Chapter 2
Resistance spot welding of AHSS
- a literature review

This chapter is a literature review to capture preliminary insight into the RSW of AHSS with emphasis on DP steels. First, it provides an introduction to the basic definitions of resistance spot weld physical and metallurgical attributes. Conventional testing methods to evaluate the mechanical proprieties of resistance spot welds and different failure modes are explained. Next, the stress distribution and damage mechanism for different failure modes during mechanical testing of AHSS resistance spot welds are discussed. The parameters that affect the microstructural evolution of fusion zone and heat affected zone and also hardness distribution over the weld zone are analysed. The effect of embrittlement elements on the mechanical response of the welds is briefly reviewed and, finally, the commonly used models developed for the prediction of the failure mode and critical weld nugget size are introduced.
2.1 Resistance spot weld attributes

The typical cross section of a resistance spot weld is shown in Figure 2.1. The fusion zone (FZ) is the region that is melted and resolidified during RSW. Weld metal composition and solidification condition strongly affect the soundness and serviceability of the joint. The solidified material in the FZ can be considered as a cast structure within a small volume. First, the solid phase nucleates at the contact surface of electrode and sheet metal and then crystal growth occurs towards the centreline. Because of the rich chemistry of DP steels, the FZ microstructure after rapid cooling mainly consists of martensite and to some extent of bainite. The FZ size ($D$) is defined as the width of the weld nugget at the sheet-sheet interface and is the most important parameter determining the mechanical performance of the weld as it governs the bonding area of the joint. The FZ size is controlled by the welding parameters; mainly welding current, welding time and electrode force. Electrode indentation depth is another factor that can affect the mechanical properties of the resistance spot weld and is dependent on the electrode force and the temperature of sheet/electrode interface. Apparently, increasing the heat input results in an increase in the temperature of the sheet/electrode interface that in turn leads to a larger penetration depth as a result of higher plastic deformation applied by the electrodes. Maximum penetration depth should be considered less than 10% of the sheet thickness as a too large penetration depth can reduce the mechanical strength of the weld.

![Figure 2-1 Resistance spot weld cross section [1].](image)

Figure 2.2 illustrates the change in weld nugget size with welding current. As shown, an increase in welding current leads to a rapid growth of the weld nugget. At higher welding current the growth rate decreases and finally there is decrease in nugget size that corresponds to the expulsion phenomenon during which the molten area is too large to be held by the electrodes and it is associated with the ejection of the molten material from between the sheets. An ideal weld nugget has sufficient diameter and penetration. A small weld nugget has too low a strength to carry the loads in crash events and shows reduced fatigue life under normal operation of the vehicle. Conversely, an oversized nugget needs higher energy and time to be welded making RSW an inefficient and costly process. The splash limit is exceeded when excessive heat input is used for welding. Macroscopic voids form because of
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splashing of molten materials; hence decrease the load-bearing surface of the weld and impair the mechanical properties of the joint.

Figure 2-2 Weldability lobe based on change in weld nugget size with welding current [2].

The weldability range or lobe is defined as the region where an acceptable joint is produced using an appropriate combination of welding time and current. The weldability range is limited by the minimum acceptable weld size and splash limit. The width of the welding lobe determines the welding window that can produce acceptable welds. Various industrial standards have been proposed for minimum weld nugget size. For example, the American Welding Society (AWS), Society of Automotive Engineering (SAE), and the American National Standards Institute (ANSI), recommend a weld nugget diameter $D = 4\sqrt{t}$, where $D$ and $t$ are the weld nugget diameter and sheet thickness respectively [3]. Also, Japanese JIS Z3140 and German DVS 2933 standards proposed the minimum weld nugget size as $D = 5\sqrt{t}$ [4,5]. All of these standards have been derived from extensive experimental testing and are empirical in nature. Based on these equations, the minimum weld nugget size is not affected by mechanical properties and it changes only as a function of sheet thickness.
The Heat affected zone (HAZ) is the zone affected only by the weld thermal cycle and is not melted during RSW. The HAZ is located adjacent to the nugget and experiences lower peak temperature. Based on the distance from the weld nugget and peak temperature, the HAZ can be divided into four regions as shown in Figure 2-3. (1) The sub-critical HAZ (SC-HAZ) at which the peak temperature is well below $A_c_1$, leading to tempering of metastable phases such as martensite and bainite as well as coarsening of carbides. Normally, the SC-HAZ does not show distinct changes, as the peak temperature does not cause any phase transformation. Hence, it is difficult to distinguish this region from the base metal (BM). (2) The inter-critical HAZ (IC-HAZ) at which the peak temperature varies between $A_c_1$ and $A_c_3$ undergoes partial transformation. Austenitization occurs in this range and increase in the peak temperature results in an increase in the fraction of ferrite dissolved into austenite. Rapid cooling leads to the transformation of inter-critically formed austenite to martensite. (3) The fine grained HAZ (FG-HAZ), where the peak temperature is above $A_c_3$ and a fully austenitized microstructure is formed during the heating stage that subsequently transforms to an ultra-fined martensitic structure. (4) The coarse grained HAZ (CG-HAZ) experiences a higher temperature compared to the FG-HAZ. Austenite grain growth is facilitated at elevated temperature and coarsened martensite is formed after subsequent rapid cooling.

