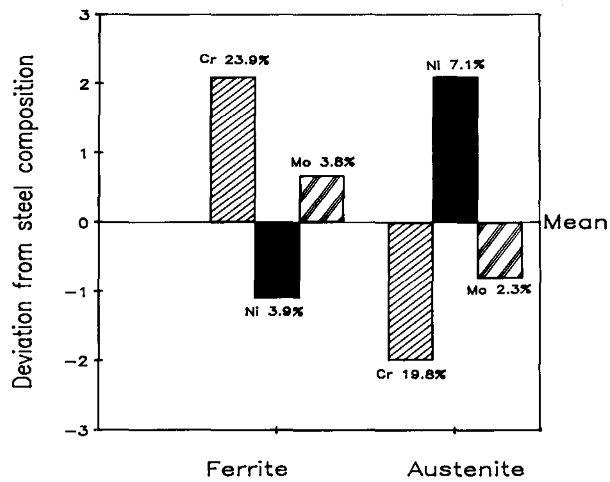


Fig. 2—The composition of the austenite and the ferrite phase of AF22 annealed at 1000 °C (mean steel composition = 21,8% Cr, 5,0% Ni, and 3,12% Mo)[5]

mechanical processing. Most modern duplex stainless steels are designed to have a 50:50 ratio of austenite to ferrite. As a result of the duplex structure, partitioning of alloying elements exists between the phases. Herbsleb and Schwaab[5] measured the phase composition of a 2205 type alloy and their results are shown in Fig. 2. Although the nitrogen content was not measured, significant segregation of nitrogen to austenite can be expected owing to its low solubility in ferrite at room temperature.

It is important to note that the solidification mode of duplex stainless steels differs from that of austenitic steels having residual delta ferrite. This has been discussed in some detail by Suutala *et al.*[6−8] for stainless steel weldments. They found that a Schaeffler alloy (Cr_{eq} = Cr + Mo + 1,5Si + 0,5Nb; Ni_{eq} = Ni + 30C + 0,5Mn[9]) with a Cr_{eq}/Ni_{eq} ratio of less than 1,48 would solidify primarily as austenite, and that delta ferrite is formed from the chromium- and molybdenum-enriched residual melt between the austenite cells or dendrites[6]. As expected, this delta ferrite would therefore be enriched in chromium and molybdenum.

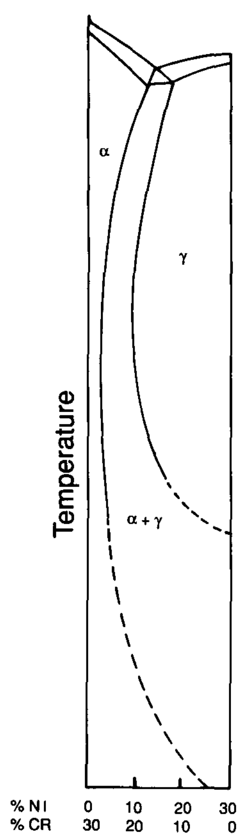


Fig. 1—A pseudo-binary phase diagram for chromium and nickel at 70 per cent iron[4]

If the alloy has a Cr_{eq}/Ni_{eq} ratio of between 1,48 and 1,95, primary solidification occurs in the ferritic–austenitic mode in which ferrite forms first and austenite forms as a result of the peritectic reaction of ferrite with the liquid or it forms eutectically[8]. This process also results in the segregation of alloying elements upon solidification, with the gamma phase enriched in austenite-forming elements and the alpha phase in ferrite-forming elements. For steels of this Cr_{eq}/Ni_{eq} ratio, the remaining ferrite can also undergo solid-state transformation to austenite[7].

Weldments of duplex stainless steel having a Cr_{eq}/Ni_{eq} ratio of greater than 1,95 solidify as single-phase ferrite. Because of the high diffusivity of chromium and molybdenum in ferrite, the ferrite solidification is not accompanied by significant segregation. The austenite is formed from the ferrite in the solid state by a Widmanstätten mechanism[7]. Segregation of ferrite stabilizers to alpha, and of austenite stabilizers to gamma, occurs during the solid-state transformation, and not during solidification as with Cr_{eq}/Ni_{eq} ratios of less than 1,95. Because of this segregation in the solid state, the phase balance and amount of segregation will be controlled by factors such as cooling rate for castings and weldments, whereas the thermo-mechanical processing conditions and annealing treatment will be very important for wrought products. The fact that solidification of duplex weldments occurs in the single-phase ferritic mode also has an important influence on the precipitation of other phases.

