

HAZ Effects in Hot-Rolled Dual-Phase Steel During Flash Butt Welding of Wheel Rims

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Abstract: The embrittlement and softening behaviour in simulated heat-affected zones (HAZ) of a newly-designed dual phase DP680 steel for wheel rim applications with different flash allowances were investigated to determine weldability, and offer valuable information for the steel design and its subsequent flash butt welding (FBW). The characterisation of microstructure and mechanical performance for the simulated HAZ was conducted by means of optical microscopy, scanning electron microscopy, electron backscatter diffraction, hardness distribution and Charpy V-notch (CVN) values at selected temperatures. The investigation demonstrates that the toughness of coarse-grained HAZ was kept at an average level of 25.3 J when the prior austenite grain size was controlled to 60.54 μm at a flash allowance of 14 mm (equivalent to heat input of 15.14 kJ/cm based on real welding process), which exhibits the worst toughness when the flash allowance was changed from 4 mm to 14 mm. Further, with a higher martensite fraction (>30%) in base material (BM), the softening occurs in inter-critical HAZ (ICHAZ) instead of sub-critical HAZ since most of martensite in ICHAZ has decomposed, and the rest ferrite and newly-formed bainite with remaining martensite reduce the hardness to a larger extent compared to SCHAZ, whose martensite has only partly decomposed. Even if the softening degree is up to 21.6% compared to the BM (average 233 HV_{0.5}), the work hardening during a series of forming processes after FBW has alleviated the softening evidently (work hardening degree>10%). However, the failure location is still in ICHAZ after forming extension which has been confirmed in practical applications of DP680 FBW and subsequent forming processes.

Key words: Dual-phase steel, flash butt welding, weldability, HAZ embrittlement, HAZ softening

1. INTRODUCTION

Dual phase (DP) steels which consist of martensite in ferrite matrix belong to the family of advanced high strength steels that offer improved mechanical properties such as high strength, continuous yielding, low yield/tensile ratio and high initial hardening rate, making them suitable

for automotive applications [1]. Ultimate mechanical properties of the designed DP steels are primarily dependent on the volume fraction, morphology, size, distribution and performance of the constituent phases. The balanced phases of designed soft ferritic matrix and islands of hard martensite are the key reason to obtain the final properties of DP steels. In general, in the design of the DP steels with high strength, the ferrite fraction is from 70% to 90%, and the remaining phase is martensite [2]. However, when the DP steel deforms, the strain is concentrated in the lower-strength ferrite phase that surrounds the islands of martensite, which generates the unique high work-hardening rate [3].

As an important part of resistance welding technology, the flash butt welding (FBW) is a solid-state welding method that relies on high current to produce a solid phase weld joint between two welding ends [4]. During the flashing stage, effective metallic vapor flow generated at the welding interface prevents oxidation and other contamination. Subsequent upsetting deformation leads to the formation of a plastic zone in the welds. Also, a large enough upsetting force is used to expel the liquid metal and oxide inclusions [5]. The heat-affected zone (HAZ) is regarded as the weakest part in the obtained weld joint, and determines the overall performance of the target steels [6]. The optimised FBW parameters are key to make sure the subsequent forming process after welding [7,8]. Although several factors can influence HAZ performance, the intrinsic properties of the steel itself and the heat input (HI) during welding can affect the HAZ directly. Further, the higher strength the DP steel has, more metallurgical and mechanical variation of characteristics occur in the HAZ [9]. Understanding the variations of phase transformation in the HAZ is therefore necessary to widen the processing windows for acceptable welds [10].

The poor HAZ toughness is considered to be one of the major issues that would limit the application of the high-grade DP steel in the application of wheel rims. Currently, the highest grade applied in wheel rim is DP590. No information is available for DP680, a higher grade, to be applied as a wheel rim. In other words, the application of DP680 in wheel rim cannot be found from any references. It is, thus, important to evaluate the weldability and develop effective techniques for improving the HAZ toughness of these high-grade wheel steels. As the first step for the wheel application, the toughness in the HAZ, i.e. coarse-grained HAZ (CGHAZ), has to be evaluated for comprehensive understanding of the DP steel.

Besides toughness, HAZ softening is unavoidable in DP steels. Biro *et al.* [11] found that for two DP600 and DP780 materials, HAZ softening occurs generally in the tempered or sub-critical HAZ (SCHAZ) where the heating temperature is experienced during welding does not exceed the A_{c1} temperature. The HAZ softening by martensite tempering is a function of cementite nucleation. Xia *et al.* [12] showed that HAZ softening and the size of the soft zone increase with increased strength of the steel. Jha *et al.* [13] concluded that the degree of HAZ softening in a dual-phase steel is also closely related to the chemical compositions of the steel and that the resistance to softening increases with increasing alloy content. In addition, the heat that is required for HAZ softening in the DP steel decreases with increased martensite carbon content in the steel. The existence of carbide-forming elements, such as Cr and Mo [11,14], in a DP steel can promote the resistance to softening. Further, the degree of softening in the welding HAZ increases with increased HI. During the FBW process, the parameter that mainly determines the HI is the flash

allowance (FA). FBW is an electrical resistance welding process used for joining components, where the energy transfer is provided primarily by the resistance heat from the parts. The parts to be joined are clamped and brought together slowly while a flashing voltage is applied. When small contacts are made between the components, there is a high current density and the material resistance heats, melts and blows out of the joint in a shower of melted particles, giving the characteristic flashing action, which is a key to set the FA. This flashing progressively introduces heating and a softened zone at the component end, while eliminating oxides and contaminants from the interface. After a preset ‘burn-off’ length of the parent material, a forge force is applied to the parts to consolidate the joint, which is the upsetting process.

In the application of high strength steels as automotive components, in particular wheel rim, there are the problems of the HAZ brittleness and softening near weld interface [15]. In order to establish the correlation between FA and weldability, the simulated CGHAZs subjected to different thermal cycles were produced to investigate the influence of HI (can be converted to FA). The present study seeks to characterise the kinetics of HAZ coarsening, tempering and phase transformation by conducting a series of thermal cycles which can simulate the flashing stage in the FBW process. Ultimately, the produced wheel rim was analysed to confirm the influence of HAZ obtained.

II. EXPERIMENTAL PROCEDURE

A. Material

The material used in this investigation was a novel hot-rolled DP steel DP680. The thickness of the hot-rolled plate was 11.0 mm. The chemical composition of the studied DP680 grade is in the list of lean DP steels, which saves the alloying cost due to a reduction of Cr and lack of Mo. The coiling temperature is controlled to be higher than 400 °C to keep the strength of the parent material. Meanwhile, the alloy design balances the microstructure and carbon equivalence to keep the weldability, embrittlement and softening in the HAZs.

The designed DP680 was alloyed with C, Mn, Si, Cr and Nb, listed in Table I. The chemical composition design and tensile tests were conducted according to GB/T 20887.3 (Continuously hot rolled high strength steel sheet and strip for automobile -- part 3: Dual phase steel). The carbon equivalent (CE) can be calculated by the following equations:

$$CE_{PCM} = C + \frac{Si}{30} + \frac{Mn+Cu+Cr}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B \quad (1)$$

$$CE_{IIW} = C + \frac{Mn}{6} + \frac{Cr+Mo+V}{5} + \frac{Ni+Cu}{15} \quad (2)$$

The fraction of martensite and solute carbon during production can be controlled by cooling rates and coiling temperature below 681 °C based on Thermo-Calc calculations, seen in Figure 1. Still from the calculation, it is clear that the A_{c1} and A_{c3} from the calculation are 681 °C and 829 °C, respectively, which can be referred to in the dilatometry tests. To further improve strength, the DP680 steel has a finer and more deformed microstructure, together with more martensite in the ferrite matrix compared to a lower grade, for example DP590. From Figure 2, the microstructure under optical microscope (OM) and scanning electron microscope (SEM) show that the fraction of martensite is approximate 33.88 vol.% according to the Image J software. Based on the optimised

design, a high tensile strength and low yield/tensile ratio (Table II), which contribute to formability and lightweighting [16], can be obtained for this novel and cost-saving DP680 grade.

Table I. Chemical composition of DP680 (wt.%)

Element	C	Si	Mn	Nb	Cr	Fe	CE _{Pcm}	CE _{IIW}
DP680	0.07	0.1	1.5	0.035	0.25	Bal.	0.161	0.37
Standard	0.23	2.00	3.30	0.15	1.00	Bal.	0.51	0.98

CE: carbon equivalent.