![Figure 2-3 A schematic image showing various weld zones based on experienced peak temperature for a low carbon steel [6].](image)

Vooids, cracks and shrinkages are also influential on the mechanical response of resistance spot welds. In most of the cases, crack leads to lower mechanical performance of the weld. However, surface cracks, which originate from liquid metal
embrittlement, do not affect the weld mechanical behaviour as they are not located at the position of maximum loading. Voids can be formed in the weld nugget because of two possible reasons: expulsion and solidification shrinkages. As already discussed, expulsion is associated with the ejection of the molten material from the weld nugget resulting in the formation of voids and deterioration of the weld quality. The shrinkages are formed because of differences in the contraction rates of the weld nugget and surrounding solid material. It was found that the materials with a higher content of alloying elements are more prone to solidification shrinkages. It was shown that longer holding time and higher electrode forces help to reduce shrinkage voids [7].

As seen in Figure 2-1, a notch or pre-crack is formed at the circumference of the weld. The failure behaviour and mechanical properties of the weld are strongly affected by the shape of the notch, e.g. sharp versus square. A sharp notch increases the stress concentration at the weld edge leading to preferential crack propagation through the weld nugget [8].

2.2 Mechanical testing of resistance spot welds

Different test methods have been proposed to evaluate the mechanical behaviour of the resistance spot welds. The most widely applied loading modes to assess spot welds are shear and peel. To apply these loading modes, several sample types have been suggested. Among them lap-shear or tensile-shear samples and cross-tension samples are the most frequently used (Figure 2-4).

![Figure 2-4 Tensile-shear (lap-shear) and cross-tension sample geometries [9].](image)

Figure 2-5 represents a typical load-displacement curve of a spot weld during mechanical testing. The mechanical performance of the joint is evaluated based on the following parameters achieved from the mechanical tests:

- Peak load $P_{\text{max}}$: The maximum force measured during testing;
- Ductility: This could be described by the maximum elongation at the peak load or failure energy at the peak load (surface below tensile curve);
- Weld nugget size: One of the most important indexes to evaluate the weld quality;
- Failure mode: Failure mode is the qualitative measure of mechanical performance of weld joints.

It is worthwhile to note that mechanical testing of a welded sample is somehow different from the bulk material, since the welded sample is composed of non-homogeneous materials with strong gradient in properties. This is why the strength of the weld is often expressed in load instead of stress and displacement instead of strain [6]. The maximum load and displacement strongly depend on the sample dimensions and loading conditions. Therefore, the reported values must always be accompanied by the detailed description of the testing method.

![Figure 2-5 A typical load-displacement curve of a mechanically tested resistance spot weld [1].](image)

The AWS has proposed a list of eight potential failure types that could occur during destructive testing of AHSS spot welds [3]. The most characteristic failure modes are: pull-out failure (PF) (mode 1); partial interfacial failure (PIF) (mode 5) and interfacial failure (IF) (mode 7) as shown in Figure 2-6. In the IF mode, the weld fails at the interface of two sheets leaving half of the weld on one sheet and half on the other. In the PF mode, failure may occur in the BM or HAZ at the perimeter of the FZ. In this case, the weld nugget is completely torn from one sheet. It is also possible to get combinations of these two modes, in which the crack propagates through the FZ and is then redirected through the thickness (PIF).

The failure mode has a very strong influence on the load bearing capacity and energy absorption capability of the weld. Generally, the PF mode is considered as the preferred failure mode, as it is associated with larger plastic deformation and energy absorption capability as opposed to the IF mode, that is an indication for the reduced crashworthiness in the automotive industry [1]. Thus, welding parameters are always adjusted to produce a weld that guarantees the PF mode as it can be a demonstration of a good quality weld.
Although the process-structure-property relationship is well investigated and understood for conventional mild steels, resistance spot welding and also quality control of AHSS come with some issues as the microstructure of AHSS resistance spot welds show non-equilibrium phase transformations that completely change the sophisticatedly designed original structure of the BM [10]. They also suffer from a higher tendency to fail in the IF mode [11]. Furthermore, the high susceptibility of AHSS resistance spot welds to the formation of shrinkage voids in the FZ because of their rich chemistry has been reported [12]. As described already, the equation $4\sqrt{\ell}$ is proposed by standards for the minimum weld nugget size to ensure the PF mode of resistance spot welds. However, it was shown that this criterion does not always reliably predict the failure mode in the case of AHSS and larger weld nugget size is required to guarantee the desired failure mode of the weld. Failure modes of resistance spot welds for different steel grades from low carbon to high strength steels with strength ranging from 206 to 655 MPa were investigated and the minimum weld nugget sizes to ensure the PF mode were determined [13]. Figure 2-7 shows the results of the required minimum weld nugget size for a high strength steel with different thicknesses. The $4\sqrt{\ell}$ recommendation is also superimposed on the graph. As shown, no direct correlation can be made between the thickness of the sheets and critical weld size for the PF mode on high strength steels. In particular, when the sheet thickness is higher than 1.5 mmm, the $4\sqrt{\ell}$ recommendation cannot guarantee the PF mode. It can be inferred that other metallurgical parameters in addition to the sheet thickness may play a role in determining the critical weld nugget size.

Marya et al. [14] studied the failure mode of DP600, DP780, DP800, DP980 and TRIP800 resistance spot welds and found that the general recommendation of $4\sqrt{\ell}$ is not sufficient to obtain the PF mode; they developed an empirical model that also considers the hardness ratio of the weld nugget and failure location. Sun et al.
[15] examined the effect of the weld nugget size on the peak load and energy absorption of DP800 steel during tensile-shear testing as shown in Figure 2-8. A gradual increase in peak load and energy absorption versus weld nugget size was found despite the large scattering in data. Similarly, they showed that the weld size guidance of $4\sqrt{\ell}$ cannot produce nugget PF mode for DP800 resistance spot welds.