The Precipitation of Other Phases

In addition to austenite and ferrite, numerous other phases may occur depending on the thermal history of the steel. Examples of such phases are chromium nitrides, carbides, or carbonitrides; gamma-phase, chi-phase, R-phase, alpha-prime precipitation; γ_2 precipitation in alpha; copper precipitates; and martensite in gamma phase. The occurrence and effect of these phases have been reviewed by Hochmann *et al.*[10] and by Solomon and Devine[3]. Temperature–time precipitation curves for various phases observed in a 2205 type alloy are shown in Fig. 3[5].

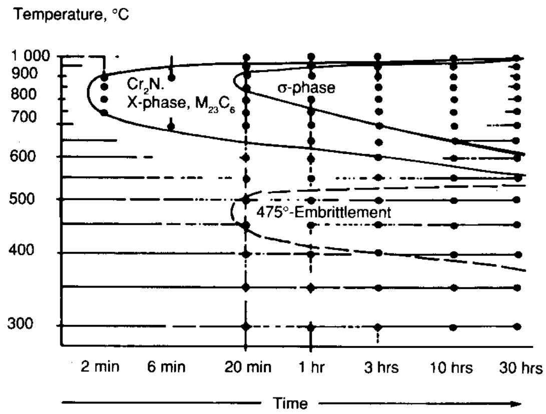
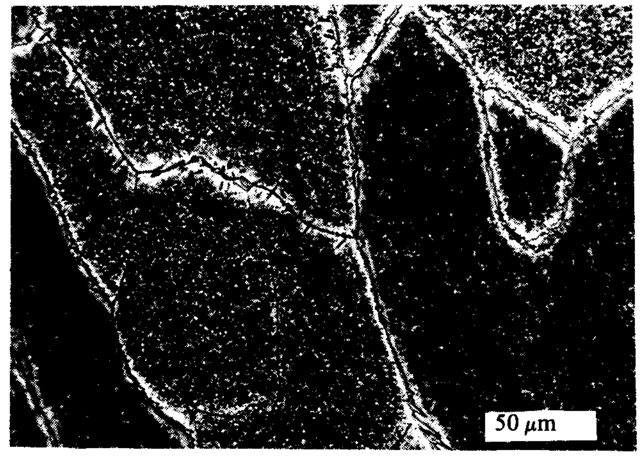


Fig. 3—Time-temperature-precipitation diagram of AF22 solution-annealed for 30 minutes at 1050 °C (water-quenched) and heated as shown[5]

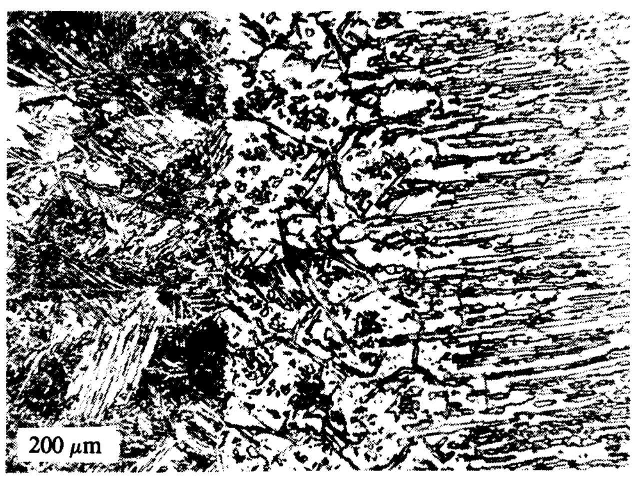
Most of the modern duplex stainless steels contain high levels of nitrogen (0,1 to 0,2 per cent). Nitrogen, an austenite stabilizer, is added as both a solid-solution hardener and to increase resistance to pitting corrosion in chloride-containing media. It has been shown that chromium nitrides precipitate in the ferritic phase when a duplex stainless steel is quenched from very high annealing temperatures[5,11], or when rapid cooling occurs in weld metal or the HAZ of welded duplex stainless steels[12-15].

