Fatigue Strength of Friction Stir Welded Joints in Aluminium

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Abstract

Solid state Friction stir welding (FSW) is of major interest in the welding of aluminium since it improves the joint properties. Many applications where Al-alloys are used are subject to varying load conditions, making fatigue failure a critical issue. In the scope of this thesis, the fatigue performance of friction stir welded AlMgSi-alloy 6082 has been investigated. Static and dynamic properties of different joint configurations and welds produced with varying process parameters have been determined. Microstructures of fractured surfaces have been studied to evaluate the effect of weld discontinuities on fatigue. The mechanical strength of the friction stir welds was set in relation to that of conventional fusion welds, and that of other FS welded Al-alloys.

The friction stir process produced aluminium butt welds with high and consistent fatigue strengths, which exceeded the strengths of similar fusion welded samples. A smooth weld geometry showed to be of great importance for the fatigue performance, favouring the friction stir welds. Welding speed in a tested range of 0.15-1.4 m/min had only a modest influence on the properties of the friction stir welds; properties were not deteriorating at the highest speed. The softening of the alloy around the weldline was modelled. A fair description of the hardness profiles across the weld was obtained. At a low and high welding speed a full and partial softening respectively was predicted, indicating that full softening is not required to obtain a flawless weld.

In case of friction stir overlap welds, tool design is even more important than in butt welding to secure weld quality. A broad tool shoulder with a concave pin end gave the best performance. In particular, the minimal influence on the sheet interface when welding with such a tool was beneficial for the fatigue strength. The stress distribution in overlap and T-type test specimens has been modelled. The stress intensity factors were determined. The corresponding crack propagation rates were in fair accordance with the experimental results. It was found that a simplified approach, developed to estimate ΔK for overlap spot welds, could be used also for friction stir overlap joints.

Keywords: Friction Stir Welding, fatigue, mechanical properties, welding speed, softening, crack propagation, overlap joints, stress distribution
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I dedicate this work to my wife Sophie and our son David.
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Additional papers that have been published


M. Ericsson, R. Sandström, J. Hagström, Fatigue of Friction Stir Welded AlMgSi-Alloy 6082, Proceedings of the 2nd Int. Symp. on Friction Stir Welding, Gothenburg, Sweden, June 2000, TWI Ltd.

Distribution of work

Paper 1
Sandström initiates the research. Sapa supplies the material and performs the welding. Ericsson conducts the mechanical testing. The metallographic exam is done at Sapa (Gränges Aluminium) and KTH by Ericsson with assistance from the staff. Ericsson evaluates the results, and writes the paper with assistance of Sandström.

Paper 2
Ericsson and Sapa define the project. Sapa performs the friction stir welding and CSM Materials Technology the MIG- and TIG welding. Ericsson conducts the mechanical testing and metallographic exam. Sandström models the softening behaviour of the alloy. Ericsson evaluates the results, and writes the paper with assistance of Sandström.

Paper 3
Ericsson initiates the research; crack propagation measurements are performed on welded material from Sapa. Ericsson evaluates the results, and writes the paper with assistance of Sandström.

Paper 4
Ericsson and Sapa define the project. Sapa supplies the material and performs the welding. Ericsson conducts the mechanical testing and metallographic exam. Lai-Zhe Jin describes the stress distribution in a finite element model. Ericsson and Sandström evaluate the stress intensity factors. Ericsson writes the paper with assistance of Sandström.

Paper 5
Sapa supplies the material and performs the welding. Ericsson conducts the mechanical testing and metallographic exam. Lai-Zhe Jin describes the stress distribution in a finite element model. Ericsson and Sandström evaluate the stress intensity factors. Ericsson writes the paper with assistance of Sandström.
1. Introduction

Friction stir welding (FSW) is a solid state welding method developed in the early 1990’s by TWI (the Welding Institute) in conjunction with a group sponsored project. The friction stir process is of major interest in welding of aluminium since it improves the joint properties. Aluminium has traditionally been viewed upon as difficult to weld. This is due to the low melting point and low hardness of the material. Conventional fusion welding methods such as meta inert gas (MIG) and plasma-arc welding often produce unfavorable cast microstructures in aluminium. Large deformations are caused by shrinkage in the weld metal and heat affected zones (HAZ). In addition extensive softening will occur in the HAZ, lowering the mechanical properties. Due to the comparatively low process temperature during FSW, approximately 0.7 to 0.9 of T_m, a large temperature gradient is avoided. Residual stresses are therefore kept at a low level. The pressure build-up under the tool shoulder creates a smooth weld, free of voids. Oxides and smaller inclusions are effectively disrupted and dispersed by the rotating tool. Since the forces exerted by the tool are large a powerful fixture is required to hold the components in correct position throughout the welding operation.

Its high strength-to-density value, as well as good formability makes aluminium a commonly used construction material. Additional benefits may include minimal corrosion attack in many environments, and a relatively low life cycle cost due to low maintenance requirements. Within the transportation industries there is a strong desire to minimize vehicle or vessel weight in order to reduce running costs, thereby supporting the use of “light” alloys. Also “unweldable”, high-strength aluminium alloys (Al-Cu-Mg and Al-Zn-Mg-Cu alloys) have been successfully joined. The implication of FSW as a joining method will often lower production times and manufacturing costs. Initially the method was used primarily for large, thin-walled Al profiles, e.g. in shipbuilding, offshore constructions, railway wagons and rolling stock. The areas of applications are constantly evolving, however, as a result of process developments. Recent innovations include “self-reacting” and 3-dimensional type equipment which broadens the scope of FSW beyond linear and planar joints. Also other light metals (e.g. Mg, Cu) and steels have been welded in encouraging initial trials. Design of the tool shoulder and probe is important, not the least in the
case of overlap and T-type joints where the formation of a good quality weld is dependant on the material flow in the sheet interface.

Many applications where Al-alloys are used are subject to varying load conditions, making fatigue failure a critical issue. In bridge constructions for example, friction stir welding has been tried since fatigue cracks quite frequently are initiated at the fusion welds. FSW has been shown to give improved fatigue resistance.