![Critical weld nugget size for a high strength steel with different sheet thickness](image1)

*Figure 2-7 Critical weld nugget size for a high strength steel with different sheet thickness [13].*

![Peak load (a) and energy absorption (b) variation of DP800 resistance spot weld](image2)

*Figure 2-8 Peak load (a) and energy absorption (b) variation of DP800 resistance spot weld during tensile-shear testing[15].*
2.3 Stress distribution and damage mechanism during mechanical testing

The real stress/strain distribution during mechanical loading of resistance spot welds is so complicated due to complexities in sample and loading geometry and also non-homogenous properties of different weld zones. The presence of the notch at the weld edge and also its shape can vary the stress concentration around the edge even 20 times higher than the nominal stress [16]. A model was proposed by Chao [9] for the stress distribution around a resistance spot weld during tensile- shear and cross-tension testing in the case of PF mode. It was assumed that the PF failure during tensile-shear testing occurs due to uniaxial tensile stresses around the weld nugget, assuming a rigid circular weld nugget. A harmonic tensile stress distribution around the weld nugget was proposed as shown in Figure 2-9a. In the case of the cross-tension test, the PF is predominantly governed by the shear around the weld nugget and a harmonic tensile stress distribution around the circular weld nugget was proposed as shown in Figure 2-9b.

![Figure 2-9 Stress distribution around the weld nugget for PF mode during tensile-shear (a) and cross-tension (b) testing [9].](image)

Finite element modelling and fracture mechanics calculations were carried out by Radakovic and Tumuluru [17] to predict resistance spot weld failure modes in tensile-shear testing of AHSS. It was found that the required load for the PF mode is proportional to the tensile strength, thickness of the sheet and weld nugget size. The force that leads to IF failure mode is a function of the fracture toughness of the weld, sheet thickness and weld diameter. They showed that as the strength of the steel increases, the fracture toughness of weld to avoid the IF mode must also increase. According to their model, the maximum local strain is concentrated in the weld nugget in the case of the IF mode, whereas for the PF mode, the maximum plastic strain is concentrated outside the weld nugget at the inner surface of the sheet and decreases in direction to the outer surface leading to necking and failure at this location (Figure 2-10).
Figure 2-10 Predicted plastic strain distribution that occurs during IF (a) and PF (b) mode in tensile-shear loading [17].

Lin et al. [18] also conducted a two-dimensional plane strain elastic-plastic finite element analysis to simulate the PF mode of a DP resistance spot weld in tensile-shear loading (Figure 2-11). It was shown that the necking outside the weld nugget starts with localized shear bands from the surface of the sheet along the 45° lines as schematically shown in Figure 2-11a. As the shear forces increase, the shear bands develop and finally necking occurs outside the weld nugget near the intersection of the two 45° bands. The location of necking failure was modelled as about one thickness away from the nugget when the finite deformation of the sheet or the elongation of the neck region is accounted for as shown in Figure 2-11b.

Figure 2-11 A schematic image of the shear bands in the location of necking (a) and a contour of the equivalent plastic strain [18].

Brauser et al. [19] investigated the local surface deformation behaviour of similar and dissimilar spot welded steels in tensile-shear loading via an optical strain field measurement system. Figure 2-12 illustrates the maximum local strain ($\varepsilon_{x,\text{max}}$) versus tensile-shear load together with visualization of measured strain field. For example, the TRIP steel HCT690T shows a less ductile behaviour and fails after a small range of plastic deformation. The maximum surface deformation is obtained in the HAZ and BM transition zone and its value is nearly 5% (Figure 2-12b). In contrast, a micro-alloyed HX340LAD spot weld exhibits a larger plastic deformation until failure and fails in a very ductile manner compared to the other welds. The maximum strain concentration is again concentrated at the HAZ-BM transition zone outside the weld and it reaches a maximum value of 15% (Figure 2-12d). The method is an excellent approach to visualize the deformation behaviour of the spot weld during mechanical loading for the PF mode outside the weld nugget. However, it fails to monitor the plastic strain when the failure occurs in the weld nugget at the faying surface of the two sheets.
Oversimplified loading condition and stress distribution during mechanical loading can be summarized as shown in the schematic sketches of Figure 2-13. During tensile-shear testing, the weld nugget is subjected to shear stresses, while the HAZ and BM experience shear in the thickness direction and tensile stress in the loading direction (Figure 2-13a). In the cross-tension test, the main loading type in the weld nugget is tensile and mode I crack tip opening occurs at the weld edge. In the HAZ and BM the main loading state is shear as well as bending moments during testing (Figure 2-13b).
Damage of AHSS spot welds during tensile-shear loading was investigated by Dancette et al. [20], by means of combination of microtomography, metallography and fractography. Strain localization and failure in the BM was observed for large spot welds except for DP980 steels. Presence of relatively soft BM and SC-HAZ promoted the PF mode of these spot welds. The development of PF damage during tensile-shear loading of DP450 spot weld is shown in Figure 2-14. The sample is loaded elastically until point A with no evidence of damage. It is followed by significant plastic deformation and beginning of necking outside the weld from point A to B as presented in the optical microscopy image. Due to the initially not aligned opposite tensile forces the weld starts to rotate under the action of the bending moments. As seen in the 3D microtomographic view, no damage is visible in the weld itself at point B, despite slight notch opening at the weld edge. At the maximum load, necking leads to failure in the BM/SC-HAZ and subsequently a sudden drop in load (point C). Finally, the specimen tears from the BM upon further loading at point D.