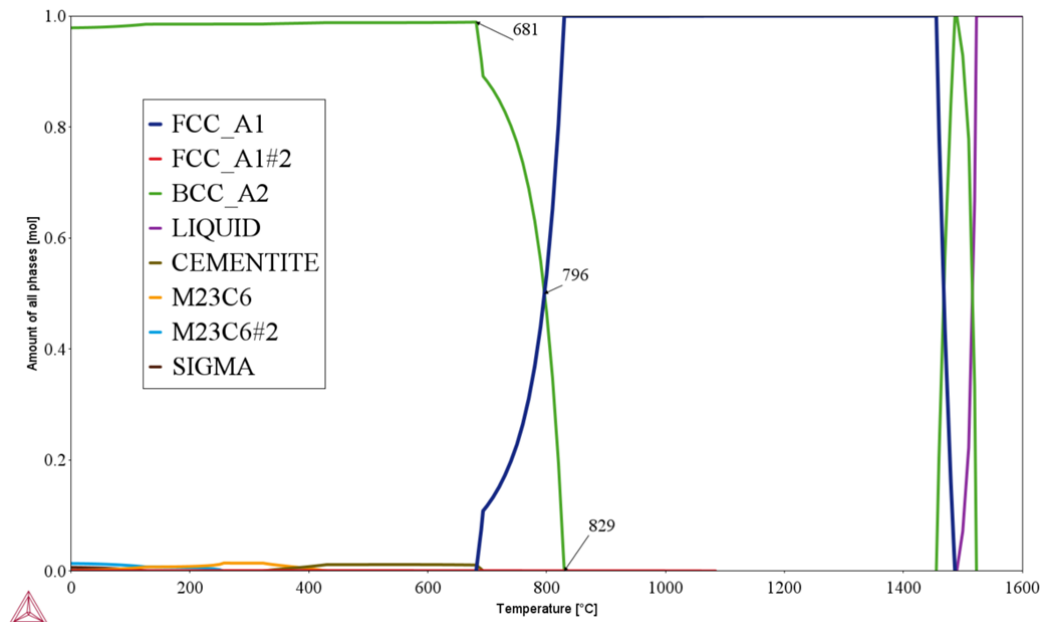


Fig. 1 Phase fraction in DP680 as a function of temperature.

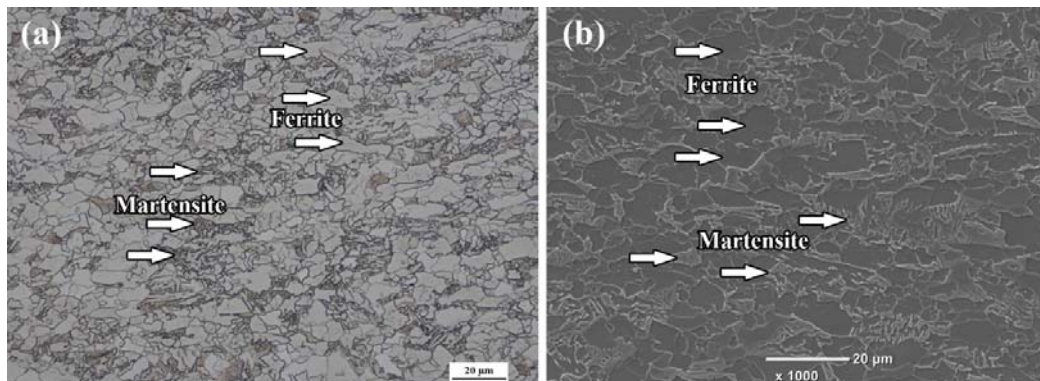


Fig. 2 Microstructure of DP680: (a) under OM ($\times 500$) ; (b) under SEM ($\times 1000$).

Table II. Tensile properties of DP680

Materials	YS/MPa	UTS/MPa	Y/T ratio	EL/%
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DP680/TD	418±7	734±6	0.569	26.9±0.1
DP680/RD	425±1	737±2	0.577	31.9±1.7
Standard	380-500	680-800	-	≥19

TD: transverse direction, RD: rolling direction, YS: yield strength, UTS: ultimate tensile strength, Y/T: yield strength/tensile strength, EL: elongation.

B. Simulated HAZ

In order to evaluate the HAZ microstructure and properties of the DP680 steel, the square bar specimens (10.5 mm×10.5 mm×75 mm) extracted from the plate along the transverse direction were subjected to weld thermal cycles with the parameters from the real FBW. Simulated CGHAZs were produced in a vacuum chamber of the Gleeble 3500 simulator at University of Wollongong, and the adjacent HAZs, i.e. FGHAZ (fine-grained HAZ), ICHAZ (inter-critical HAZ) and SCHAZ, were classified based on the peak temperature ranges. In this case, the flash allowance values, including 14 mm, 10 mm and 4 mm, are selected based on measured real weld thermal cycles [17] and the real situation at the customer sites. Among them, 14 mm and 4 mm are the values that most customers selected as two sides of the flash allowance range. And 10 mm is in the middle of the range. As listed in Table III and displayed in Figure 3, the thermal cycles for the CGHAZ simulation are composed of three processes, they are: heating to the peak temperature of 1,380 °C at a matched heating rate for each flash allowance; holding at the peak temperature for 0.5 s; and controlled cooling through $\Delta t_{8/5}$ (from 800 °C to 500 °C). Based on these data, the HIs calculated are 15.14 kJ/cm, 12.37 kJ/cm and 8.74 kJ/cm for the three selected flash allowances. It should be noted that since overshooting occurs when the heating rate is high, the peak temperatures were revised according to the overshooting behaviour (20 °C for the heating rate of 150 °C/s, 80 °C for 350 °C/s, and 240 °C for 500 °C/s). The red and black curves in Figure 3 are the set and real Temperature-Time curves captured from the simulation. After simulation, three samples were named based on the flash allowances used, i.e. FA14, FA10 and FA4.

Table III. Key parameters for Gleeble simulation

Sample No.	Simulated flash allowance (mm)	Heating rate (°C/s)	$\Delta t_{8/5}$ (s)	Peak temperature (°C)
FA14	14	150	30	1,380
FA10	10	350	20	1,380
FA4	4	500	10	1,380

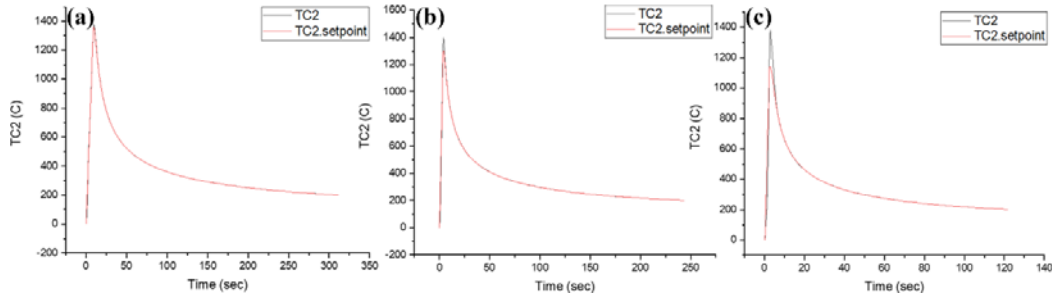


Fig. 3 Thermal cycles used for CGHAZ simulation resulting from different HIs by variation of flash allowance: (a) FA14; (b) FA10; (c) FA4.

C. Microstructural and mechanical experiments

The simulated HAZ samples for metallographic examination were sectioned at the location of the thermocouples, ensuring that the examined samples were subjected to required thermal cycles. The specimens for microstructural observation were cut using a blade, mounted in bakelite for handling, polished in a standard metallographic method by Struers automatic polisher TegraPol 21. The polishing process was conducted according to ASTM MNL 46. The specimens were ground by 220# SiC paper, roughly polished with a DP-Plan cloth and 9 μm diamond spray, finely polished with a MD-Mol cloth and 3 μm diamond spray, a MD-Dur cloth and 1 μm diamond spray in turn. Subsequently, the specimens were etched in 5 pct nital solution. The microstructures were characterised using OM and SEM. Further, the grain size was measured using the software of Image J. Electron backscatter diffraction (EBSD) study of selected samples was conducted with a TESCAN MAIA3 machine, and a step size of 0.3 μm was used for the CGHAZ. The software of AZtecCrystal was used to analyse the obtained data.

The Vickers hardness testing was performed using a load of 0.5 kg and a dwell time of 10 s according to ASTM E384, and the corresponding instrument is Matsuzawa Via-F microhardness tester. The Charpy-V impact toughness was investigated by the machine of Instron-9250. Three parallel Charpy-V notch (CVN) specimens with a size of 10 mm $\times$ 10 mm $\times$ 55 mm with the T-L orientation have been taken according to ASTM E23, and a 2 mm deep CVN was cut for the impact testing at 20 $^{\circ}\text{C}$ and -20 $^{\circ}\text{C}$. The impact measurements were conducted based on ASTM A370. The impact energies were recorded during CVN testing.

III. RESULTS

A. Transformation Temperature

When the DP680 steel is heated during welding, four HAZ areas can be classified as CGHAZ, FGHAZ, ICHAZ and SCHAZ as mentioned. The classification of these four areas depends on the peak temperature. Combined with the results from the dilatometry experiments (Figure 4), the A_{c1} and A_{c3} for DP680 are 734 $^{\circ}\text{C}$ and 887 $^{\circ}\text{C}$, respectively, which means that the peak temperature of ICHAZ is located between 734 ~ 887 $^{\circ}\text{C}$, and the peak temperature of SCHAZ is below 734 $^{\circ}\text{C}$. CGHAZ has a higher peak temperature, i.e. higher than 1,100 $^{\circ}\text{C}$. And FGHAZ has a peak temperature range of A_{c3} to 1,100 $^{\circ}\text{C}$. It is understandable that all four HAZ areas can be positioned according to the microstructure characteristics even if it is difficult to differentiate the HAZ areas relying only on the temperature. The widths of ICHAZ and SCHAZ for three flash

allowances are displayed in Figure 5. It is evident that the increased flash allowance widens the scopes of ICHAZ and SCHAZ, which may influence the softening behaviours in the practical application. Generally, the wider softening area, together with softening degree, decides the possibility of localised necking during forming.