Herbsleb and Schwaab[5] found fine Cr_2N precipitates in AF22 after being quenched from an annealing temperature of 1100 °C. The amount of Cr_2N precipitation increases with increasing annealing temperature. As the annealing temperature increases, the volume fraction of austenite decreases and the ferrite must take up higher amounts of nitrogen in solid solution. Although these levels of nitrogen are soluble in ferrite at high temperatures, Cr_2N precipitates in the ferrite upon rapid cooling because nitrogen is relatively insoluble in ferrite at lower temperatures. In the case of weldments, the same mechanism is operative, and regions in the HAZ experience temperatures of 1300 °C and higher, at which the steel would be fully ferritic with nitrogen in solid solution. Under certain welding conditions, this region could experience a very rapid cooling cycle and severe Cr_2N precipitation can be observed in an almost fully ferritic HAZ[13]. If the HAZ experiences a slower cooling rate, austenite forms and nitrogen diffuses to and is dissolved in the austenite, thus reducing the amount of Cr_2N precipitation.

Fig. 4 is the microstructure of a laser and manual metal-arc weld on SAF 2205 plate. Laser welding results in a very fast cooling rate, and the microstructure reflects this effect. An almost fully ferritic structure with some precipitation of acicular austenite on the grain boundaries can be observed. Copious precipitation of chromium nitrides within the ferrite grains and a precipitate-free zone near reformed austenite are also evident. The manual metal-arc weld produced much more austenite reformation due to the slower cooling rate.



(a)



(b)

Fig. 4—The microstructures resulting from (a) laser welding and (b) MMA welding of 2205 plate. In (b) the parent metal is on the right, with the weld deposit on the left

Properties of Duplex Stainless Steel Weldments

Pitting Corrosion

Sridhar *et al.* studied the effect of heat input on the pitting corrosion of autogeneously bed-on-plate GTA welded Ferralium 255 in a 10 per cent $FeCl_3.6H_2O$ solution[12,16,17]. The results of this investigation are shown in Table II, and it is evident that the rate of pitting corrosion increases with a decrease in heat input. The sample with the lowest heat input (No. 8) exhibited a very small area of weld metal. Because the base-metal corrosion was negligible, this sample did not have the highest percentage mass loss since the mass loss was limited by the amount of weld metal available for corrosion.

The microstructure of the weld metal indicated that the higher heat input produced a coarser ferrite grain, as well as coarser austenite. The austenite was lathy and had a definite crystallographic orientation with respect to the ferrite. A precipitate-free zone adjacent to the ferrite grain

TABLE II

EFFECT OF HEAT INPUT[12] ON THE PITTING RESISTANCE OF AUTOGENOUSLY GTA WELDMENTS OF FERRALIUM 255

Solution: 10% $FeCl_3.6H_2O$ at 15 °C Time: 120 h

Specimen no.	Heat input kJ/mm	Loss in mass %	Remarks
1	4,4	0	No pitting
2	1,8	0	No pitting
3	1,06	0,20	Slight pitting of weld metal
4	0,91	0,24	Corrosion of weld metal-fusion line
5	0,55	0,42	Corrosion of weld metal-fusion line
6	0,47	0,42	Corrosion of weld metal-fusion line
7	0,24	0,86	Corrosion of weld metal-fusion line
8	0,12	0,63	Corrosion of weld metal-fusion line

boundary was also observed[16]. The corrosive attack occurred mainly inside the ferrite grains around precipitates, and the precipitate-free zone was left intact. This corrosion morphology was explained on the basis of the depletion of chromium around precipitates of chromium nitride in the ferrite.

The effect of the heat input shown in Table II was rationalized in terms of cooling rate. For the low heat inputs, there was a fast cooling rate, leading to significant precipitation of chromium nitride as discussed in the previous section, with resultant high corrosion rates. The beneficial effect of the higher heat inputs resulted from slower cooling rates, which allowed sufficient time for healing (redistribution of chromium) of the depleted zones around the precipitates of chromium nitride. Sridhar *et al.* pointed out that selective attack cannot be explained by partitioning of elements during phase transformation since the ferrite is normally enriched in chromium and molybdenum, two elements that enhance resistance to pitting corrosion[17].