2. Aim of the work

The aim of the work was to evaluate mechanical properties and microstructures of friction stir welds. The commonly used Al-Mg-Si-alloy 6082 was chosen for the study. The properties have been set in relation to known data for other welded alloys, both FSW and conventional welds. Some issues that have been investigated are:

- Influence of microstructure and macroscopic features on fatigue initiation and propagation
- Influence of process parameters; welding speed and tool rotation speed on the joint properties. There is a continuous strive within the industry to enhance production speed
- Weld softening behaviour
- Crack propagation rate in the weld, HAZ and base material
- Mechanical strength and weld formation in FSW overlap- and T-joints welded with different tool designs
- The stress distribution in overlap welds and its influence on fatigue

Another purpose was to share light on the premises of the relatively nouvelle friction stir welding technique, including its advantages and limitations.
3. Survey on the fatigue strength of aluminium alloy welds

3.1. Influences on fatigue

3.1.1 Stress concentrators

There are several factors that make the weld the critical area for fatigue initiation and progression. The associated stresses can be characterized as either global or local depending on their extent. The weld is located at a structural (global) stress concentration. It typically introduces a severe notch, and frequently contains planar and/or volumetric discontinuities. The welding often generates high tensile residual stresses. Furthermore, the change in microstructure introduced by the welding procedure is often unfavorable. In general, the influence of local effects, such as local stress fields, defect conditions and residual stress state, is of great importance in determining fatigue performance. [1]

Welding flaws are classified as either planar or volumetric. A third category is shape imperfections, which covers weld toe undercut and misalignment of a joint. Planar flaws include cracks, lack of fusion, and incomplete penetration. Non-planar or volumetric flaws include e.g. slag inclusions and porosity. These factors act to produce severe internal stress concentrations, which may or may not override the effects produced by external factors such as structural discontinuities. [1,2]

![Fatigue fracture surface](image)

Fig. 1. Fatigue fracture surface. Fatigue has initiated around the large sites of inclusions seen in the left picture. To the right, fatigue striations, marking the successive advancement of the crack front, are visible within one of the voids.
It should be noted that the majority of cases of fatigue cracking in service are associated with the inherent stress concentration features of the welded joint geometry, and only a relatively small number with welding flaws [2]. Research on the effect of weld defects on the fatigue behavior of aluminium butt welds shows that the European Convention for Constructional Steel (ECCS) and the International Institute of Welding (IIW) recommendations are very conservative; giving nominal endurance limits nearly half of those produced in experimental programs [3]. This research has made it possible to quantify the effect of weld defects on the distribution of local loads in the sensitive zones of the weld. By numerical evaluation of the geometric parameters characterizing the welded assembly (e.g. undercut depth and radius in the case of an unsound weld) reference fatigue curves have been established where actual manufacturing quality of the welds is taken into consideration. Another mean of fatigue assessment has been proposed from measurement of flaws in laser welded Al-joints [4]. A previous procedure (based on British Standard PD 6493 and a document by the International Institute of Welding) was modified with regards to the criteria for flaw dimensions measurement and flaws interaction. Since gas pores may be considered as notches rather than cracks, a fillet radius different from zero was here adopted to give the stress intensity factors. It provided a better correlation with experimental results. In the context of a simplified engineering assessment procedure, the threshold nominal stress range was derived by means of linear elastic fracture mechanics (for planar flaws).

3.1.2 Parent material

Base alloy properties and alloy temper are factors considered to exert only a secondary influence on the fatigue performance of aluminium welds (the relation is closer at short fatigue lives). This is because a majority of the fatigue life of a welded joint is spent in the propagation of fatigue cracks to critical size, and crack propagation rates are relatively insensitive to changes in alloy composition, heat treatment, or temper [1]. D. Colchen showed an absence of base material effect (in between 5000-, 6000-, and 7000-series Al alloys) on the endurance limits of various welded assemblies, regardless of the defect types and material thicknesses he considered [3]. However, at least one investigation indicates that temper condition can have a considerable impact on the fatigue strength of common arc weldments [5].
3.1.3 Joint configuration

Stress concentrations associated with the weldment are besides residual stresses, the most significant factor in determining the fatigue life of aluminium alloy weldments. The stress concentrations are strongly influenced by a number of other parameters, such as joint type, internal and external geometry, and joint size [1].

Nonsymmetrical joint configurations, such as strap, tee and lap joints show the poorest fatigue resistance. This is due to the development of large secondary stresses. The weld metal itself provides a source of localized stress concentration. If there is an abrupt transition at the weld toe, i.e. the junction of the plate surface and the weld metal, the stress concentration will be high. By removing the face the fatigue life can be prolonged.

![Image of the five basic types of joints](image)

Fig. 2. The five basic types of joints

In general, if they are made according to established practice, the choice of filler metal and method of edge preparation exert little influence on welded joint fatigue behavior. Fatigue is relatively unaffected by the tensile strength of the welded material. Hagström et al though showed that the static and fatigue strength of Al 6082 TIG-welded square tubes was lowered by a filler metal under matching the base alloy
strength. In case of the dynamic performance, this was attributed to the smaller penetration and poorer groove geometry compared to that obtained with a matching filler wire. Elongation properties of the weld metal, relevant for the redistribution of stresses in the joint, were also named as possible reason for the observed behaviour [6].