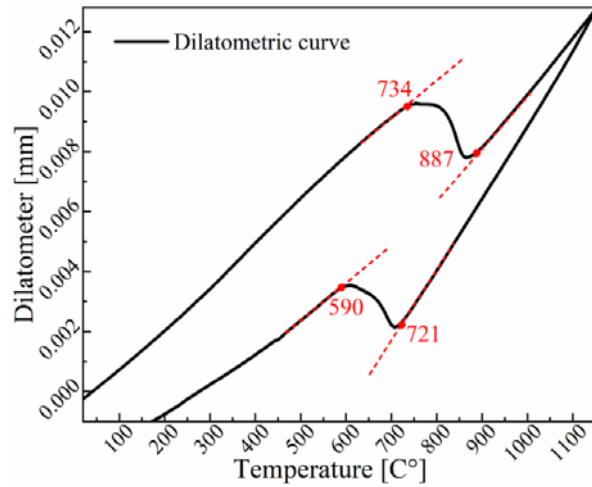


Fig. 4 Phase transformation curves of DP680 when heating and cooling.

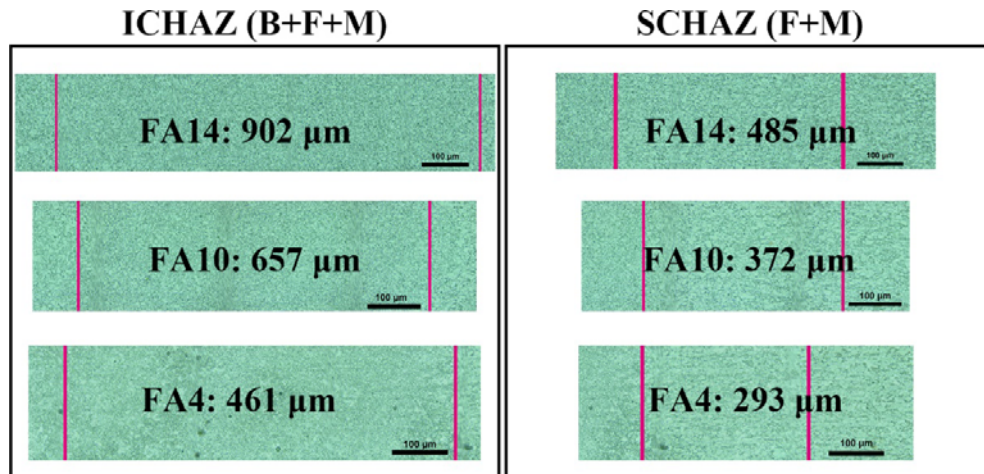


Fig. 5 Widths of ICHAZ and SCHAZ at three flash allowances.

B. Microstructural and Mechanical Analysis

Figure 6 shows the microstructures of the simulated HAZs that have experienced different welding cycles. The original dual phase microstructures in the parent material (Figure 2(a)) were transformed to relevant microstructures located at different HAZ areas. All of the images of different zones were observed by optical microscope under a same magnification rule, i.e. $\times 500$. The phases can be divided into four different types, which are lath martensite (LM), martensite (M), bainite (B) and ferrite (F). From Figures 6(a)~(c)-1 and 6(a)~(c)-2, CGHAZ and FGHAZ have the similar microstructures of lath martensite, bainite and ferrite. With the increase of the HI (from FA4 to FA14), the content of lath martensite decreases, and the content of bainite increases, accompanied with grain growing. Figures 6(a)~(c)-3 display the microstructure in the ICHAZ,

228 which includes bainite, ferrite and martensite. Since ICHAZ is an area where the peak temperature
 229 is in the range of $A_{c1} \sim A_{c3}$, part of ferrite and martensite was transformed to austenite, and then
 230 new phases, i.e. bainite and ferrite, formed based on varied cooling rates. Figures 6(a)~(c)-3 have
 231 a similar microstructure as described above. However, the grain size at ICHAZ for FA4 is smaller
 232 than FA14 due to fast cooling rate after thermal cycle. SCHAZ is an area where the peak
 233 temperature is below A_{c1} , in this case, the original microstructures of (ferrite + martensite) do not
 234 change except for tempering of the microstructure. It cannot be observed that the microstructure
 235 changes evidently under the OM, seen in Figures 6(a)~(c)-4. However, the microstructure under
 236 SEM will be investigated further to testify the microstructural evolution.

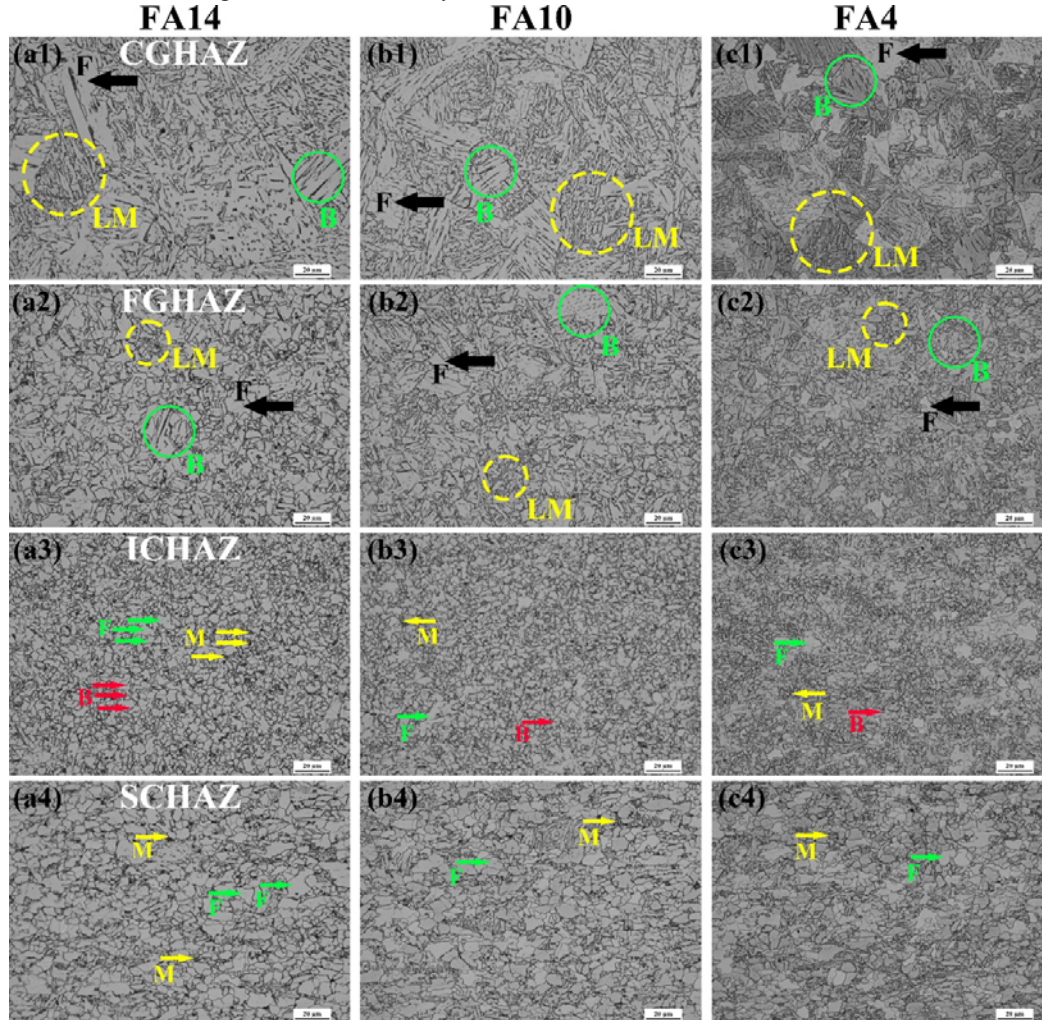


Fig. 6 Microstructures at HAZs: (a1) FA14-CGHAZ; (b1) FA10-CGHAZ; (c1) FA4-CGHAZ; (a2) FA14-FGHAZ; (b2) FA10-FGHAZ; (c2) FA4-FGHAZ; (a3) FA14-ICHAZ; (b3) FA10-ICHAZ; (c3) FA4-ICHAZ; (a4) FA14-SCHAZ; (b4) FA10-SCHAZ; (c4) FA4-SCHAZ.

In order to compare the prior austenite grain sizes (PAGS) or original austenite grain size of CGHAZs for the DP680 steel at roughly three flash allowances, the grain boundaries of three CGHAZ were highlighted, and the average grain sizes ranked as $FA14 > FA10 > FA4$, seen Figures 7(a) and (b). The precise PAGS results will be evaluated based on the EBSD results.

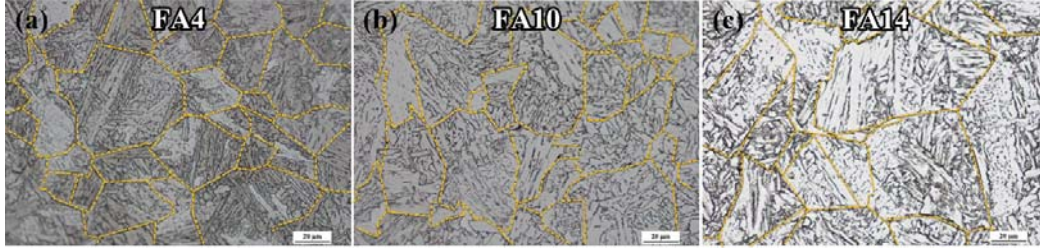


Fig. 7 PAGSs and their distribution of DP680 CGHAZ at three flash allowances: (a) FA4-CGHAZ; (b) FA10-CGHAZ; (c) FA14-CGHAZ.