Yasuda *et al.* studied the effect of heat input on the resistance to pitting corrosion of a 2205 alloy[18]. They also found a high heat input to be beneficial for resistance to pitting corrosion. Simulated HAZs were then prepared by the heating of samples of 2205 to various temperatures, followed by quenching in water and in compressed air and by air cooling to yield cooling rates of 300 °C/s, 40 °C/s, and 20 °C/s respectively. The beneficial effect of slower cooling rates on the resistance to pitting corrosion were confirmed on these samples, and selective attack of ferrite was observed for the higher cooling rates. Chromium nitrides were identified in the ferrite phase of the samples cooled at fast and slow rates. The beneficial effect of the slower cooling rates on resistance to pitting corrosion were therefore explained by the healing of chromium-depleted regions around precipitates. Higher austenite volume fractions taking more nitrogen into solid solution with a subsequent decrease in the amount of chromium nitride precipitated in ferrite was also given as possibly contributing to the beneficial effect of slow cooling rates. In terms of weld heat input, it was concluded that heat inputs greater than 1 kJ/mm improved pitting resistance.

Lundqvist *et al.* also studied the effect of welding conditions on the resistance to pitting corrosion of 2205[2,13] and 2304[2] duplex stainless steels. The influence of heat input in the range 0,5 to 3,0 kJ/mm was investigated for TIG welded bead-on-tube welds with and without filler metal. A marked improvement in pitting resistance was obtained with increasing heat input, as is shown in Table III. Only weld beads made with the highest heat input of 3,0 kJ/mm passed the $FeCl_3$ test for duplicate specimens. When evaluated at 25 °C, welds made at 2,0 kJ/mm with filler metal and at 2,5 kJ/mm autogeneously were resistant to pitting. This indicated that, when filler metal is used, lower heat inputs can be tolerated without affecting the resistance to pitting corrosion. The reason for the detrimental effect of low heat inputs was once again found to be an appreciable amount of chromium nitride precipitation in ferrite grains. The amount of precipitation diminished for higher heat inputs and was virtually absent at 3,0 kJ/mm. This was explained by the reformation of austenite at the expense of nitride precipitates.

TABLE III

PITTING TESTS ON TIG WELDED BEAD-ON-TUBE WELDS OF 2205

$x/2$ = SPECIMENS ATTACKED/SPECIMENS TESTED[13]

Solution: 10% $FeCl_3.6H_2O$

Filler metal	Temp. °C	Heat input, kJ/mm 0,5	1,0	1,5	2,0	2,5	3,0
Sandvik 22.8.3.L	25	–	2/2	2/2	0/2	0/2	0/2
	30	–	2/2	2/2	2/2	1/2	0/2
None	25	2/2	2/2	2/2	2/2	0/2	0/2
	30	2/2	2/2	2/2	2/2	1/2	0/2

Lundqvist *et al.*, in addition to TIG welding, performed SMA butt welding on 20 mm thick 2205 plate using heat inputs from 2,0 to 6,0 kJ/mm. Although all the top surfaces of the weld metal passed the pitting test in 10 per cent $FeCl_3$. $6H_2O$ at 30 °C irrespective of heat input, the weld metal on the root side, which was the first to be deposited, did not pass this test. To investigate this, tests on critical pitting temperatures were carried out in 3 per cent NaCl at 400 mV (SCE). Critical pitting temperatures of 48, 43, and 40 °C were obtained for heat inputs of 2,0, 4,0, and 6,0 kJ/mm respectively. Microstructural evaluation revealed that extremely fine austenite had been precipitated in the first- and second-weld beads. The higher the heat input during subsequent weld passes, the more reformed austenite was present. It was thus concluded that not only nitrides give rise to negative effects on the pitting resistance, but also fine precipitates of austenite, which were presumed to be reformed at temperatures as low as 800 °C. However, it was thought that the reformed austenite is less detrimental to pitting resistance than Cr_2N precipitates.

The use of N_2 as a shielding gas was also investigated in this study[2]. Although it was found that nitrogen gives adequate protection to oxidation, it also diffuses into the weld metal and HAZ. This increased the amount of austenite in the root run and effectively suppressed the amount of chromium nitride precipitated, and it was claimed that this would enhance the resistance to pitting corrosion.