Literature from recent years on new welding methods, e.g. electron beam, laser beam and friction stir welding, shows that the use of these may substantially improve the fatigue properties of aluminium weldments [5,7,8,42]. In current design codes and standards no distinction is made between welding method, temper condition and alloy composition with respect to the fatigue behavior (IIW Fatigue Recommendations, and BS 8118).
3.1.4 Weld softening

During welding of precipitation hardening aluminium alloys, microstructural modifications and property changes occur in the heat-affected zone (HAZ) of the weld. These are dependent on the thermal cycle experienced by the material. The most important effect of the thermal cycle caused by welding is strength reduction or softening. Softening can be caused by three mechanisms: dissolution of precipitates, growth of precipitates (overageing) and formation of intermetallic phases. Weld and HAZ softening plays an important role for the tensile strength, since static fracture often coincides with the weakest area in the material. The width of the HAZ, and consequently extent of softening, is determined by several parameters, the most important being material thickness, the amount of heat transferred into the material, and number of heat paths. Fatigue performance of the welds, although negatively affected by the softening, is more closely related to the consistency in weld quality and groove geometry. This is realized when comparing MIG and TIG welds, the latter having much wider heat affected zones but longer fatigue lives [9].

By post weld ageing treatment (PWAT), hardness for the heat-treatable aluminium alloys can be recovered. Through dissolution and ageing the strength of the heat-affected zone is enhanced by reprecipitation. This positive alteration of the microstructure raises the fatigue strength, approximately to the same extent as the introduction of compressive residual stresses by shot peening [10]. In fusion welds mere postweld ageing is not effective because of the solute loss in the matrix introduced by solute segregation to dendrite boundaries in these welds. Furthermore, postweld solution heat treatment (SHT) is not practical for many applications of Al alloy welds, which makes softening a great problem.

Time and temperature play a very important role in the precipitation hardening process of the Al-alloy. The initial increase in the mechanical- and fatigue properties is due to vacancies assisted diffusion mechanisms and formation of high volume fraction of Guinier Preston (GP) zones, which disturb the regularity in the lattices. During overageing the size of individual particles increases, but their total number decreases leading to fewer obstacles to the movement of dislocations. Aging at a temperature between 447-473 K for 6-8 hours has produced the best combination of mechanical- and fatigue properties in an unwelded 6000-series aluminium alloy [11].
According to common practice this temperature and time range may also be applied for the post weld ageing treatment [12].

In the zone within the HAZ border to the fusion zone, a certain hardening can be seen in a MIG welded Al-Mg-Si alloy. This may be explained by alloying elements going into solid solution, and/or, in the case of multiple welding passes, precipitates being dissolved and maintained in supersaturated solution following rapid cooling. In a subsequent pass, peak temperatures and heat gradients may be lower than those of the proceeding pass resulting in a certain amount of ageing for this zone [13].

3.2. Improvement of fatigue performance

3.2.1 Significance of residual stresses

The level of residual stresses present in the weldment significantly affects fatigue performance [1]. If tensile residual stresses are present the crack initiation phase is shortened, resulting in a reduction in fatigue life. A way to improve this is to introduce compressive stresses in the surface material. Such techniques include peening, e.g. hammer-, shot- and needle peening. A recently introduced method is ultrasonic impact treatment (UIT) [14]. Peening techniques may also modify the weld toe geometry, e.g. eliminate cold laps in lap joints (see next paragraph).

Stress relaxation seems to be slightly less effective than peening in improving the fatigue strength of Al-alloy welded joints. In as-welded material the residual stresses are likely to be slightly compressive in the region of preferential crack initiation (nearby lateral edges of specimens). These compressive stresses are relieved through the relaxation, resulting in relatively low fatigue strength. A convergence between stress relieved and shot peened fatigue curves can be seen for high loads. This effect is believed to depend on the yielding of material zones immediately underlying the shot peened layer, causing a progressive redistribution of residual stresses toward tensile stress values [10].

The residual stresses acting transverse to the weld are typically low. They have no effect on the propagation rate of a crack along the weld [15]. In the case of a crack propagating perpendicular to the weld the situation is different. The longitudinal residual stresses are much higher, typically with a peak in the heat-affected zone for the aluminium alloys. To examine the effect of residual stress fields on fatigue crack propagation in welded plates, crack propagation rates were measured on large plasma
arc welded CCT (center crack tension) specimens and small CT (compact tension) specimens [15]. The presence of residual stresses in the former raises the actual value of the stress ratio R, and it also varies due to relaxation. Evaluation of the residual stresses acting in uncracked and cracked welded plates provided a comparison of the specimens. By using Walker’s law, which takes the actual effect of the stress ratio on crack propagation into account, the plots were found to coincide; the only difference between the two types of specimens was in the residual stress field. The effect of residual stresses on crack propagation can then be evaluated by taking into account the actual value of the stress ratio.

The significance of residual stress on the fatigue performance affects the applicability of data based on small test specimens, since residual stresses are not present in them. This has led to the proposal that scale correction factors can be used to verify the results of components of differing thickness. For large-scale beams the fatigue strength is influenced by the different degrees of restraint imposed on the weld by the structural fabrication techniques, as well as the complex stress distributions associated with the gradients [1].

3.2.2 Weld detail
As has previously been said, the fatigue life of weldments is to a great extent determined by the stress concentrations associated with the weld detail. Therefore, techniques that modify the weld toe geometry, and thereby reduce the stress concentration effect, improve fatigue strength. For additional effects it is common to remove slag intrusions and other defects that form at the weld toes.

Techniques for improvement of the weld detail include:
- Machining (including different peening techniques, see below)
- Grinding
- Water jet erosion
- Remelting techniques

In general improvement methods are used mostly as a means to upgrade repair welds, whose fatigue performance is lower than the original weld. However, weld improvement methods can also be viewed as weld quality enhancement that in some case can be integrated in production. For example, needle peening and ultrasonic peening have shown to be highly effective for fatigue life improvement of thin plate
aluminium lap-welded joints (giving an improvement of 60 and 70 % respectively compared to as-welded material) [14]. In the same investigation, two other methods: weld toe grinding and hammer peening, showed little effect on the fatigue strength. In the case of the hammer peening the result was due to additional bending stresses caused by the angular misalignment following from local deformation. A standard peening tool was used. It is normally used on steel specimens with larger plate thickness. That is to say it proved unsuitable for these thin plates, measuring 6-8 mm.