The impact toughness (at 20 °C) of DP680 CGHAZ with flash allowances is shown in Figure 8 and Table IV. The rank in CVN at 20 °C is FA4>FA10>FA14. Since with higher flash allowance, the CVN at CGHAZ decreases accordingly. However, when the test temperature is -20 °C, the CVN at CGHAZ decreases to a great degree. The CGHAZ CVN values at -20 °C for FA14 are 31 J, 24 J and 21 J, respectively. And the average value is 25.3 J, dropping off evidently compared to the values at 20 °C. Even though, the CVN still meets the requirements of the application. In terms of hardness, FA4, FA10 and FA14 are 227 HV_{0.5}, 209 HV_{0.5} and 191 HV_{0.5}, respectively, which is decreased gradually with increased heat input.

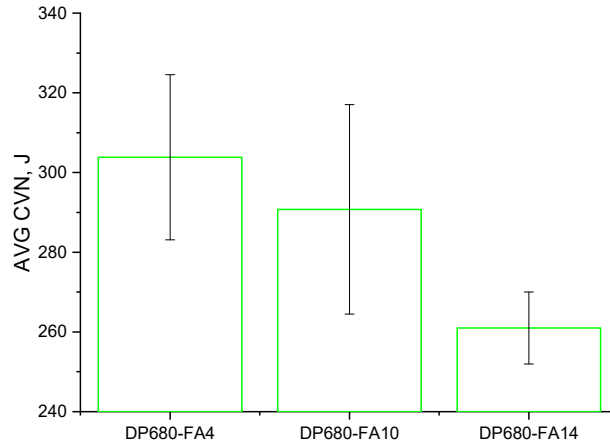


Fig. 8 CVN impact toughness (20 °C) at three flash allowances.

Table IV. CVN impact toughness (20 °C) at three flash allowances.

No.	Single CVN, J			AVG CVN, J
DP680-FA4	312	319	280	304
DP680-FA10	260	304	308	291
DA680-FA14	259	271	253	261

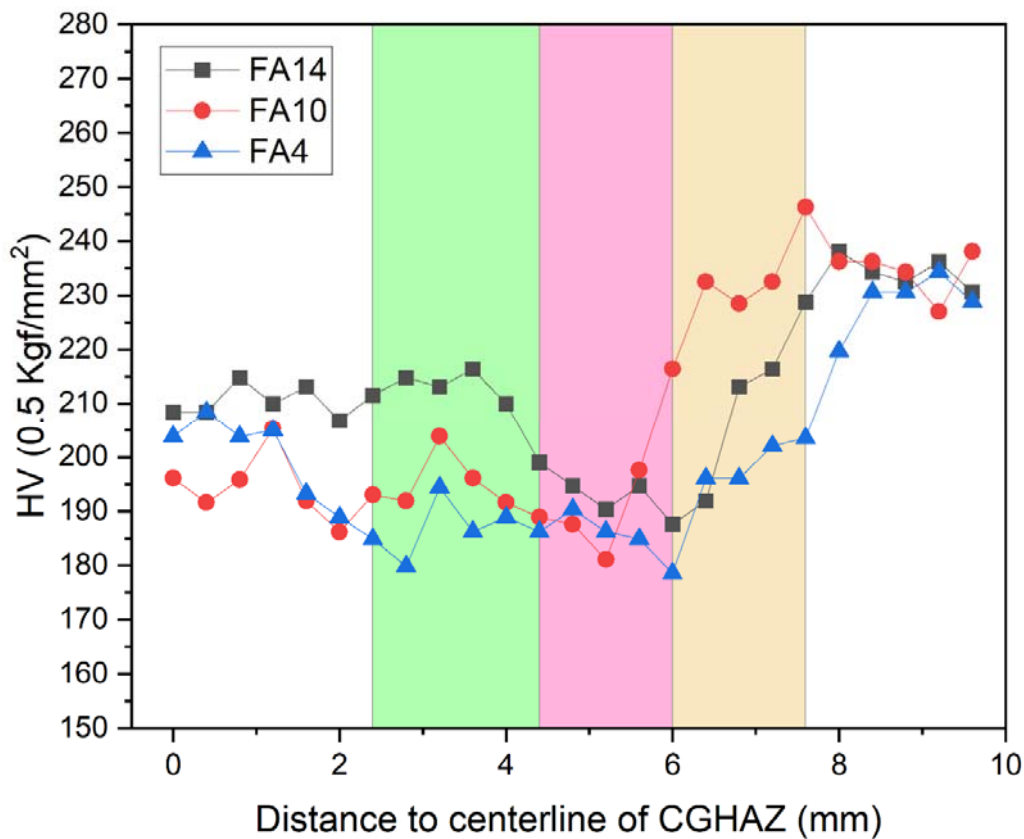
The CGHAZ is a key area to determine the toughness of the steel since the weld joint through FBW has to resist wheel rim impact. Besides impact resistance, the softening of the steel is

another concern for designers.

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269 The hardness distribution profile of the HAZs for the three samples is shown in Figure 9. To
 270 compare the hardness of different HAZs after thermal simulation using different flash allowances,
 271 the hardness tests were conducted along a centreline in the specimens from CGHAZ to BM at a
 272 load of 0.5 kg for 10 s. The distance of each two adjacent hardness test points was 0.4 mm.
 273 Compared to the parent material, whose average hardness is 233 HV_{0.5}, the hardness of the HAZs
 274 decreases greatly. The regions with the lowest hardness are ICHAZ according to microstructural
 275 investigation, with average hardness values of the softening zones for FA14, FA10 and FA4 being
 276 183 HV_{0.5}, 185 HV_{0.5} and 194 HV_{0.5}, respectively. Compared to the parent material, the three
 277 samples have the softening degrees of 21.6%, 20.4% and 16.6% in the ICHAZ.

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279

280 **Fig. 9** Hardness distribution profiles of three samples with different flash allowances.

281

282 For FA14, the lowest hardness of 176 HV_{0.5} occurs at ICHAZ (195 HV_{0.5} at SCHAZ), seen in
 283 Figure 10. This is different with other studies which show the softening occurs at SCHAZ for DP
 284 high strength steel, the DP680 steel has the lowest hardness in ICHAZ when the flash allowances
 285 from 4 mm to 14 mm were selected. The hardness of the materials is related to the factors that can
 286 influence the strength.

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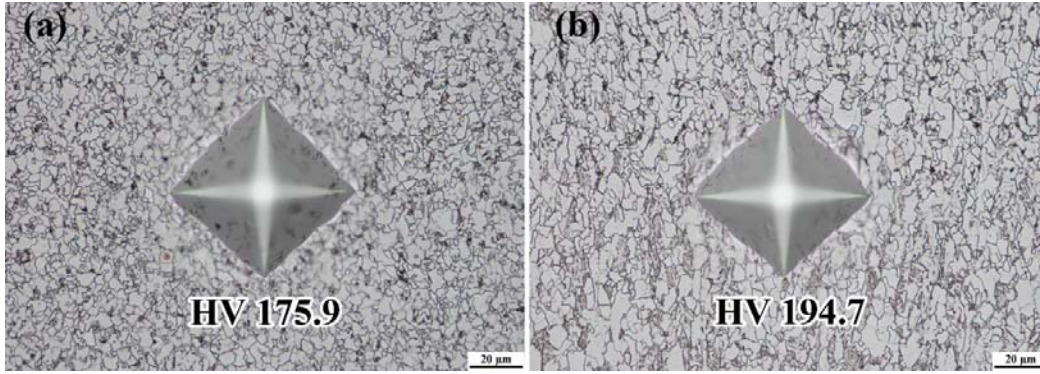


Fig. 10 Hardness difference at ICHAZ and SCHAZ for FA14: (a) FA14-ICHAZ; (b) FA14-SCHAZ.

Combined with the microstructure from OM (Figures 6(a)-3 & (a)-4), the SEM of the ICHAZ and SCHAZ of FA14 was observed in order to display more details, seen Figure 11. In the SCHAZ, it can be observed that the martensite starts to decompose after being tempered. Compared to the SCHAZ, in the ICHAZ, ferrite is finer, part of martensite decomposes, and some bainite forms due to higher peak temperature than A_{c1} and appropriate cooling rate.

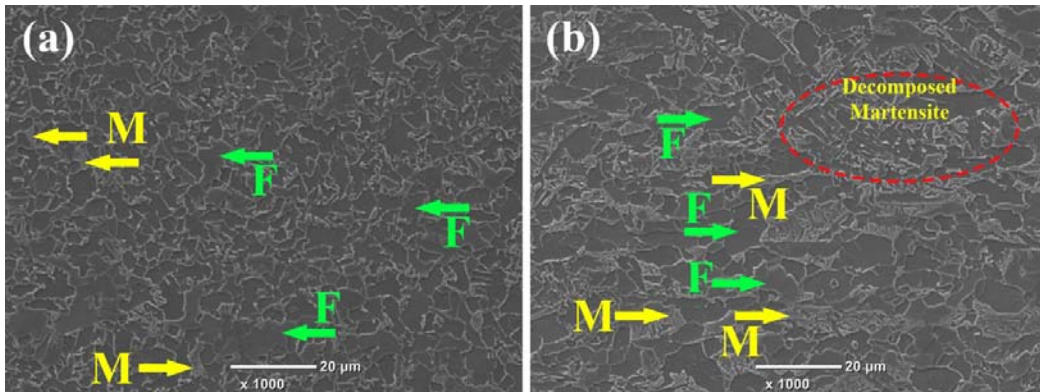


Fig. 11 Microstructure comparison at ICHAZ and SCHAZ under SEM for FA14.

IV. DISCUSSION

A. Conclusions from microstructural and mechanical analyses

As mentioned, DP680 steel has more deformed microstructure (elongated grains) that includes a finer grain size, more martensite in the ferrite matrix compared to the lower grade DP590 steel for further strength increase.