A similar study on the effect of welding conditions was carried out by Ume *et al.* on a 2205 type alloy[15]. TIG welding with 9 per cent nickel filler metal using heat inputs of 0,6 to 6,2 kJ/mm was performed, and the resistance to pitting corrosion was evaluated in 10 per cent $FeCl_3.6H_2O$. Like previous investigators, they found the pitting resistance to increase with increasing heat inputs; however, excessive heat inputs once again decreased the pitting resistance. Both types of degradation were found to be related to the precipitation of chromium nitrides. For fast cooling rates (low heat input), precipitation occurred at α/γ boundaries near the fusion line, and pitting was initiated at this position. For slow cooling rates (high heat inputs), precipitates formed at the α/γ boundaries in the HAZ at a distance of about 3 mm from the fusion line, and pitting was initiated at this position. Pits were observed to propagate into the ferrite matrix. In the case of intermediate cooling rates, the superior resistance to pitting was attributed to the healing of chromium-depleted regions.

The inferior resistance of welds of low heat input has been shown to be due to the presence of nitrogen in the steel. Tsuge *et al.*[19] studied the effect of nitrogen content on the localized corrosion behaviour of 2205 simulated weldments. Samples containing 0,047 to 0,238 per cent nitrogen were heated to 1375 °C, and the cooling rate was such that cooling through the range 800 to 500 °C took 10 seconds. Alloys containing less than 0,077 per cent nitrogen had a fully ferritic microstructure, whereas alloys containing more nitrogen exhibited precipitation of acicular austenite. Pitting tests in 5 per cent NaCl and in 10 per cent $FeCl_3.6H_2O$ showed that the low-nitrogen alloys had an inferior resistance to pitting corrosion. This was attributed to the formation of chromium nitride and chromium carbide precipitates on α/α grain boundaries in these fully ferritic structures. This result shows that, for a constant cooling rate, low-nitrogen duplex stainless steels are more susceptible to the precipitation of chromium nitride than high-nitrogen alloys. Tsuge *et al.*[19] concluded that 2205 should contain at least 0,12 per cent nitrogen to allow sufficient reformation of austenite in the weldments. This explains why modern or second-generation duplex stainless steels are deliberately alloyed with nitrogen as opposed to the older grades (e.g. AISI329), which contained residual nitrogen.

Intergranular Corrosion

Intergranular corrosion due to the precipitation of chromium carbides at grain boundaries can be a serious problem in austenitic stainless steels if these materials are held for prolonged periods in the sensitization region (approximately 500 to 750 °C). This form of corrosion results from the depletion of chromium adjacent to $M_{23}C_6$ precipitates. One way of combating intergranular corrosion is for stainless steels to contain very little carbon (typically less than 0,03 per cent). The low carbon content retards the kinetics of $M_{23}C_6$ precipitation, thus allowing longer times within the sensitization temperature range without any detrimental effect.

Duplex stainless steels exhibit a remarkable resistance to sensitization. This phenomenon has been reviewed by Solomon and Devine[3], and a model for the superior resistance of duplex stainless steels to sensitization has been proposed. This model is based on the chromium concentration profile at an austenite-ferrite interface containing $M_{23}C_6$, as depicted in Fig. 5. Because of the high chromium content in the ferrite phase, as well as the high diffusivity of chromium in ferrite, most of the chromium in the carbide is contributed by the ferrite phase. Consequently, a very wide chromium-depleted zone exists on the ferrite side of the interface. A very small amount of chromium is contributed by the austenite phase. This results in the formation of a very narrow but deep chromium-depleted zone on the austenite side of the interface. Because the zone is so narrow, it is quickly replenished by chromium diffused from the interior of the austenite grain. The narrowness and rapid healing of this depleted zone explain the superior resistance of duplex alloys to sensitization. Despite this inherent resistance, many commercial duplex stainless steels contain less than 0,03 per cent carbon, thus reducing the risk of sensitization even further.