The IF mode was observed for small welds and also for large DP980 spot welds influenced by the strong tangential component of the resultant load at the faying surface. The damage development for IF mode of DP980 steel with higher thickness compared to DP450 in tensile-shear test is shown in Figure 2-15. Similar to the PF mode, the weld is elastically loaded until point A, without any damage evidence. Slight rotation of the weld nugget is observed at point B close the maximum load as
observed in the optical micrograph. A slight decrease of the slope of the loading curve at point B indicates some level of plastic deformation. Because of higher sheet thickness and also higher stiffness the degree of rotation is much smaller compared to the DP450 weld. As observed in the 2D microtomographic image at point B, there is small notch tip propagation in the transverse direction. Because of macroscopic resultant shear loads, a mode III anti-plane shear loading of the crack tip occurs as opposed to the mode II in-plane shear that is expected in the loading direction. The final fracture occurs suddenly and in an unstable way at the faying surface of two sheets at point C leading to the formation of sheared fracture surface.

Figure 2-15 Damage development in the IF mode of DP980 spot weld in tensile-shear testing [20].

Figure 2-16 illustrates the fracture surface of DP980 spot weld after tensile-shear loading failed in the IF mode. While the flank side of the weld shows a complex fracture surface with limited ductility, elongated dimples in the central part of the weld in accordance with the in-plane shear loading can be observed.

A similar approach was used to evaluate the fracture mode during cross-tension loading of AHSS [21]. The damage process upon PF mode for the DP450 steel is presented in Figure 2-17. The vertical displacement of the grippers that clamp the horizontally positioned two sheets of cross-tension specimen results in superimposed bending and tension in both the lower and upper sheets. The stress is concentrated around the weld and a plastic zone starts to develop in the initial stages of loading. This leads to folding of the sheet in this area that also changes the slope of the curve at the displacement around 4 mm. Further loading results is strain concentration at the BM/HAZ outside the weld nugget as illustrated at point A. Microtomography at point B, close to the maximum load, reveals development of a crack from the inner side of the sheet. Also, the coating is cracked on the outer side of the sheet. Point C corresponds to the stage that the crack reaches the outer surface.
of the sheet leading to a sudden drop in load. Further loading tears the weld in the BM until the sample completely fails.

![Figure 2-16 Fracture surface of the DP980 spot weld failed in IF mode during tensile-shear loading [20].](image)

**Figure 2-16** Fracture surface of the DP980 spot weld failed in IF mode during tensile-shear loading [20].

Another commonly observed failure type during cross-tension testing is the PF mode by ductile shear in the HAZ caused by the notch tip propagation at the weld edge. The damage mechanism of this failure mode for an IF260 spot weld is observed in Figure 2-18. As observed at point A, the weld at its initial configuration shows a sharp crack tip at the weld edge. Mode I loading applied during cross-tension testing tends to open the notch tip. However, the ductile microstructure of the weld nugget and HAZ in the case of IF260 steel leads to blunting of the crack tip (point B). Crack initiation from the notch tip is firstly observed at point C, very close to the maximum load. Finally failure develops from the cracks initiated from the notch tip in a sudden
manner, shortly after point C. DP495 and DP590 also failed in the same manner except that their HAZ showed less ductile behaviour due to higher hardness.

![Image](image1.png)

**Figure 2-18** Ductile shear fracture of large IF260 spot welds during cross-tension testing [21].

![Image](image2.png)

**Figure 2-19** PIF crack propagation in the weld nugget of DP980 spot weld during cross-tension loading [21].

If the weld nugget has low fracture toughness, mode I loading of the notch tip during cross-tension loading may result in brittle or semi-brittle fracture at the faying surface. The damage mechanism of the DP980 spot weld, which failed in the PIF mode is shown in Figure 2-19. Mode I loading leads to crack initiation at the
notch tip after few millimetre displacement at point A. The complex solidification structure of the weld together with the asymmetric nature of the loading during cross-tension testing results in a complicated crack path upon loading, as shown at points B and C. As shown in 3D reconstruction of microtomograph of the weld nugget, a single complex crack deviated upward in the upper sheet and downward in the lower sheet. The maximum load during this failure mode is obtained once the crack reaches the outer surface in the weld nugget (point C). Final fracture occurs when the crack is sufficiently developed and allows complete separation of the sheets (point D).

Figure 2-20 depicts the fracture surface of three different kinds of fracture during cross-tension test. The fracture surface the PF mode by necking in the BM is covered by dimples showing ductile fracture behaviour (Figure 2-20a), while the PF mode by shear in the HAZ shows typical elongated dimpled structure for ductile shear failure (Figure 2-20b). PIF is associated with cleavage as well as a small area of ductile fracture as illustrated in Figure 2-20c.