There are two key areas that need more attention during welding HAZ simulation, i.e. brittle area and softening area. The brittle area is always CGHAZ since the area is heated to a temperature near the melting point [18,19], and coarse grains with high hardness form at this area. In the Gleeble simulation, the dual phases, i.e. ferrite and martensite, transforms to austenite at a high temperature, and then into various microstructural constituents upon cooling. The microstructures in the CGHAZ of the three samples with different flash allowances are all lath martensitic, bainitic and ferritic. When the flash allowance is changed from 4 mm to 14 mm, lath martensite decreases while bainite increases. Meanwhile, the grain size increases with decreased cooling rate. The

PAGS is another important factor influencing the CGHAZ toughness. Finer PAGS correlates to higher toughness, which can be clarified in Figure 7. The flash allowance of 4 mm to 14 mm is equivalent to an HI from 8.74 kJ/cm to 15.14 kJ/cm. From the analysis, the reduction of lath martensite, increase of bainite and refinement of grain size are beneficial for toughness improvement since in general, bainite has a higher toughness compared to lath martensite if grain size is not taken into account. According to the CCT curve for the DP steel and References [20] and [21], lath martensite and bainite are changed based on the welding process. However, the results of toughness for final welded sample are still dependent on the details of these two phases, including chemical composition, grain size, etc. The prior austenite grain boundaries are considered to be high angle grain boundaries (HAGBs), which can arrest fracture propagation and thus increase the energy for a microcrack to grow to the critical Griffith crack length [6]. Be noted that the proportion of grain boundaries stands for grain size. More grain boundaries mean higher grain refinement. Hardness variation can indicate the microstructural constituents, and in the CGHAZ, the hardness decreases from FA4 to FA14. The trend of hardness variation was generally consistent with the HAZ hardness model of Yurioka *et al.* [22] correlating the maximum HAZ hardness and the HI/cooling time. It is acceptable that the hardness variation in the HAZ is related to carbon partitioning and transformation products formed during cooling process in the various thermal cycles [23], which leads to the phase transformation between ferrite/ martensite and ferrite/bainite/lath martensite.

The EBSD experiments were conducted to further characterise the CGHAZs under varied FAs. The EBSD analysis was performed on selected samples to examine the microstructures of simulated CGHAZs in more detail.

As shown in Figures 12(a), (c) and (e), the inverse pole figure (IPF) maps exhibit that there is not apparent preferred orientation in the CGHAZs of DP680, which means that the change of flash allowance does not influence the texture evolution in that location. In order to accurately analyse the effect of flash allowance on grain size, the reconstruction of parent grains was launched according to the Kurdjumov-Sachs (K-S) crystallographic correlation, which is $\{111\}_{FCC} // \{011\}_{BCC}$ and $\langle -101 \rangle_{FCC} // \langle -1-11 \rangle_{BCC}$. The PAGS is an important factor influencing the CGHAZ toughness. The finer the austenite grain size is, the more benefits to the toughness performance. Figures 12(b), (d) and (f) display the reconstruction results of the original austenite which show that the average grain size (60.54 μm) of CGHAZ original austenite grain with the flash allowance of 14 mm is larger than the other two (56.26 μm in FA4 and 53.67 μm in FA10). Even though there is no difference in type and size of the martensite phases in the CGHAZ, the change of PAGS still leads to the adjustment in the mechanical properties, especially toughness. Therefore, it can be concluded that the increased flash allowance has an opposite effect on the HAZ CVN for DP680.

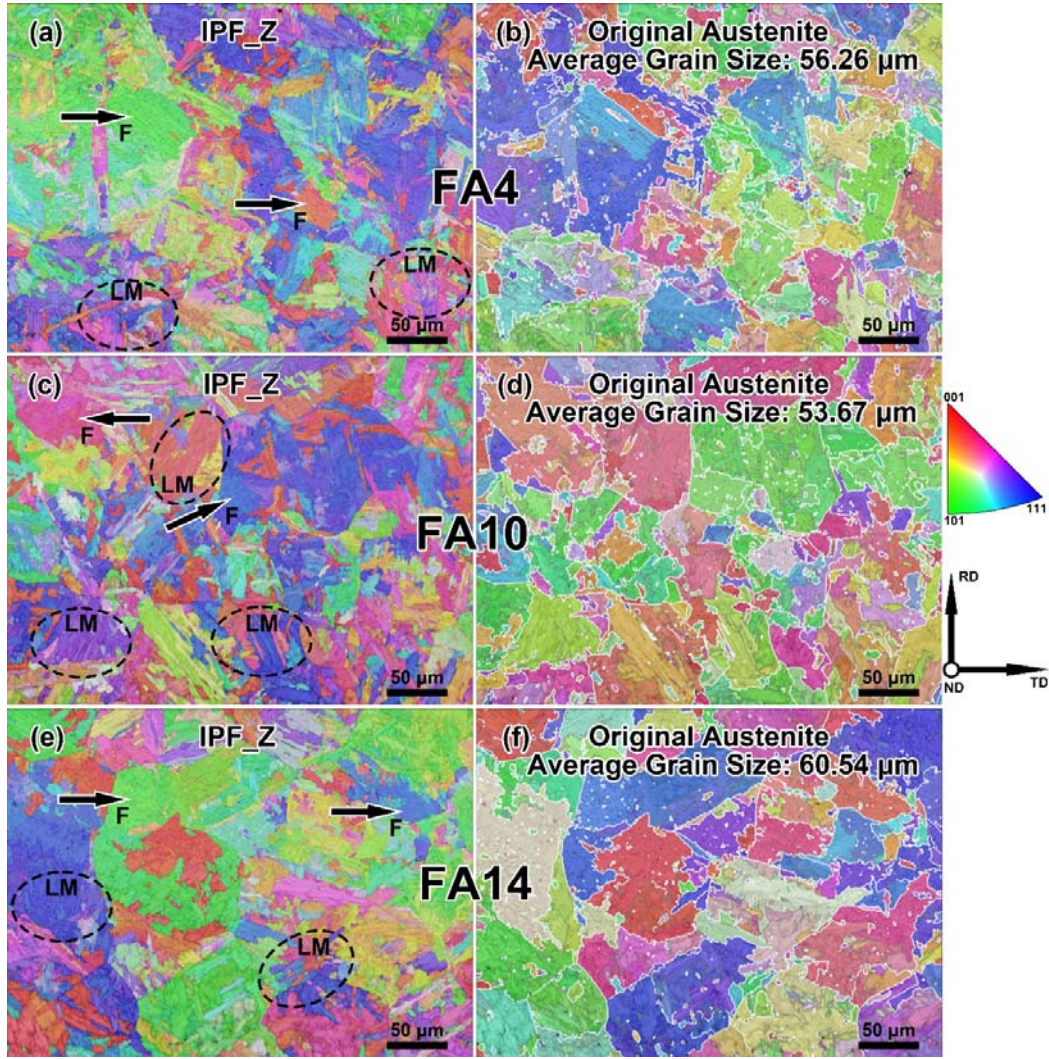


Fig. 12 IPF+BC map and PAGS: (a) and (b) FA4; (c) and (d) FA10; (e) and (f) FA14.

Figures 13(a), (c) and (e) provide the distribution of the recrystallised grains in the three CGHAZs. Note that the grains with mean orientations from 0° to 1.5° were defined as recrystallised grains in this study. Generally, the angle of mean orientation of recrystallised grains can be defined by the user. According to the reference, the recrystallised grain commonly differs from neighbouring cells in lattice orientation by around 1° . In addition, the grain boundaries whose misorientation was 2° have been defined as low angle grain boundaries, which is a form of expression of dislocations. Therefore, we defined mean orientation of recrystallised grains as 1.5° , which is between 1° and 2° [24]. The recrystallised grains mainly located at HAGBs whose angles are higher than 15° [25-27]. HAGBs can be regarded as the grain boundaries of original austenite, as shown in Figure 12. The static recrystallisation progressing under high temperature softens the materials. As shown in Figures 13(a), (c) and (e), the degree of recrystallisation has not been changed with an increase in flash allowance. However, a high HI apparently decreases the dislocation density. Figures 13(b), (d) and (f) illustrate the distribution of kernel average misorientation (KAM) in CGHAZs. The geometrically necessary dislocation (GND) can be calculated based on Eq. 3:

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$$\rho^{GND} = \frac{2KAM_{ave}}{\mu b} \quad (3)$$

373

374 where μ is the step size (0.3 μm) in the scanning, b is the Burgers vector, and KAM_{ave} is the
375 average KAM value. KAM_{ave} has a linear proportion with GND.

376

377 Figures. 13(b), (d) and (f) indicate that the KAM_{ave} of CGHAZs with FA4, FA10 and FA14 are
378 0.6031° , 0.6250° and 0.5885° , respectively, and the ρ^{GND} s for FA4, FA10 and FA14 are
379 $3.5 \times 10^{14}/\text{m}^2$, $3.6 \times 10^{14}/\text{m}^2$, and $3.3 \times 10^{14}/\text{m}^2$, when the b in Eq. 1 is set to be $0.5\langle 111 \rangle$. It is
380 well known that the progress of grain growth consumes a lot of dislocations, therefore the
381 CGHAZ with higher flash allowance exhibited a large average grain size and a lower KAM_{ave} ,
382 which represents a lower dislocation density. The role of dislocation can be explained by
383 dislocation strengthening and work hardening. Hence, the CGHAZ of DP680 duplex steel with a
384 flash allowance of 14 mm revealed the lowest microhardness under a same load. However, from
385 FA4 to FA10, the dislocation density has not been changed evidently. In other words, the flash
386 allowance from 4 mm to 10 mm does not have an influence on dislocation density.