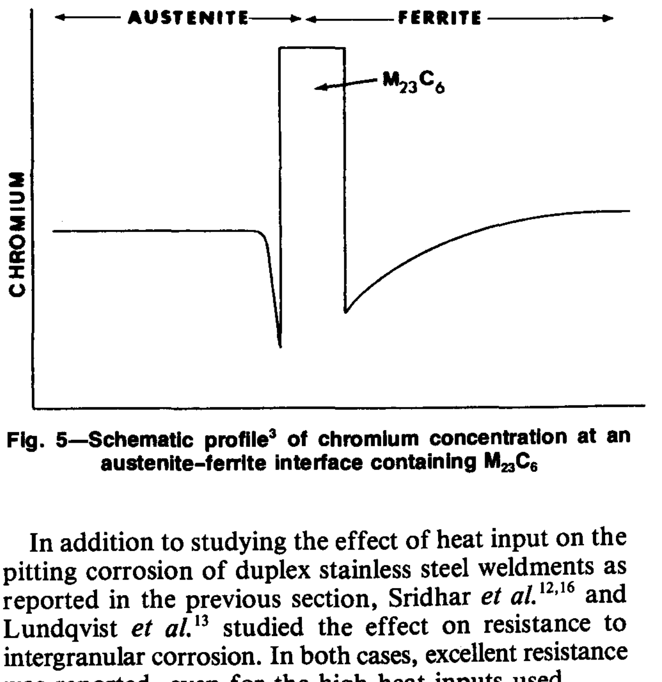


Fig. 5—Schematic profile[3] of chromium concentration at an austenite-ferrite interface containing $M_{23}C_6$

In addition to studying the effect of heat input on the pitting corrosion of duplex stainless steel weldments as reported in the previous section, Sridhar *et al.*[12,16] and Lundqvist *et al.*[13] studied the effect on resistance to intergranular corrosion. In both cases, excellent resistance was reported, even for the high heat inputs used.

Stress-corrosion Cracking

Duplex stainless steels have very good resistance to stress-corrosion cracking. Gooch[20] reported that no marked adverse influence of welding cycle on the cracking propensity of low- and high-alloy duplex stainless steels could be detected in $CaCl_2$ solution at 130 °C. In the work by Lundqvist *et al.*[2,13], it was found that welds had a slightly inferior resistance in comparison with the parent metal in 40 per cent $CaCl_2$, but this was considered negligible from a practical viewpoint. The heat input was found not to influence the susceptibility to stress-corrosion cracking.

Duplex stainless steels are being used increasingly in offshore oil exploration to combat sulphide stress-corrosion cracking (SSCC) in sour-gas wells. Wilhelm and

Kane[21] found that the susceptibility of 2205 to SSCC increased after the alloy had been heat-treated at 1315 °C followed by a water quench. This heat treatment resulted in an increase in the volume fraction of ferrite, and the increased susceptibility to SSCC was presumed to be due to the greater sensitivity of the ferrite phase to hydrogen. From this result, it can be expected that low heat inputs may have a detrimental effect on the resistance of duplex stainless steels to SSCC.

Mechanical Properties

It is well known that the ferrite content of a duplex stainless steel has a significant influence on the yield strength, tensile ductility, and toughness of the alloy[3]. For this reason, it can also be expected that the heat input will affect these properties in a welded joint since the austenite-to-ferrite balance depends on the cooling rate. Sridhar *et al.*[12] found that increased austenite contents in welds on Ferralium 255 resulted in increased tensile elongation and impact toughness. Lundqvist *et al.*[13] found an increasing toughness as the heat input was increased on SAF 2205. This was also attributed to a reduced ferrite content. These authors also point out that, owing to well-balanced chemical compositions and favourable austenite-to-ferrite ratios, sigma phase and 475 °C embrittlement are not a problem during the welding of modern duplex stainless steels. They concluded that, from a strength and toughness point of view, there seems to be no need to maximize arc energies for 2205.

Discussion

Owing to the different solidification modes and transformation characteristics of duplex and austenitic stainless steels, weld-heat input has different effects on the properties of these two types of alloys. The most significant effect is found in the corrosion properties. For austenitic stainless steels, it is desirable to minimize the heat input in order to have a fast cooling rate in the weld metal and adjoining HAZ. This ensures that deleterious phases such as $M_{23}C_6$ do not precipitate during the welding process. In contrast, it has been shown that very low heat inputs could have disastrous effects on the resistance of duplex stainless steel weldments to pitting corrosion.