![Figure 2-20 Fracture surface of PF mode by necking in the BM (a), PF mode by shear in the HAZ (b) and semi-brittle PIF in cross-tension loading [21].](image)

2.4 Microstructural evolution and hardness distribution in FZ

The final microstructure of the FZ is mainly governed by the thermal cycle of the welding and chemical composition of the BM. A severe thermal cycle is applied to the materials during the RSW process. The heating and cooling rate of RSW is much higher than conventional welding techniques. The cooling time from 800 °C to 500 °C ($\Delta t_{8-5}$) for 0.8 mm sheet thickness is $\sim 0.06$ s, which is much shorter than $\Delta t_{8-5}= 8$ s for shielded metal arc welding [1]. The heat dissipation during RSW is controlled by two mechanisms: heat dissipation through the water-cooled electrodes and heat dissipation via the adjacent colder BM. It was shown that ratio of the heat
loss via the electrodes to the heat loss via the adjacent BM is a function of the electrode diameter divided by the square of the sheet thickness. Therefore, for small welds with a diameter smaller than that of the electrode tip and thin sheets, the heat loss mechanism is dominated by the water-cooled electrodes, whereas welds larger than the electrode tip and thicker sheets are cooled mainly through heat loss by the adjacent BM. Because of technical difficulties in the experimental measurement of the cooling rate, several analytical and finite element models have been developed to determine the thermal cycle of the welding process [22–24]. An analytical approach was proposed by Gould et al. [10] to predict the cooling rate during RSW:

$$\frac{\partial \theta}{\partial t} = -\left(\frac{\alpha \pi^2}{4 \Delta x^2}\right) \left(\frac{\theta}{\theta_p}\right) \left(\frac{\theta}{1 + \left(\frac{\Delta x}{\Delta x_E}\right)^{-1}}\right) \cos\left(\frac{\pi}{2 \Delta x^2}\right)$$

(2-1)

where $\theta$ is the temperature, $t$ is time, $\alpha$ is the thermal diffusely, $\Delta x$ and $\Delta x_E$ are the sheet and electrode face thicknesses, respectively, $\theta_p$ is the peak temperature in the spot weld, $k_E$ and $k_S$ are the thermal conductivities of the electrode material and steel, respectively. An increase in the sheet thickness decreases the cooling rate as it increases the distance of the molten area from the water cooled electrodes. Based on this analytical model, decreasing the sheet thickness from 2 mm to 0.8 mm increases the cooling rate from 2000 to 10000 K s$^{-1}$. Increasing the welding current, welding time and decreasing the electrode force reduce the cooling rate.

The solidification path of most of low carbon steels can be expressed as:

$$\text{Liquid} \rightarrow \delta_{\text{ferrite}} \rightarrow \text{austenite} \rightarrow \text{ferrite + bainite + martensite}$$

If the cooling rate is high enough, depending on the carbon concentration, it is possible for austenite to be formed directly from molten material without $\delta$ ferrite transformation [25]. Figure 2-21 illustrates the weld cross section and optical macrograph of the FZ and HAZ of resistance spot welded IF steel. The central region of the FZ is composed of equiaxed grains connected to the columnar grains that penetrate into the HAZ. The FZ microstructure is characterized as Widmanstätten ferrite (Figure 2-21b), while the arrows in Figure 2-21c show the presence of allotriomorphic ferrite in the HAZ.

Due to very high cooling rates of RSW, fully martensitic microstructure can be easily formed in the FZ of AHSS. Higher alloying element contents added to the chemical composition of these steels increase their hardenability and facilitate the formation of martensite even at lower cooling rates [27]. The microstructure of the FZ and FZ/HAZ transition zone for the resistance spot welded DP600 steel is shown in Figure 2-22 presenting a fully martensitic structure.
Figure 2-21 Optical micrograph of weld cross section (a), FZ (b) and HAZ (c) of IF steel resistance spot weld [26].

Figure 2-22 Electron microscopy micrograph of FZ (a) and FZ/HAZ transition zone (b) for DP600 resistance spot weld [12].
The FZ hardness is an important factor concerning the weldability of AHSS and mechanical performance of the welds for automotive applications. The hardness of martensite formed in the FZ of a spot weld is mainly but not only influenced by carbon content as well as cooling rate. Den Uijl et al. [28] carried out extensive work on the relationship between chemical composition and post weld hardness of high strength steels for automotive applications and developed empirical relations of hardness prediction for a specific welding process (i.e. RSW, laser welding and plasma arc welding). The following empirical equation was derived for the weld hardness of RSW with forced cooling time according to their work:

$$HV_{FZ} = 229 + 1088 \times \left( C + \frac{Si}{88} + \frac{Mn}{102} + \frac{Cr}{91} + \frac{Mo}{99} \right)$$ \hspace{1cm} (2-2)

Yurioka et al. [22] proposed the following expression for carbon equivalent (CE) as an indication of the hardenability of the steel:

$$CE_Y = C + A(C) \times \left[ 5B + \frac{Si}{24} + \frac{Mn}{6} + \frac{Cu}{15} + \frac{Ni}{20} + \frac{Cr+Mo+Nb+V}{5} \right]$$ \hspace{1cm} (2-3)

where

$$A(C) = 0.75 - 0.25 \tanh[20 \times (C - 0.12)]$$ \hspace{1cm} (2-4)

![Figure 2-23 CE_Y versus FZ hardness [27].](image)

The accommodation factor of A(C) allows the CE_Y to be applicable for a wide range of carbon content from 0.02 to 0.2 %wt. Figure 2-23 shows the CE_Y and average FZ hardness of different steels gathered by Khan et al. [27]. It is shown that the FZ hardness increases with richer chemistry of the BM with higher CE_Y. CE_Y shows a linear relationship between fusion zone hardness and BM chemistry (r = 96:1%). Extracting the linear relationship between FZ hardness and CE gives the following equation:

$$HV_{FZ} = 630CE_Y + 188$$ \hspace{1cm} (2-5)

where HV_{FZ} is FZ hardness and CE_Y is carbon equivalence calculated using Eq.(2-3).
2.5 HAZ softening

The HAZ hardness is also an important parameter that effectively determines the failure mechanism and weld strength. Several investigations reported the decrease in the hardness of the SC-HAZ with respect to the BM hardness in DP steel spot welds \([15,29–32]\). Figure 2-24 illustrates a typical hardness distribution of different weld zones with reduction of hardness at the SC-HAZ of a DP980 resistance spot weld.