Fig. 4. Above: Effect of ultrasonic impact treatment (UIT) on material surface. Down to a depth of 12 mm, residual weld stresses and strains are reduced up to 50 % of their initial state. This together with the plastic deformation enhances the cyclic endurance of the material. In addition a wear and corrosion resistant “white layer” is created. Right: UIT being applied on weld toe in a bridge construction
3.3. Welding methods

3.3.1. Fusion welding – commonly used methods

Gas Metal Arc Welding (GMAW) including MIG (Metal Inert Gas) and TIG (Tungsten Inert Gas) has over the years grown to become one of the most popular and widely used welding techniques. However, in the case of welding aluminium there are some problems related to conventional arc welding. The melting of the weld area may produce severe heat distortion in thin metal sheets leading to residual stresses (especially in the high strength 2000- and 7000-series Al-alloys) [16]. A fusion zone consisting of an as-cast, coarse microstructure with solute gradients near the dendrite boundaries is obtained in the precipitation-hardenable Al alloys. The amount of porosity may also be high. Porosity occurs in aluminium alloy fusion welds mainly due to the rejection of hydrogen during the weld solidification. With excessive porosity in the fusion zone the fatigue of Al-alloy 6382 was seen to exhibit a greater weakness there than in the HAZ, due to the stress concentrating (notch) effect of the many pores. GMAW welding of Al-Mg-Si alloys may cause the formation of intergranular cracks in the HAZ due to penetration of low fusion constituents or diffusion of alloy elements present in the filler metal [13].
Fig. 6. Dendritic microstructure in an AA6082 MIG weld

Good reproducibility can be difficult to achieve with arc welding techniques if they are not fully automated. On the other hand, a manual metal arc process is generally the most reliable method for production of welds with a minimum of excess metal and a smooth transition at the weld toe, features that raise the fatigue strength. Furthermore, in recent years these techniques have been improved so that the main problems connected with them, such as residual stresses and defects, have been reduced [2].

Gas-tungsten arc welding (TIG) is often used when high weld quality is required, e.g. with thin wall thickness. TIG welding is a slower process than MIG. The operator can produce low-distortion welds, since the method gives precise control of welding heat. The weld is typically free of the spatter associated with other methods [17].

In Plasma Arc Welding (the plasma keyhole method) a tungsten electrode is also used, but a nozzle to form a highly collimated arc column constricts the arc. The plasma is formed through the ionization of a portion of the plasma gas. The keyhole mode of operation gives greater penetration. This allows a reduced amount of joint preparation, but also makes the method less tolerant to joint gaps and misalignment. The process produces high weld quality similar to TIG. Results show that in plasma arc welding there is no detrimental effect in fatigue crack propagation rate as
compared to unwelded material, which was seen for TIG welded specimens [13]. In addition, plasma arc welds have demonstrated superior fatigue performance compared to MIG welds during S-N testing. This is concluded to be mainly due to the larger root radius of the former [2].

3.3.2 Point shaped joints

The main joining methods used in automotive and airplane construction are resistance spot welding (RSW) and riveting respectively. Both industries are currently struggling with needs of cost reduction. Today there is a strong interest in the use of aluminium alloy sheet for car applications, particularly the body. Aluminium body structures offer significant weight reduction compared to equivalent steel structures. A unibody construction, similar to present steel designs has been suggested [18]. This approach features a process called weld bonding, a combination of adhesive bonding and resistance spot welding. For the development of the process the effect of weld parameters on spot weld properties has to be established more thoroughly.

The fatigue strength of point shaped joints is expected to depend linearly on the distance of the crack from the centre of the joint. Experiments show that the tensile shear fatigue strength of spot-welded joints increases with increasing nugget diameter [19]. Fatigue fracture is, as in general with aluminium alloys, transgranular. Grain coursening did not occur in the HAZ during spot welding of an Al-Mg alloy. Grain boundary melting, on the other hand, did occur resulting in coarse grain boundaries (indicating that spot welding increases the tendency towards intergranular tensile fracture for Al-alloys, as do other welding methods).

Crack initiation for spot welds is assumed to come very early in life, at the location where the binding between the two joined profiles stops and where a sharp notch is present. A smoother geometry at crack initiation for the self-piercing rivet joint may lead to its higher fatigue strength, noticed in a recent investigation [7]. The amount and character of residual stresses are also different in spot welding and riveting, in the latter caused by the metal working operation.

3.3.3 High-energy beam welding

As with the plasma keyhole method, one of the main advantages of the high-energy density fusion processes is their ability to make deep and narrow welds in a single
pass. The total heat input is much lower than that required in arc welding. This leads to a more narrow heat affected zone and noticeably fewer thermal effects.

With laser-beam welding very precise welds (relative to position, diameter and penetration) can be obtained. There is little or no contamination. Difficult-to-weld materials can be joined. A precise fit-up and alignment of the parts is much more critical than in ordinary arc welding though, since the focal spot diameter is very small [17].

Laser welds with only partial penetrations have shown higher fatigue strength than full penetration welds in the case of aluminium-extruded profiles joined with sheets (for car industry application) [7]. This can be attributed to the significant thinning of the sheets at full penetration.

3.4. Fatigue assessment – new trends

Procedures for fatigue assessment of structural aluminium alloys include e.g. component testing, nominal stress-, geometric stress- and notch stress methods, and the fracture mechanics approach. In order to assess the fatigue, S-N (stress-life) curves or calculations involving crack propagation can be used. At the present nominal stress and fracture mechanics methods are well established. For the notch stress method more basic investigations are needed [20]. New welding procedures, such as electron beam, laser beam and friction stir welding should be incorporated in fatigue recommendations.