There is general agreement that very low heat inputs should be avoided on duplex stainless steels and that much higher heat inputs than for austenitics can be tolerated. This could influence the costs of weld fabrication since joints can be made in fewer passes. It should be kept in mind that heat input alone does not determine the cooling rate, but that the thickness of the parent metal and the interpass temperatures also play a role. Lundqvist *et al.*[2] specified heat inputs of 0,5 to 2,5 kJ/mm for SAF 2304 and SAF 2205, with the upper limit not considered critical. It was stated that the choice of heat input should be related to the thickness of the material, thick material requiring heat inputs in the upper part of the range in order to allow a sufficient amount of austenite to reform. It is not the aim of this paper to specify welding conditions for duplex stainless steels, but ràther to draw attention to the difference in transformation behaviour of austenitic and duplex stainless steels and how this is affected by heat input. Detailed welding specifications are available from the producers of duplex stainless steels.

In terms of resistance to intergranular corrosion and stress-corrosion cracking, no detrimental effect was found for high heat inputs. The tensile elongation and toughness of welds on duplex stainless steels were found to increase with increasing heat input.

The only detrimental effect of very high heat inputs or multipass welding reported was the formation of secondary austenite[13] and the formation of chromium nitrides in the HAZ[15]. It is interesting to note that no detrimental effect due to the precipitation of sigma phase was reported.

Conclusions

(1) Duplex stainless steels solidify in the single-phase ferritic mode, whereas austenitic stainless steels solidify in the austenitic or austenitic–ferritic mode.
(2) The austenite phase in weldments of duplex stainless steel is formed by a solid-state transformation that is strongly affected by cooling rate.
(3) In duplex weldments, low heat inputs result in high volume fractions of ferrite and severe precipitation of chromium nitrides, adversely affecting mechanical and corrosion properties.
(4) Duplex stainless steels should be welded by the use of high heat inputs to allow sufficient time at high temperature for austenite to reform.

Acknowledgements

The authors are grateful to the University of the Witwatersrand and to Middelburg Steel & Alloys (Pty) Ltd for permission to publish this paper.

References

1. Hoffmann, J.P. *J. S. Afr. Inst. Min. Metall.*, vol. 86, no. 11. 1986. pp. 433–442.
2. Lundqvist, B., Norberg, P., and Nyström, R. Weldability and weld joint properties of duplex stainless steels. ASM Materials Week '87, Oct. 1987, Cincinnati (USA).
3. Solomon, H.D., and Devine, T.M. Jr. *Duplex stainless steels.* Lula, R.A. (ed.). Metals Park (USA), ASM, 1983. pp. 693–756.
4. Pugh, J.W., and Nisbet, J.O. *Trans. AIME*, vol. 188. 1960. pp. 268–276.
5. Herbsleb, G., and Schwaab, P. *Duplex stainless steels.* Lula, R.A. (ed.). Metals Park (USA), ASM, 1983. pp. 15–40.
6. Suutala, N., Takalo, T., and Moisio, T. *Met. Trans. A*, vol. 10A. 1980. p. 1173.
7. *Ibid.*, 1979. p. 1183.
8. *Ibid.*, vol. 11A. 1980. p. 717.
9. Schaeffler, A.L. *Metals Progress*, vol. 56. 1949. p. 680.
10. Hochmann, J., Desestret, A., Jolly, P., and Mayoud, R. *Stress corrosion cracking and hydrogen embrittlement of iron base alloys.* Conference held at Unieux-Firming, France, Jun. 1973. NACE-5, 1977. p. 956.
11. Sridhar, N., and Kolts, J. *Corrosion*, vol. 43, no. 11. 1987. p. 646.
12. Sridhar, N., Flasche, L.H., and Kolts, J. *Corrosion '84.* NACE, 1984. Paper no. 244.
13. Lundqvist, B., Norberg, P., and Olsson, K. *International Conference on Duplex Stainless Steel, The Hague, Oct. 1986.* Nederlands Instituut voor Lastechniek, 1986. Paper 10, pp. 16–29.
14. Hertzman, S., Roberts, W. and Lindenmo, M. *Ibid.*, Paper 30, pp. 257–267.
15. Ume, K., Seki, N., Naganawa, Y., Hyodo, T., Satoh, K., and Kuriki, Y. *Mat. Perf.*, Aug. 1987. pp. 25–31.
16. Sridhar, N., Flasche, L.H., and Kolts, J. *New developments in stainless steel technology.* Lula, R.A. (ed.). Metals Park (USA), ASM, 1985. pp. 341–350.