![Figure 2-24 Hardness distribution of different weld zones of DP980 resistance spot weld \([36]\).](image)

A comprehensive study was carried out by Baltazar et al. \([33–35]\) using a nanoindentation technique. Figure 2-25 shows the microstructure of the SC-HAZ at different distances from the \(Ac_1\) line together with that of the BM of resistance spot welded DP980 steel. The \(Ac_1\) line was determined as the boundary between the IC-HAZ and the SC-HAZ. The SC-HAZ at a distance of 100 µm from the \(Ac_1\) line was characterized as a ferrite matrix and severely decomposed martensite or tempered martensite (TM) as shown in Figure 2-25a. Sub-micron particles arising from nucleation and growth of carbides are shown inside the broken martensite phases by arrows. Similar microstructure with no distinguishable change was found at 200 µm from \(Ac_1\) (Figure 2-25b). The microstructure of the SC-HAZ at 400 µm from the \(Ac_1\) line was characterized by more clear boundaries of martensitic phase and finer decoration of carbides (Figure 2-25c). Considerable reduction in the volume fraction of sub-micron sized particles were detected at 600 µm distance from \(Ac_1\) (Figure 2-25d). At a distance of 800 µm, sub-micron sized carbides can barely be seen. However, the martensite phase still shows decomposed characteristics compared to that of the BM with solid and undecomposed martensitic phase dispersed in the ferrite matrix (Figure 2-25e, f). It can be inferred that the tempering of martensite is responsible for the reduction of hardness in the SC-HAZ.
Nanoindentation testing was performed at different distances from the $A_c_1$ line toward BM. The change in the hardness of martensite, tempered martensite and ferrite at different distances from the $A_c_1$ line towards BM is shown in Figure 2-26. The nanohardness of martensite in the BM was measured as $7.2 \pm 0.8$ GPa, while the nanohardness of tempered martensite (TM) at the distance of 100 µm from the $A_c_1$ line revealed a reduction in hardness to $4 \pm 1.2$ GPa. There is gradual increase in nanohardness value of martensite from the $A_c_1$ line towards the BM. Slight reduction in nanohardness was observed in ferrite near the location of $A_c_1$. It was attributed to the reduction in the dislocation density of ferrite in the SC-HAZ.
The degree of softening is strongly dependent on the volume fraction of martensite in the BM. Figure 2-27 depicts the hardness profile over the weld zones for three different grades of DP steel. As shown, the SC-HAZ of DP780 and DP980 resistance spot welds are considerably softened, as opposed to the SC-HAZ of DP600 with no trace of softening. The degree of softening is greater for the DP980 resistance spot weld compared to that of the DP780 steel, which is attributed to the larger martensite volume fraction of higher DP grades that shows higher potential for non-isothermal tempering of martensite during the welding process.
An increase in the heat input also intensifies the degree of softening as the reduced cooling rate increases the time that the SC-HAZ is subjected to a non-isothermal tempering process [36]. It was also reported that the SC-HAZ is softer in the case of thicker sheets because of the lower cooling rate of the spot weld [31]. The BM chemistry is another parameter affecting the degree of softening as the BM with higher Cr or Mn content shows higher resistance to SC-HAZ softening [37].

2.6 Segregation

Embrittlement elements such as phosphorous and sulphur tend to segregate at grain boundaries of the solidifying fusion zone. In the case of materials with a high hardness (martensitic microstructure), the presence of small amounts of phosphorous and sulphur can have detrimental effects on the integrity of the welds. Experimental work carried out by van der Aa et al. [38] identified phosphorus segregation as one of the main causes for brittle weld metal failure during cross-tension testing of resistance spot welds. Elemental segregation leads to decreased coherency between the grains in the nugget zone of the spot weld. Two mechanisms aggravate the segregation effect: the solidification mechanism and widening of the solidification trajectory. In the case of low alloyed steel with the carbon as the main alloying element, the primary phase formed during the solidification is δ ferrite. Further cooling leads to the transformation of the δ phase to austenite, which subsequently transforms to martensite. Segregation occurs at the first grain boundaries during liquid to δ transformation. It is believed that the subsequent transformation of δ to austenite replaces the initially segregated boundaries with the ones depleted of P and S. Steels with higher alloying element contents solidify directly from liquid to austenite and finally martensite. This solidification path retains the boundaries that are rich in P and S. With increase in alloying element content, especially carbon, the solidus temperature decreases, while the liquidus temperature is less affected, which results in wider solidification trajectory leading to more segregation of P and S at the grain boundaries [39,40]. Amirthalingam et al. [41] studied the cross-tension properties of three different high strength steels labelled as CP, 2CP and CPB. CP steel contains 0.07 wt% carbon and 0.08 wt% phosphorous, while 2CP with same amount of P has carbon twice that of the CP steel (0.14 wt%). CPB has the same chemical composition as CP steel with addition of 0.0027 wt% boron. Figure 2-28 illustrates the cross-tension strength of the three steels as a function of weld diameter together with plug ratio (%). The 2CP steel with the highest carbon content shows the lowest cross-tension strength, whereas the CPB spot welds have the highest maximum load and also the largest plug ratio. Based on the quasi-binary phase diagram, in 2CP steel, austenite forms from the liquid/δ-ferrite by a peritectic reaction. However, with the super-fast cooling rate of resistance spot welds, it is possible for austenite to form directly from liquid. The lower mechanical performance of 2CP steel was attributed to the higher carbon content and consequently higher carbon segregation at the grain boundaries of the resistance
spot welds. Phase field simulations also showed that the addition of boron to the chemical composition of the CPB steel is able to decrease the phosphorous segregation during the solidification process and thereby reduce the embrittlement of the weld and enhance the mechanical properties.