The geometric structural (hot spot) stress concept is e.g. used by the car industry, for fatigue assessment of welded sheet steel structures. It is also well established and used for tubular structures, shipbuilding and pressure vessels [20]. The determination of geometric (i.e. structural) stress has by some been regarded as the most important information for fatigue strength assessment of a welded detail [21]. For complex structures the nominal stress cannot be defined and the stress field is multiaxial. Therefore a local stress is defined. The term "hot spot" refers to the critical point in the structure where fatigue cracking is expected. The geometric or hot spot stress includes all stress rising effects of a structural detail excluding the non-linear peak stress caused by the weld profile itself, and it is provided by finite element calculations on the whole component using some special meshing rules. The verification is performed by comparison with the Wöhler S-N curve associated with
the base material and the welding process. Few codes propose design values for aluminium based on geometric stress however. Partanen et al. [22] published a compilation of aluminium data for hot spot assessment, indicating that the fatigue class FAT40 (40 MPa at 2 million cycles and slope m=3) could be recommended in the case of thin plates and extrusions. In a current paper, a method to apply the hot spot approach for welded aluminium automotive components is presented [23]. This method compiles previous results and assesses new ones. Multiaxial stresses in the weld are taken into account by using a criterion extended to the hot spot concept [24], since it is often limited by the use of only the maximal principal stress at the weld toe. The S-N curve proposed by Partanen is verified by a set of experiments. Although in principle the hot spot concept should only be used for weld toe failures, the experiments show that the fatigue performance of welds exhibiting root failure is equivalent to that of welds exhibiting toe failure, provided that the root failure mode is unavoidable for the structural detail.
4. Friction Stir Welding (FSW)

4.1 The FSW process

4.1.1 Operational basics

Friction Stir Welding (FSW) was invented by TWI in 1991. The method was further developed through a group sponsored project in conjunction with several interest companies. Friction welding, however, has existed much longer. The ability to use high-pressure frictional interaction to locally develop a heavily plasticised extrudant-like zone was early recognised as being of engineering use.

![Diagram of friction stir welding process](image)

Fig. 7. Schematic illustration of the friction stir welding process

In FSW a cylindrical, rotating (non-consumable) tool is used, constituting a shoulder with a central probe. The probe or pin is being forced down between the two workpieces (Fig. 6). Sometimes a hole is drilled were the probe should enter. The pressure and friction heat of the rotating tool plasticises the material in the weld. As the tool moves along the weldline the material is stirred by the probe, forming a consolidated joint in its wake. Some probes have flutes, which are designed to produce a smooth material flow and enhanced vertical mixing in the weld region. Being a solid-state
process, the material is never allowed to melt. When welding starts, a steady state condition is rapidly reached in which the temperature at the tool contact section and around the stirring pin remains constant. This is due to a dynamic balance between heat conduction and heat generation at the interface [25]. Since there is no melting of material, some problems associated with conventional fusion welding are avoided; the formation of a cast microstructure and subsequent shrinkage of the weld zone during solidification. Due to the relatively low energy input, heat distortions are kept to a low level, and thereby the amount of residual stresses. Under optimized welding conditions, the pressure build-up under the tool shoulder creates a smooth weld, free of voids. Oxides and smaller inclusions are effectively disrupted and dispersed by the rotating tool. The weld top surface is smooth. The concentric tool marks and stirred up flash at the end of the tool shoulder constitute the only macroscopic stress concentrations.

The welding speed is adjusted to the dimensions and the alloys that are used. It is also affected by the spindle speed. Single pass butt joints with aluminium alloys have been made in thicknesses ranging from 1.2 mm to 50 mm without the need for a weld preparation. Thicknesses of up to 100 mm can be welded using two passes, one from each side, with 6082 aluminium alloy [25].

![Image](image_url)

**Fig. 8.** Friction stir welding, tool traversing to the right. A semi-circular pattern marks the successive advancement of the tool shoulder. This pattern is created since the tool system is angled (2 or 3 degrees behind the normal position) to increase the pressure at the back of the tool where the weld is formed [34].
4.1.2 Temperature cycle

Process temperatures have been found to vary between approximately 0.7 to 0.9 T<sub>m</sub>, where T<sub>m</sub> is the absolute melting temperature of the aluminium material [26,27]. A temperature of the order of 500°C is reached during a very short time approximately 2 mm from the weld centre. Further away from the weld, the base material in the heat-affected zone (HAZ) will reach 200-300°C, but only for a few seconds. Different alloys will have somewhat different temperature distributions [26] (in addition, T<sub>max</sub> can vary quite a bit).

More recently, a combined experimental/analytical method was developed for determining power, heat input, tool shoulder temperature and process efficiency in FSW. According to the trials in this investigation, heat inputs (kJ/cm) for FSW appear to be similar to those for arc welding despite the lack of melting during FSW. This seemingly contradictory observation may stem from differences in energy density distributions for the heat sources of the two processes. While the highest energy density for the arc is focused along the arc centreline, it is spread around the shoulder periphery for FSW. Furthermore, the intimate contact at the work-piece/tooling plate interface leads to increased conductive loss [28].

4.2 Metallurgy

4.2.1 Microstructure

The residual grain size within the FSW zone varies due to the variations in deformation and temperature. The weld and HAZ can be divided into three different zones, each with its distinctive structure:

- The weld nugget

Around the weld centre the structure is fully plasticised, and on the top surface heavily strained. This area recrystallises to a very fine-grained equiaxed structure, with a grain size of ~3-6 μm. The dislocation density in the weld nugget is low. Therefore it is concluded, that the intrinsic, stored energy associated with dislocations generated by prior deformation (during pressing) do not contribute significantly to the efficiency of the FSW process [27,29].

- The thermomechanically affected zone (TMAZ)
The TMAZ is a relatively thin region that surrounds the recrystallised nugget. This zone undergoes a similar heat treatment to the nugget, but less plastic deformation. The amount of strain energy is not high enough to initiate recrystallisation, and a partially deformed, inhomogeneous grain structure remains after welding [30]. The outer border of the TMAZ often coincides with the weakest zone in the material; the minimum is probably due to a combination of loss of hardening contribution from precipitates and decreases in dislocation density [31].

- The thermally affected zone (HAZ).

This zone is only affected by heat but to a lesser extent than the TMAZ. Therefore dissolution of hardening precipitates is not complete, and tends instead to lead to coarsening and strength loss.