17. Sridhar, N., Kolts, J., and Flasche, L.H. *J. of Metals,* Mar. 1985. pp. 31-35.
18. Yasuda, K., Tamaki, K., Nakano, S., Kobayashi, K., and Nishiyama, N. *ZIS Mitt.,* vol. 29, no. 3. 1987. pp. 229-237.
19. Tsuge, H., Tarutani, Y., and Kudo, T. *Corrosion,* vol. 44, no. 5. 1988. pp. 305-314.
20. Gooch, T.G. *Duplex stainless steels.* Lula R.A. (ed.). Metals Park (USA), ASM, 1983. p. 573.
21. Wilhelm, S. Mark, and Kane, R.D. *Corrosion,* vol. 40, no. 8. 1984. pp. 431-439.

Video on coal

The World Coal Institute has produced the world coal industry's first international video, 'Coal—fuel for the future'. The 8-minute video is directed to non-technical audiences. It presents a lively and comprehensive background to one of the world's greatest fuel resources and one of the world's most international industries.

Public-interest groups, students, and anyone concerned with energy and environmental issues will find the video a short but very informative introduction to world coal. The commentary is being translated into several languages so that the video can be used worldwide.

Coal is used to provide 44 per cent of the world's electricity, and a substantial increase in the use of coal is being forecast for the 1990s and beyond—especially in the developing world. The video makes the point that coal can power the future cleanly, safely, and economically.

Filming took the production team to 4 continents, and the final product contains footage from 7 different countries. It highlights the facts that coal reserves are vast, easily accessible, and spread generously around the world. The film also makes clear that coal is the safest fuel to transport, store, and use, while remaining the cheapest fuel for the large-scale generation of electricity.

Given current international concern about the greenhouse effect and global warming, the video makes a timely contribution to the debate. It explains the success that has been achieved in the development of clean coal-burning technology. A modern coal plant is designed to eliminate 90 to 95 per cent of the sulphur dioxide, and much of the nitrogen oxides, emitted from burning coal, as well as substantially improving the efficiency with which fuel is used.

The video points out that all the world's coal-fired power stations together emit only around 8 per cent of the total man-made radiative gases that contribute to the greenhouse effect. (CFCs and deforestation contribute a much greater share of such gases.)

For further information please contact,

Mr Richard Tallboys
Chief Executive
World Coal Institute
Vicarage House
58 – 60 Kensington Church Street
London W8 4DB
England.
Telephone: + 44 1 937 4600
Facsimile: + 44 1 376 0454.

Environment Day Symposium

The Sixth World Environment Day Symposium will be held under the auspices of the Associated Scientific and Technical Societies of South Africa (AS&TS) with the Environmental Planning Professions Interdisciplinary Committee (EPPIC), the Habitat Council, and the Society of Professional Engineers as participating organizations.

The Symposium has a twofold aim: firstly, to make engineers and other members of the planning professions aware of their responsibility to the environment; and, secondly, to make the public aware of what is being done by these professions.

Enquiries should to be addressed to

Mrs T.Y. Poole
Secretary
Symposium Organizing Committee
P.O. Box 61019
Marshalltown
2107.
Telephone: 832-2177 (mornings).

New building for Wits*

The Faculty of Engineering's new Chamber of Mines Building on the University of the Witwatersrand's West Campus was opened recently.

The construction of the new building was made possible with the assistance of the mining industry. Comprising 5810 m^2 of modern teaching and research facilities on one basement level and four floors, it currently houses the Departments of Mining Engineering and Electrical Engineering, the Faculty Office, and the offices of Engineering Support Programmes and Continuing Engineering Education.

At the opening function, Professor Glasser said the Engineering Faculty aimed to find integrated solutions to the country's engineering requirements. This represented a change from the discipline-orientated divisional approach that had existed in the past. He said this aim would be facilitated when the whole faculty was eventually housed on the West Campus, with all its staff in the Chamber of Mines Building.

* Released by Lynne Hancock Communications, (P.O. Box 1564, Parklands 2121).

Pictured at the official opening of the Chamber of Mines Building were (from left) Mr K. Maxwell, President of the Chamber of Mines; Mr C. Fenton, Immediate Past President of the Chamber of Mines; Professor D. Glasser, Dean of the Faculty, and Professor P. Tyson, acting Vice-Chancellor of Wits University.