Figure 2-28 Cross-tension strength of three steels with different chemical composition[41].

Nippon developed the most widely used relation for Carbon-Phosphorous-Sulphur equivalent to predict the failure mode of the weld in peel mode tests [42]:

$$C_{\text{eq}} = C + \frac{Si}{30} + \frac{Mn}{20} + 2P + 4S < 0.24 \text{ (in %wt)} \quad (2-6)$$

Based on this empirical equation, the compositional limit to ensure PF mode during peel test is a $C_{\text{eq}} < 0.24$ for typical hold time (T) of 25 cycles. Increase in sheet thickness and decrease in hold time, reduces the cooling rate of the weld shifting the compositional limit to $C_{\text{eq}} < 0.31$ (Figure 2-29).

Figure 2-29 Effect of steel chemical composition on the failure mode according to Nippon [42].


2.7 Weld fracture toughness

The fracture toughness of the weld is one of the most important factors that strongly affects the strength and failure mode of resistance spot welds. However, its measurement is a challenging task due to geometry and sample size restrictions. Furusako et al. [43] carried out a computational study on the fracture toughness of the weld edge during a cross-tension test using an elastic-plastic fracture mechanics model. Assuming that the crack starts to propagate when the crack propagation driving force ($J$) around the nugget under a tensile load reaches the fracture toughness ($J_c$) of the nugget edge, they attempted to derive the value of $J$ and measure the value of $J_c$ of the edge during the cross-tension test. It was also assumed that the stress that is applied to the nugget increases not axi-symmetrically, but at four points $90^\circ$ apart in the same plane. Two kinds of virtual cracks were allowed to propagate from one of these points; interfacial cracks and vertical cracks. Figure 2-30 a and b show the simulated distribution of maximum principal stress at the weld edge under a load of 4kN during a cross-tension test for interfacial and vertical cracks, respectively. The virtual crack is opened during loading and the decrease in potential energy due to opening was divided by the crack area to calculate the $J$-value. The variation of $J$-value with nugget diameter under a load of 5 kN for each of the two types of cracks is shown in Fig. 2-30c. As presented, the $J$-value decreases under the same load with the increase in weld size in accordance with the experimental results. The $J$-value for the nugget diameter of $3\sqrt{t}$, when the crack propagates through the weld nugget, is slightly higher than the vertical crack. However, for the larger weld sizes ($4\sqrt{t}$ and $5\sqrt{t}$), the $J$-value for the crack propagating through the sheet thickness is larger than the interfacial crack value. Results obtained are in agreement with the experimental results show that the cross-tension strength increases and the failure mode changes from IF mode to PF mode with increase in weld nugget size. Therefore, they believed that it is possible to improve cross-tension strength by restraining the increase in the $J$-value or increasing the fracture toughness, $J_c$, of the weld edge.

Figure 2-30 Distribution of maximum principal stress during cross-tension testing for interfacial (a) and vertical (b) virtual cracks. (c) $J$-value dependence on the nugget diameter under the load of 5 kN [43].
They also used scale compact tension specimen (CT) to measure the $J_c$ at the weld edge of two types of 980 MPa steel sheets with different chemical compositions, one with 0.13 wt% carbon and another with 0.30 wt% carbon content (Figure 2-31). The $K_c$ values for 0.13 % C and 0.30% C nuggets were measured as 84 MPa.√m and 29 MPa.√m, respectively. The corresponding $J_c$ values were also converted as 30 kN/m and 3.7 kN/m, respectively. The cross-tension strength of two welds was measured as 2.4 kN for 0.30% C and 6.6 kN for 0.13% C. Consequently, it was shown that the cross-tension properties can be improved by the increase in the fracture toughness of the weld edge.

Figure 2-31 CT specimen preparation position (a) and appearance of miniature CT sample (b) [43].