The grain structure obtained by friction stir welding is finer than that produced by static recrystallisation. Therefore it has been suggested that deformation continues in conjunction with recrystallisation; so called dynamic r. (there have been some debate regarding the exact mechanisms involved). Some research teams suggest that the grains during FSW are reformed by dynamic recrystallization and static growth, which should provide a mechanism for superplastic flow [27,32]. Dynamic recrystallization can assist solid-state flow. Efforts have been made to characterize the deformation microstructure in 7075 T6 [32]. After post weld annealing (500°C for 60 min), the initial equiaxed fine grains were replaced by coarse irregular grains in the nugget. The base material was unaffected however. This phenomenon in the weld nugget bears close resemblance to secondary recrystallisation (or abnormal grain growth). In a textured material anisotropy in both the grain boundary energy and mobility is introduced due to orientation difference between high and low angle grains. However, in friction stir welded 7075 the texture components are distributed in such a way that high angle grain boundaries dominate the microstructure. Therefore the abnormal grain growth does not arise from anisotropy. Second phase particles (here dispersoids) have an inhibiting effect on normal grain growth, and are thus very important in maintaining a very fine-grained matrix. It has been shown, that grains with more than six faces or a size advantage of more than 1.4 have a potential of outgrowing their neighbors when the matrix grains are fully pinned. By electron back-scattered diffraction (EBSD) technique such grains were observed in the FSW weld nugget [32].
Fig. 9. Microstructure developed during friction stir welding. A. Unaffected material B. Heat affected zone (HAZ) C. Thermo-mechanically affected zone (TMAZ) D. Weld nugget (fully recrystallized, fine-grained part of the TMAZ). The different weld affected zones, from left to right: HAZ, TMAZ, and nugget are seen below. In the middle picture all three zones are represented throughout different thickness levels of the sample. Note the widely varying grain size.

Friction Stir Processing (FSP) is a new interesting technique which main objective is to control the microstructure rather than joint parts. It can be used to locally tailor the properties of a material. For fabrication of complex shapes using aluminium alloys superplastic forming is of great importance. The problem with this kind of forming is that it is limited to thin sheets, and the microstructure of the entire sheet is altered. For optimum superplasticity the goal is to create a minimum grain size with grain growth stability at superplastic temperatures. The fine grain microstructure achieved by FSP
(3 to 5 μm in 5 mm thick 7050 Al) is applicable to superplastic forming at moderately high strain rates and relatively low temperature (450°C). FSP can be applied selectively to thick section Al alloys without altering the performance characteristics of other structural parts [33]. The method can also be used to modify the microstructure and resultant properties of fusion welded aluminium alloys.

4.2.2. Softening

For precipitation-hardened alloys such as 6000-series (Al-Mg-Si-alloys), the strengthening particles in the centre of the weld will be redissolved into solid solution. Subsequent reprecipitation gives strength recovery in this area. Further out, where the temperature is lower, the precipitates will merely coarsen and thereby lose much of their hardening effect. The hardness distribution across the weld gives an indication of the width of the HAZ. Hardness will also decrease as a result of decrease in dislocation density in the weld zone in case of cold work.

The strength can be regained if an ageing treatment is carried out after welding. However, the low hardness areas with coarsened particles will not react to this. The reason is that the coarsened precipitates consume large quantities of the solutes (i.e. Mg and Si) so that strengthening precipitates can hardly be formed. Instead these particles have to be redissolved using a solution heat treatment. This is somewhat costly, and cannot be carried out e.g. for Al-composite structures. For T6 material the post weld heat treatment involves a secondary ageing; the elongation is not fully restored. The best weld strength is obtained from welding in the T4 condition (supersaturated solid solution) and then subject the material to a normal ageing. By post weld heat treatment (PWHT) at 185°C for 5 hours, a 90 % HAZ strength recovery may be achieved [34].

Although the typical grain size produced by FSW is significantly smaller than that of MIG weldments, precipitates in FS welded AA6082 and 5083 were found to be somewhat larger [35]. The unaffected base materials contained the coarsest precipitates. FS welds contained smaller but more numerous precipitates. The finest precipitates were found in the MIG weld metals. Both finer grains and finer precipitates will increase the tensile strength. It seems as the combined effect of the two strengthening mechanisms results in similar strength levels as observed for MIG- and FS welded samples.
Fig. 10. Above: Microstructure in FS welded alloy AA6082. The weld nugget is wider on the top surface where the tool shoulder is in contact with the work-piece. Below: Transverse Vickers hardness over weld, HAZ and base material as a function of the distance from the weld centre (mm). Hardness is measured on the middle of the sample. Please note the different scales in the two figures. The minimum hardness is located at the edge of the tool shoulder in the HAZ. Hardness in and adjacent to the HAZ fluctuates due to partial recrystallization. In the T4 temper condition, weld hardness has been restored by post weld ageing treatment resulting in reprecipitation.

The soft weld region, characterized by complete dissolution of precipitates, and the minimum hardness region in the HAZ have shown greatest hardness increase following post weld ageing [36]. The precipitation sequence is characterized by formation of spherical Guinier-Preston (GP)-I zones, which grow into needle-shaped precipitates with an increase in ageing time.
When the joints are free of defects, the static fracture locations are dependant on hardness distributions of the joints; fracture often coinciding with the area of minimum hardness. The area of maximum strain, and thereby fracture, may differ however, as was seen for alloy AA2017-T351 which fractured at the interface between weld nugget and TMAZ despite a hardness minimum in the HAZ. Although the maximum strain and plastic deformation were low in these welds, there was a large difference in microstructure between the weld nugget and the TMAZ [37].

4.2.3 Influence of process parameters on welding flaws

The surface appearance and weld formation is strongly altered by the applied process parameters, i.e. tool pressure and contact area, rotational and welding speed. These welding parameters are dependent on the alloy and its dimensions. Parameter settings, as well as tool size and design have to be chosen separately for each new alloy and wall thickness to be welded. Parameters for butt welding of most aluminium alloys have been optimized in a thickness range from 1.6 to 10 mm. Special lap joining tools have also been developed for aluminium with thicknesses of 1.2 to 6.4 mm [25].