387

388 In summary, a higher HI from larger flash allowance coarsened the grains of the original austenite
389 as well as reduced the dislocation density, which resulted in a decrease in mechanical properties
390 including impact toughness and hardness.

391

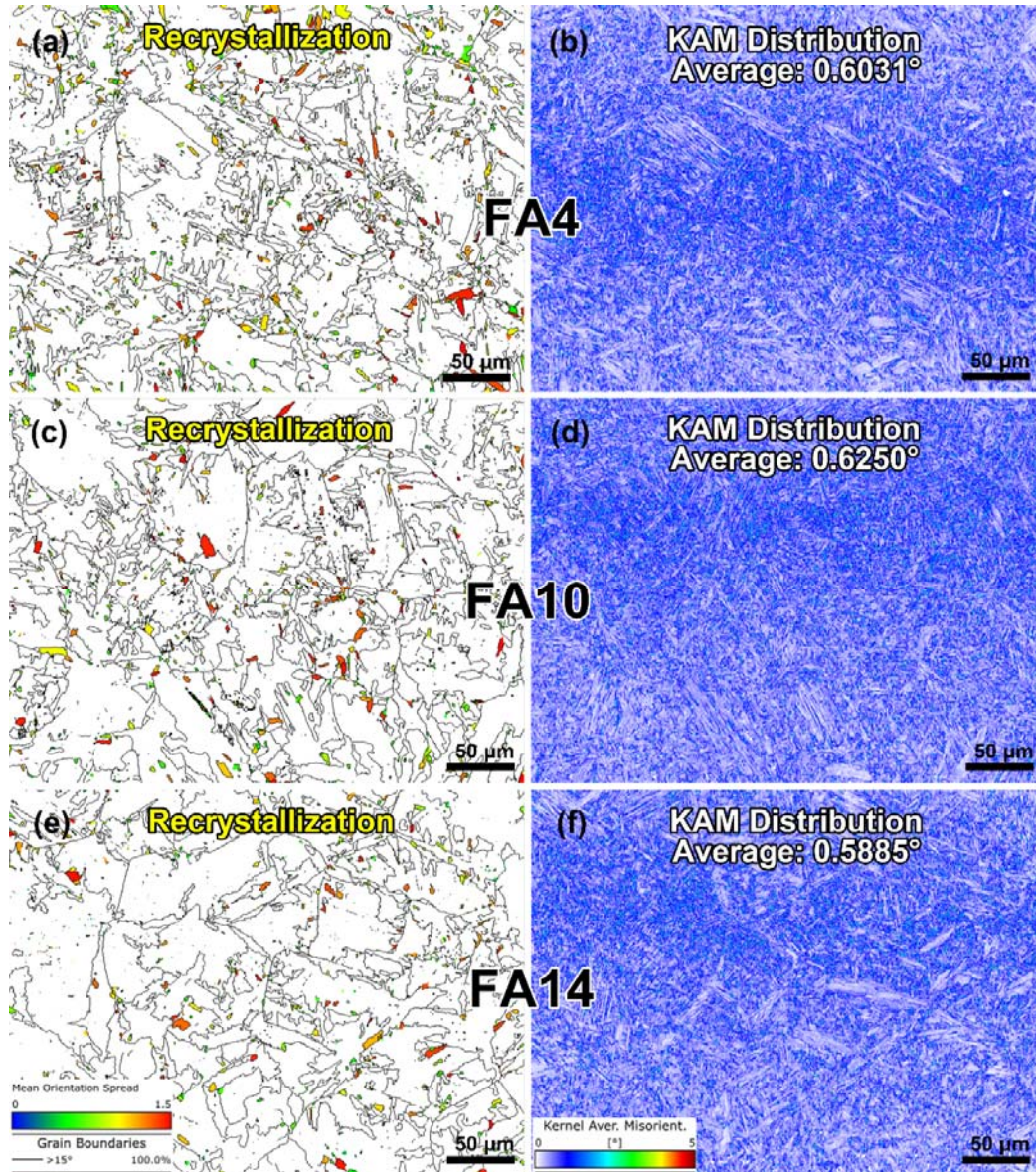


Fig. 13 Recrystallisation distribution and KAM distribution: (a) and (b) FA4; (c) and (d) FA10; (e) and (f) FA14.

In general, PAGBs are the great pathway for crack propagation since as stated the PAGBs are considered to be HAGBs, which can arrest the crack propagation. And more grain boundaries mean smaller grain size. Thus, Fig. 14 also indicates that the grains of CGHAZ were coarsened as the increased heat input. From Fig. 14, the HAGBs for FA4 to FA14 is 42.0%, 38.9% and 33.4%, respectively, and ranks as FA4>FA10>FA14. This is in accordance with the CVN of the HAZ.

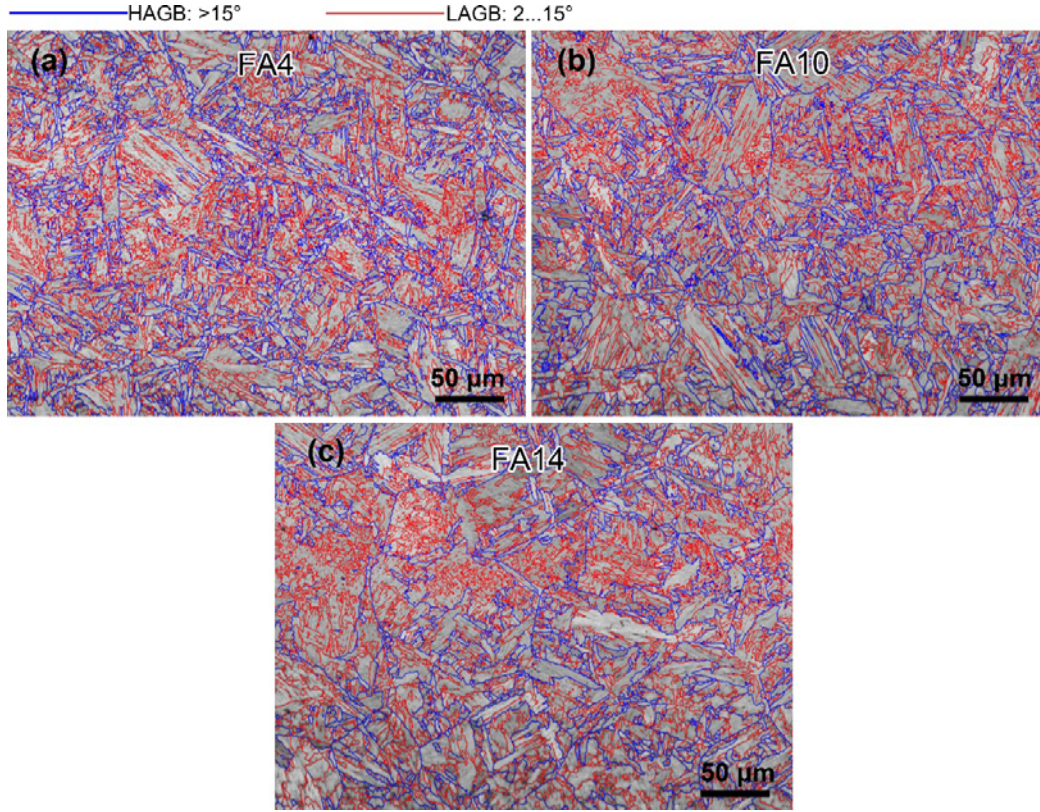


Fig. 14 HAGB distribution of: (a) FA4; (b) FA10; (c) FA14.

Besides CGHAZ, the other areas, including FGHAZ, ICHAZ and SCHAZ, all generate new microstructures during FBW. FGHAZ has a similar microstructure with CGHAZ, but the difference is the grain size due to the lower peak temperature below 1,100 °C. ICHAZ has a microstructure of bainite, ferrite and martensite. Here, when the area of the ICHAZ is heated to above 734 °C, a part of the ferrite and martensite is transformed to austenite, which turns to bainite in a fixed range of cooling rate, and even to ferrite and martensite when the cooling rate is determined by the HI. In terms of SCHAZ, its peak temperature is below 734 °C, as martensite is a thermally unstable phase, it will tend to decompose, resulting in softening of the local microstructure. This phenomenon is recognised as HAZ softening [10], which is influenced by the factors such as martensite fraction, chemical composition, welding HI, etc. [28,29], and has a negative influence on the strength of the weld joint. However, for the DP680 steel, the hardness of the ICHAZ is lower than SCHAZ, and in fact the ICHAZ is the area with lowest hardness in the HAZ areas.

Based on the study of Biro *et al.* [11], the HAZ softening occurs in the tempered or SCHAZ. As martensite tempering fundamentally changes the DP microstructure, the softening of the HAZ originates from martensite tempering, which is a function of cementite nucleation, growth, and saturation. HAZ softening kinetics could be fit to the Avrami equation [30], seen Eq. 4.

$$\Phi = 1 - \exp(-kt^n) \quad (4)$$

where Φ is the volume fraction of martensite that has been transformed, k is a fitting parameter representing the energy barrier to the tempering transformation, t is the tempering time, and n is a fitting parameter related to the rate of the transformation. To describe martensite decomposition using the Avrami equation, the value of Φ was defined to be zero when the material is in its base state. Generally, the conclusion that the SCHAZ has the lowest hardness compared to the other areas due to the occurrence of tempering of the martensite. However, in this case, the fraction of martensite is over 30%, and in such a short duration through simulated welding cycles, not all martensite can get tempered in the SCHAZ. From Figure 11(b), the fraction of rest martensite is 18.14%, which means that only 15.74% martensite has been decomposed during welding.