2.8 Failure mode prediction

The weld diameter is the most important factor governing the failure mode of the spot weld. As discussed earlier, different empirical equations are used in the automotive industries to determine the minimum weld nugget size and predict failure mode. However, although these equations work well for low-carbon steels, they do not give reliable recommendations for high strength steels. It arises from the fact that these equations only describe minimum weld nugget as a function of sheet thickness and ignore other metallurgical and geometrical factors. Therefore, some attempts have been carried out to develop analytical or empirical models to determine the minimum weld nugget size for AHSS. For instance, Marya et al. [14] proposed an experimental model to predict minimum weld nugget size of dual-phase steel to guarantee the PF mode during tensile-shear test expressed as:

$$d_c = 0.53t^{3.22} + 8.48 \left(\frac{h_{max}}{h_{min}}\right)^{-1.24}$$  \hfill (2-7)
Any spot weld that has a diameter smaller than \( d_c \) is likely to fail via IF fracture. Conversely, any weld that has a larger diameter is expected to fail via the PF mode. Based on this model, thicker sheet requires a larger weld nugget to avoid the IF mode. Besides, higher \( H_{max}/H_{min} \) ratio (\( H_{max} \) maximum hardness (FZ), \( H_{min} \) minimum hardness, i.e., BM and SC-HAZ) reduces the required minimum weld nugget size for PF mode.

Van den Bossche [44] developed and analytical criterion to determine the critical weld nugget size to guarantee the PF mode spot welds during tensile-shear testing of AHSS as following:

\[
\left( \frac{D}{t} \right)_c = 1.5 \frac{\sigma_{yBM}}{\sigma_{yFZ}} \left( \frac{w}{t} \right)^{0.5}
\]

where \( \sigma_{yBM} \) and \( \sigma_{yFZ} \) are the yield strength of the BM and FZ. The model was developed based on the assumption that the fracture occurs in the area that first yields (the FZ or the BM). It also delivers the fact that for AHSS in addition to the sheet thickness, the material properties must be taken into account to determine the critical weld nugget and failure mode.

Pouranvari and Marashi [45] proposed another analytical model for the critical weld nugget during tensile-shear test by assuming that the driving force for the IF fracture mode is shear stress at the weld nugget and the driving force for the PF mode is the tensile stress around the weld nugget. Assuming a cylindrical nugget with diameter of \( D \) and height of \( 2t \), the failure load the IF failure mode (\( F_{IF} \)) can be estimated as:

\[
F_{IF} = \frac{\pi}{4} D^2 \tau_{FZ}
\]

where \( \tau_{FZ} \) is the shear strength of the FZ. For the PF mode another simplified assumption is considered as the onset of failure occurs when the maximum radial tensile stress outside the weld nugget reaches the ultimate tensile strength of the failure location. Thus, the failure load of the PF mode (\( F_{PF} \)) can be expressed by:

\[
F_{PF} = \pi D t \sigma_{PFL}
\]

where \( \sigma_{PFL} \) is the ultimate tensile strength of the failure location outside the weld nugget. The failure mode during tensile-shear test is determined by the competition of these two driving forces. It depends on which one reaches its critical value. The PF mode occurs if \( F_{PF} < F_{IF} \). Therefore, using equations 2-9 and 2-10, the critical weld nugget size can be defined as:

\[
D_c = \frac{4t \sigma_{PFL}}{\tau_{FZ}}
\]

An equation for the critical weld nugget size for the cross-tension mechanical tests was derived by Smith as [46]:
\[ D_c = 2.93 \left( \frac{\tau_{PF}}{K_F^{FZ}} \right)^{2/3} t^{4/3} \]  

(2-12)

where \( \tau_{PF} \) is the shear strength of the PF failure location and \( K_F^{FZ} \) is the fracture toughness of the FZ. It was assumed that the failure mode is a results of the competition between shear plastic deformation outside the weld nugget (PF) and crack propagation into the weld nugget (IF). Similar models were developed by Chao [9], Ku and Chiang [47] and Sun et al. [48] for the failure load and mode prediction of resistance spot welds. As presented, all the developed models rely on the information on the local mechanical properties of the weld such as hardness, yield strength, shear strength and fracture toughness. Therefore, gaining deeper insight into the mechanical properties of different weld zones is of paramount importance that leads to achieve prediction of the mechanical response of AHSS resistance spot welds.

### 2.9 Conclusion

Conventional \( 4\sqrt{t} \) or \( 5\sqrt{t} \) recommendations for the minimum weld nugget size are not reliable for the resistance spot welds of AHSS. AHSS are more susceptible to weld metal failure and it is essential to consider the metallurgical properties of the welds in the development of models for failure behaviour of spot welds. It was shown that the formation of brittle phase of martensite in the weld nugget of AHSS leads to higher hardness, higher brittleness and lower mechanical performance. High strength AHSS shows SC-HAZ softening due to tempering of martensite during the heating cycle of RSW, which makes the microstructural evolution and deformation response of the welds more complicated. The effect of embrittlement elements such as phosphorous and sulphur on the mechanical properties of the welds was discussed and it is shown that the higher carbon equivalent of AHSS makes their microstructure less tolerant of phosphorous and sulphur.

Despite the extensive work carried on the RSW of AHSS, it still lacks a deeper insight into the process-microstructural-property correlation of spot welds. Martensite as the predominant phase formed in the FZ of AHSS plays an important role in the mechanical response of the weld in cross-tension or tensile-shear loading. A more comprehensive investigation on the microstructural characteristics of martensite can be helpful to better understand the deformation behaviour of AHSS spot welds. Besides, as presented in section 2.8, all the analytical models developed for the prediction of the failure mode for AHSS are based on information about the local mechanical properties of different weld zones including the FZ and HAZ. In-situ small-scale mechanical tests can be a rigorous approach to evaluate the mechanical behaviour of different weld zones to make an accurate correlation between the welding parameters, microstructure and mechanical response.
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