Since the heat input is a function of the shoulder radius to the third power (i.e. 
\[ q \propto R^3 \]), and only linear dependent on the forces and revolution speed applied, the weld formation is strongly dependent on the selected tool size [38]. If the tool pressure is too low, enough heat will not be generated (dry friction). This will lead to lack in bonding and welds that are not sound. Similarly, when welding with high speed, there is a risk for small, unwelded grooves in the bottom of the weld bead. If the tool pressure is too high, mechanical distortion can occur. The tool shoulder pressure is a function of the rotational speed and the traverse welding speed. The probe plunge depth relative to the material thickness is important to secure sufficient plastic deformation in the root of the weld. If the plunge depth is too small, this will generate an unbonded area in the root and an early fracture during tensile or bend deformation will appear.
Here follows a list of common flaw types which can be generated during FSW if process parameters are not optimized (Figs. 11, 12):

- **Voids**; flaws that are totally subsurface, volumetric, and contains no material

- **Lack of penetration**; in a butt weld this flaw is characterized by the full thickness of the weld not being forged. It can be caused by inadequate plunge depth, too short a pin, or poor alignment of the tool relative to the joint line. The gap between the unbroken plate surfaces may be more or less fine (see Fig. 11).

- **Faying surface flaws**; small voids below the top of the weld that can be removed by light machining of the weld surface

- **Entrapped oxides**; so called “kissing”-bonds or “lazy S” flaws originate from incorrectly broken and stirred surfaces, leaving semi-continues layers of oxides in the weld. They provide little mechanical strength. Since the surfaces are fully in contact on a molecular level, these types of flaws are very difficult to detect.

Fig. 11. Common flaw types which can be generated during FSW [39].
Fig. 12. In this figure it can be observed that incomplete pin penetration (approximately 0.8 mm from the bottom surface) has lead to the creation of an entrapped oxide type flaw [39] (“kissing-bond”).

A flaw that, depending on size and severity, is unacceptable to design codes may be termed a defect. As long as process parameters are held within a given tolerance window, friction stir welding is known to generate few weld defects. This stated it is inevitable, however, that defects will occur as the process becomes more widely used and operating conditions are pushed to their limits. Problems that are related to fusion welding, like slag formation, extensive porosity and hot cracking will not be encountered, but conventional welding defects such as voids or lack of fusion can still be present. The lack of reliable nondestructive testing (NDT) methods to detect some characteristic defects, most importantly so called kissing-bonds, should be considered a problem. In an investigation by TWI, flaws were generated (in alloy 2014A) by variation of some weld parameters outside the established process window [40]. These were welding speed, welding force (forging pressure), tool pin height and surface oxide thickness. In case of limiting the forging pressure, a void, partially surface breaking, was formed along the entire length of the weld on the advancing
side. Void formation was, in part, caused by the plates moving apart and the tool lifting. A similar void was generated when the welding speed was raised four times above optimum speed (rotational speed held constant). By the form of the void; with an almost vertical edge at the end of the nugget, it was seen that material had been swept away by the rotating tool. The void thereafter remained unconsolidated, probably since the plasticized material had received less work and hence were cooler, and less easily forged by the shoulder. The location of the voids, between the weld nugget and the TMAZ (thermo-mechanically affected zone) on advancing side, has been described as a region of chaotic motion. Bendzsak et al predicted flow singularities in this region, and attributed them to be the source of weld defects [41]. The voids referred to above were detected by standard NDT-methods; X-radiography and ultrasonic inspection. This was not the case for other flaws encountered however. The reason is that they provide little disruption to direct passage of sound or electricity, and there is negligible foreign matter of a lower density that would enable X-ray detection. Faying surface flaws were created when the welding speed was increased. Root flaws that were detected by the metallographic sectioning, can presently be revealed only by a destructive bend test with the root in tension. A new technique however, ultrasonic phased array inspection, may be used to determine whether the weld has been correctly forged, thereby reducing the probability of entrapped oxides. Root and joint line remnant flaws can exhibit partial bonding; in 6082-T6 and 5083-O mechanical properties were not affected by their presence, but the defects are still clearly undesirable. In the case of critical applications, machining of the weld root may be an effective measure to remove such features [42]. It has been suggested that acoustic emission signals can be used in the sense of monitoring the entire FSW process. The band energy variation during the traversing of the tool over the defected region reflects the existence, location, and size of the weld defects [43].

As a result of improvements in the process, FSW production speed has gradually been optimized. The applicable welding speed has increased significantly in the years since the introduction. Today 2 m/min is a typical production speed in joining of extruded Al profiles. Previous investigations have shown static strength and ductility of AA6082 to be relatively unaffected by an increased welding speed, in the interval of 0.25 up to 1.4 m/min [35,44]. The same applied to alloy 5083; no clear influence of welding speed and plate thickness was seen between 0.046 and 0.132 m/min. At the
present moment 6000-series Al alloys (trials on 6082 of up to 5 mm thickness) can be welded with a maximum speed of 6-10 m/min with mechanical properties still intact [45,46]. The latter speed was tested in welding of Al 5754 and 6181 tailored blanks (a joint between two sheets of slightly different thickness). To avoid weld defects at very high speeds, the transverse speed, welding force and rotational speed have to be gradually raised to their final values with customized ramp functions. The process parameter box becomes smaller as the speed is increased. This means that small variations in welding conditions may affect the quality. For example, joint line remnants have been encountered at increased welding speeds, resulting from less disruption of the oxide per mm advance of the tool. Therefore it is important to reduce the oxide covering layers prior to welding at high speed, especially when bordering the established process window [47,40]. (As previously stated, tool design plays a basic role in the disruption of these layers.)