Since the peak temperature is below A_{c1} , the martensite in DP680 can start to decompose, and the fraction of martensite decomposition depends on the HI used. Compared to the SCHAZ, the existing ferrite and martensite in the ICHAZ were partially transformed to austenite, and then to bainite due to the cooling rate. From Figure 11(b) and Figure 10(b), part of the martensite decomposed into tempered martensite and cementite in the SCHAZ, which apparently reduced the hardness. However, this reduction was lower than the existing ferrite and martensite transforming into bainite even if the grain size was more refined in the ICHAZ. From Figure 11(a), combined with Figure 10(a), it was observed that most of martensite in the ICHAZ has disappeared, and the rest of the ferrite and the newly formed bainite together with the remaining martensite reduced the hardness to a greater extent compared to the SCHAZ.

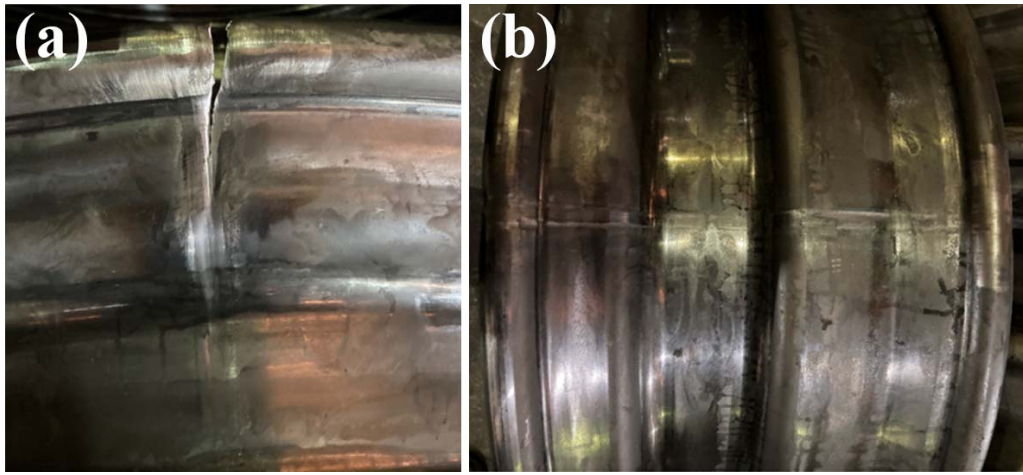
The impact toughness for three different flash allowances at 20 °C is high enough for the application. However, when the test temperature is decreased to -20 °C, the impact toughness for FA14 drops dramatically to 25.3 J, which is only 9.7% of the value at 20 °C. But this value is still acceptable for wheel rim applications. FA10 and FA4 have a higher impact toughness at CGHAZ compared to FA14, which means at lower temperature, the flash allowances of 4 mm and 10 mm are still available for practical applications. As for the CGHAZ, compared to lower flash allowance, FA14 has less lath martensite and more bainite, which is beneficial for toughness improvement. However, its grain size increases, which has greater influence on the impact toughness compared to the microstructure modification.

As a conclusion, during welding HAZ simulation, varied transition behaviours occur according to the peak temperature. For instance, when the peak temperature is over A_{c1} , the transformation of ferrite to austenite occurs, while the peak temperature is above A_{c3} , the austenitic transformation ends. However, be noted that not only the peak temperature, but the duration at this peak temperature are the key factors to determine the microstructure in the corresponding HAZ areas. Therefore, the microstructure at HAZs is not invariable, which is influenced greatly by the detailed parameters of the welding thermal cycles, especially heating and cooling rates and duration at peak temperature. The variation in microstructure can influence the final performance as well.

B. Actual weld joint

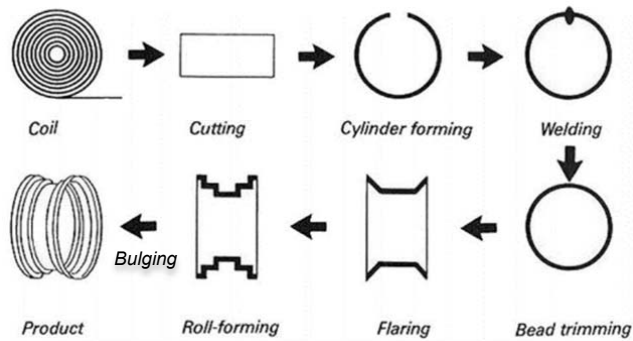
When producing wheel rims, it is of crucial importance to control the FBW parameters, especially flash allowance, which directly controls the HI. If the FBW parameters are not appropriate,

470 cracking occurs during the forming process after welding (Figure 15(a)). Even if the wheel rim is
 471 successfully formed, the localised necking may occur if the softening area is too wide (Figure
 472 15(b)).
 473



474
 475 **Fig. 15** Defects occurred during forming after FBW: (a) Cracking generated during bulging; (b)
 476 Localised necking generated along FBW-ed joint
 477

478 It needs to be clarified that work hardening occurs after a series of forming processes for both
 479 parent material and flash butt weld joint. The comprehensive forming procedures include cylinder
 480 forming, flaring, 3-pass roll-forming, and bulging [30,31] (Figure 16).
 481

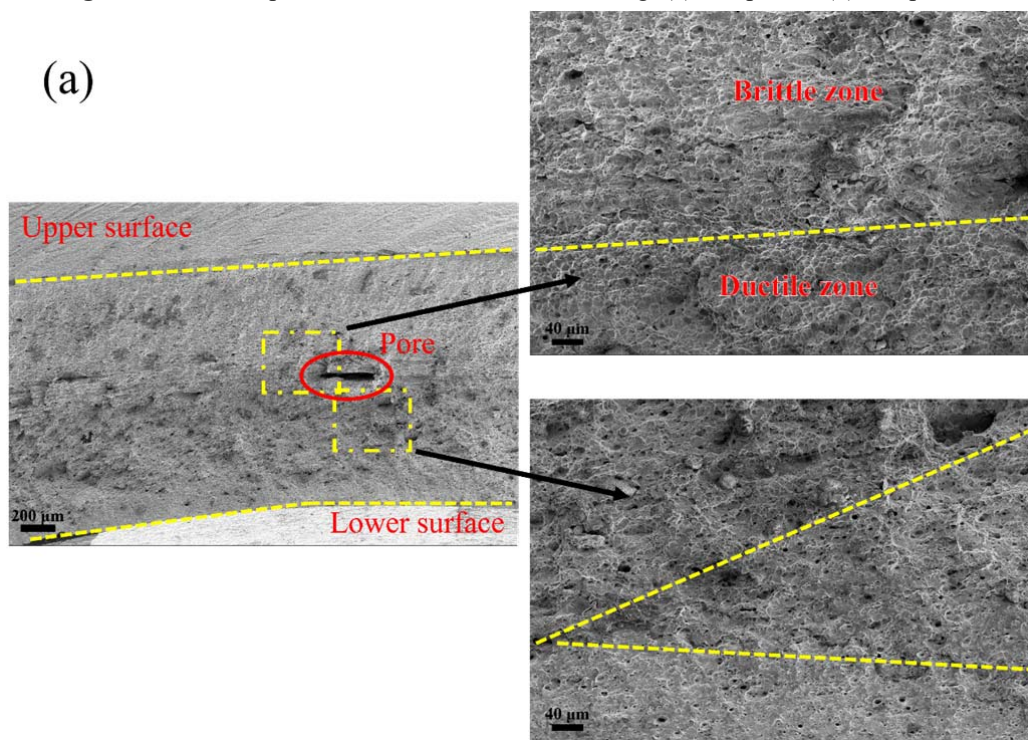


482 **Fig. 16** Forming processes of wheel rim after FBW.
 483

484 In order to investigate the root reason for cracking failure, both wheel rims with and without
 485 cracking (Sample 1# and 2#, Figure 17) were analysed with the HI used for Sample 1# being
 486 higher than Sample 2# (flash allowance is in the range of 10 ~ 14 mm). As a result, Sample 1#
 487 initially cracked from the location of Sample 1-a. Further, even if Sample 2# was successfully
 488 formed, three typical locations of Sample 2# showed slightly localised necking after forming.
 489 However, the softening degree was greatly relieved compared to the simulated sample, seen in
 490 Table V.
 491
 492



Fig. 17 Wheel rims produced with and without cracking: (a) Sample 1#; (b) Sample 2#.



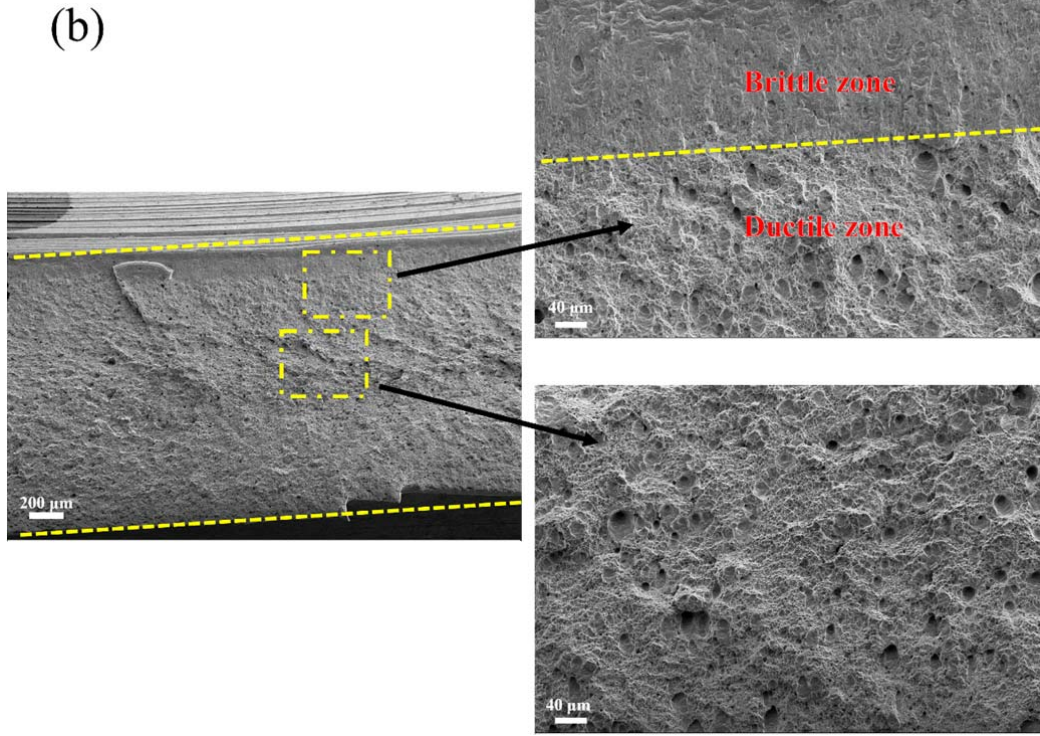


Fig. 18 Fracture morphology of Sample 1# under SEM: (a) Location 1-a; (b) Location 1-b.

Table V. Comparison of localised necking for three typical locations of Sample 2#

2#		2-a	2-b	2-c
CGHAZ Width (mm)	Upper surface	9.83	7.65	6.95
	Centre	7.39	4.94	4.65
	Lower surface	10.17	8.58	7.32
Edge (left, mm)		5.27	4.45	4.07
Reduced thickness (left, mm)		4.95	4.45	3.86
Reduced thickness (right, mm)		5.22	4.36	3.81
Edge (right, mm)		5.40	4.33	4.03
Larger thinning rate between left and right edges		8.2%	2.0%	6.6%

From Figure 17(a), it is understandable that for Sample 1#, the cracking originated from the HAZ. Besides that, some pores were found at the fracture location (Figure 18(a)), which indicates the low toughness of Sample 1-a. The fracture in Fig. 18 (a) shows a pore, and this pore can be addressed to be the defect of the original material, which is different to the typical tensile fracture and welding defect. During the deformation, the existing pores always decrease the ductility of the material [32]. As shown in Fig. 18(a), the pore exhibited an elongated shape, this means the defect had been enlarged during deformation. Meanwhile, the pore led to the stress concentration during deformation, which will result in the failure [32, 33]. In addition, the pores can be the sources of cracks and result in the fracture after yield, and this situation can also occur in high strength-toughness materials [34]. Therefore, the Sample 1-a shows a low ductility.

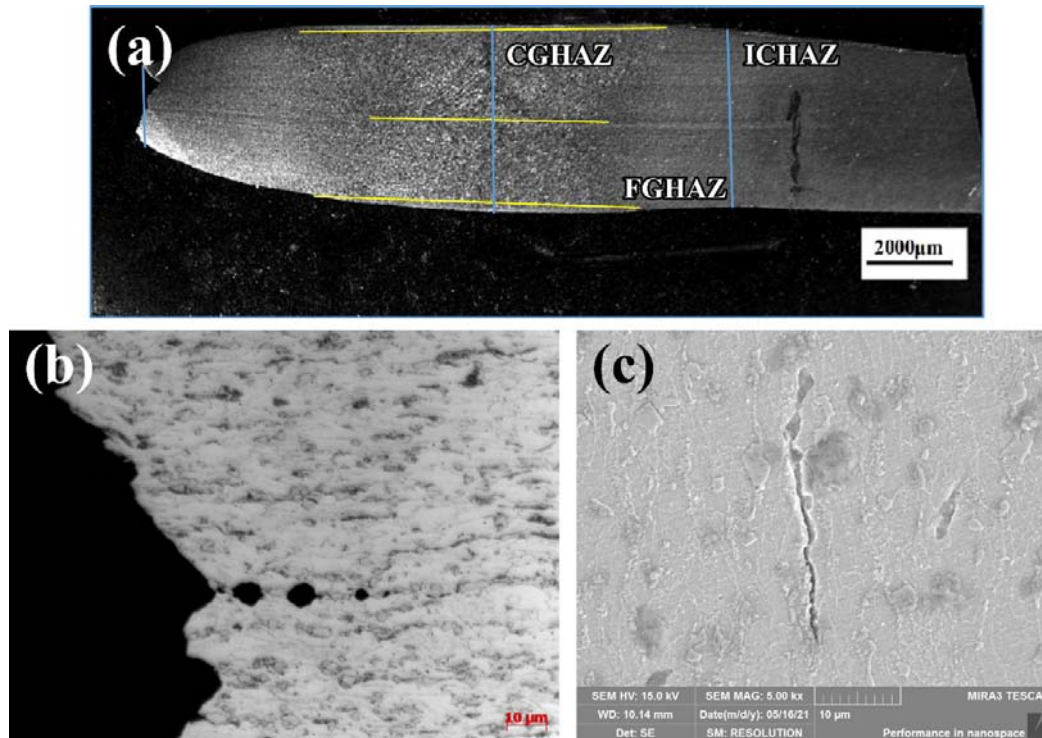
511

512 The fracture in Fig. 18 exhibited two different types, which were brittle zone with quasi-cleavage
 513 and ductile zone with dimples. It is noticed that the pore located at the boundary of brittle zone
 514 and ductile zone, as shown in Fig. 18(a). During the bulging, the strain state of brittle zone and
 515 ductile zone were different, hence the stress concentration is more likely to occur at the boundary
 516 of the two zones. This condition can further lead to the negative affect of pores on the mechanical
 517 properties. For Sample 1-b (Figure 18 (b)), there is a narrow band of cleavage fracture, which
 518 suggests the brittle microstructure. When samples are taken from near the HAZ, some
 519 embrittlement was observed when the welding parameters are not controlled well, especially the
 520 HI.

521

522 The failure location for Sample 1# was seen to be the ICHAZ, shown in Figures. 19(b) and (c),
 523 where the most softening occurs. The locations near the fractures have experienced a serious
 524 extension. Two steps during cracking determine the final fracturing, including, 1. Crack initiation,
 525 and 2. Crack propagation. It is evident that some cracking behaviours have occurred near the
 526 fracture. However, the cracking has been paused due to some possible reasons, including, 1. There
 527 is some microstructure which has a good ductility and resist the crack propagation, 2. The fracture
 528 has occurred, which led to uncompleted fracture of this paused crack. Compared to the parent
 529 material (Figure 2(b)), the typical microstructure of the ICHAZ is the retained small blocks of
 530 martensite, rest ferrite, and new bainite phases.

531



532

533 **Fig. 19** Fracture appearance and microstructure of DP680 after bulging due to localised thinning:

534 (a) Fracture appearance; (b) Fracture microstructure under OM; (c) Fracture microstructure under

535

536

V. CONCLUSIONS

The microstructures and mechanical properties of simulated HAZs of a newly-developed DP680 dual phase steel studied led to the following conclusions:

(1) The CGHAZs of the target steel produced with a Gleeble simulator under different flash allowances (HIs) varying from 4 mm to 14 mm showed fracture toughness in the range between 8.74 kJ/mm and 15.14 kJ/mm. The results indicated that superior toughness can be achieved in the selected flash allowance (HI) range at 20 °C. When the test temperature was decreased to -20 °C, the toughness dropped dramatically, however, the flash allowance range can still satisfy product requirements of the practical applications.

(2) The prior austenite grain size, together with the transformed microstructure, in the CGHAZ determines the final performance, i.e., toughness and hardness. Compared to a lower flash allowance, the sample with higher flash allowance has less lath martensite and more bainite. For DP680, the CGHAZ has an acceptable performance if the flash allowance is changed from 4 mm to 14 mm. However, the performance and dislocation density of the samples whose flash allowance is changed from 4 mm to 10 mm show no evident difference.

(3) From the simulation, the softening behaviour for the DP680 steels ranges from 16.6% to 21.6%. However, after forming, varying degrees of work hardening occur in the different HAZ areas with the local thickness decreasing with increased flash allowance and elongation. From a practical viewpoint, the flash allowance of 12 mm generates the largest localised necking of 8.2%, which is located at the ICHAZ as investigated through Gleeble simulation. The localised necking is greatly relieved compared to the original state, which can be acceptable although the failure location has not changed.

ACKNOWLEDGEMENTS

The authors would like to thank the support from UOW-HBIS Project (No. IRIS 200506001).

CONFLICT OF INTEREST

On behalf of all authors, the corresponding author states that there is no conflict of interest.

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HAZ effects in hot-rolled dual-phase steel during flash butt welding of wheel rims

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2023-12-01

Springer

<https://doi.org/10.1007/s11663-023-02914-1>

Han J, Zhu L, Wang J, et al., (2023) HAZ effects in hot-rolled dual-phase steel during flash butt welding of wheel rims. Metallurgical and Materials Transactions B - Process Metallurgy and Materials Processing Science, Volume 54, December 2023, pp. 3361-3376