With tool rotational speed adjusted, the weld softening has been quite similar for widely varying traverse welding speeds (6082 and 5083 respectively). In case of alloy AA6082, comparisons have been made at 1, 3, and 6 m/min (the rotation speed was comparatively low; 80-330 rpm) and between 0.35, 0.75, and 1.4 m/min (different batches) [45,48,44]. The minimum hardness was the same in all instances; around 70 HV, and the extent of the HAZ was also similar, except at the slowest speed where it was wider. Hardness and local strain measurements on 6181 T4 at 1, 5, and 8 m/min show a wider HAZ for the slowest speed as well. Even though a partial softening can be discerned at fast speeds its effect on mechanical properties seems to be of a second order [44,46,49].

However, there have been investigations which suggest that the dynamic strength may be lowered at an increased FSW weld speed [35,49]. This needs to be considered since it is not fully consistent with conclusions in this thesis. Article no. 2 states that the influence of increased welding speed; in the range of 0.7 to 1.4 m/min (rotational speeds 2200 and 2500 rpm respectively) for AA6082, T6 and T4+PWAT, is only minor on the mechanical and fatigue properties of the material (For 6082, T6 the main fatigue life for the high speed was below low speed between 220-180 MPa max stress, but above towards lower stress levels). It is acknowledged, however, that at a significantly lower forward speed (0.35 m/min) the fatigue strength of the material could be improved, probably due to an increased amount of heat generated (in part the
result of an increased shoulder diameter here). This is also seen in the investigation of James et al on AA5083-H321 [49], where the fatigue limit at $10^7$ cycles is significantly decreased in the interval of 0.08 to 0.13 m/min (about 18% for as-welded and 11% for polished specimens), but further decrease is small from the latter speed up to the final speed of 0.2 m/min (as-welded specimens). For polished specimens the fatigue limit relative to the min strength at 0.13 m/min actually increases in this interval, which was explained by a more pronounced onion-skin texturing at that speed. The mentioned lack of fusion type defect (see further on fatigue properties) was not associated with crack initiation, and it occurred at a similar frequency at all welding speeds. The tool rotational speed was kept constant, which may have been detrimental in case of insufficient frictional heat generated. Besides the obvious reason of increasing productivity, one motive to raise the welding speed can be a desire to achieve FSW joints with very high strength. The aim is then to minimize the line energy in order to avoid thermal material degradation. A problem however, especially at very fast welding speeds, is to avoid a cold weld without sufficient bonding. Therefore, the rotational speed must be increased in relation to the forward speed, in order to provide enough heat to each unit length along the weldline. Void formation has been reported in situations where high travel speed was coupled with slow rotational speed [50]. The applied welding speeds in the investigations referred to were from optimum to high. Furthermore, there is an influence of alloy type on defect occurrence in FSW. “Kissing-bonds” were deliberately generated in AA5083-O, 5083-H321 and 6082-T6 by using weld parameters outside the process tolerance window [42]. Nondestructive testing including X-ray radiography and dye penetrants would not detect any of these root flaws. However, the larger flaws in 5083-H321 compared to the other two alloys caused these welds to fracture completely at 90° during bend testing. In addition to reduced ductility and ultimate tensile strength, the fatigue strength of 5083-H321 flawed welds was reduced relative to flaw-free welds, which was not the case for 6082 and 5083-O. (However, while the latter exceeded the Eurocode 9 design curves for single-sided welds, even the flawed 5083-H321 welds were compliant with them, indicating the conservatism of current fatigue design curves in regard to FSW.) Though there was some evidence that partial bonding at the root flaw interface could delay fatigue cracking, it was suggested that any root flaw that affects tensile properties or is deeper than 0.35 mm should be assumed unbonded [42].
4.3 Properties and performance

4.3.1 Advantages and limitations

The FSW process can easily be adapted for automation. It does not produce any major safety hazards such as welding fume, radiation or toxic gases, and the noise level is low. No joint preparation is needed. Taking MIG as an example of a commonly used fusion welding method, it is both flexible and productive in the welding of many alloys. In the case of aluminium and other “light” metals two major disadvantages exist however; the large deformations caused by shrinkage in the weld metal and HAZ, and the extensive softening in the HAZ. Due to the comparatively low temperatures during FSW a large temperature gradient is avoided. This results in a ductile weld. The risk to raise cracks by bending is small. FSW is often better suited for the joining of large, thin-walled profiles than conventional fusion welding or large-scale extrusion. Since there is no melting of material and thereby burning off of alloy substances, one gets a homogeneous weld without variations in alloying content. In addition, no filler metal is used (the ability to influence weld metal properties by choice of filler material is beneficial, but cost and complexity will be increased). For the reasons mentioned above, friction stir welding has been successfully applied to materials and components otherwise hard to weld.

Other advantages include:

- Friction welds exhibit excellent physical properties, matching those of the parent materials involved in many cases
- Highly consistent quality
- Flat surface without weld reinforcement or spatter; no grinding or brushing needed
- High productivity

Classic limitations associated with friction stir welding are [51]:

- Since the forces exerted by the tool are large, a powerful fixture is required to be able to hold the components in correct position throughout the welding operation
Since the method requires a stable welding machine with a powerful fixture, the welding equipment should preferably be stationary. This may be a serious limitation in the field of certain potential applications.

The method leaves an end hole when the tool is pulled away from the work piece. In many cases this hole can be cut off, but in other cases it has to be sealed using another method. For horizontal welds run on/run off-plates can be used. Results though show that a severe binding fault appears when the tool passes the slit between work-piece and plate. Therefore, material approximately corresponding to the thickness of the work-piece has to be machined away from the weld ends [52].

The back of the object must be accessible if 100% penetration is essential.

There is no good NDE method to detect lack of fusion in these welds, especially so called "kissing bonds", a defect where the surfaces show intimate mechanical contact, but weak molecular bonding.

Some technological aspects in friction stir welding are critical for the result and properties of the welds. The design of joint configurations, fixtures and clamping devices has to be established through substantial experimental experience. Parts to be welded must be clamped to a backing plate using a fixture, rigid and powerful enough to resist the considerable downward force exerted by the tool. This force and forces that try to split up the parts are related to the welding speed and to the material thickness [26]. A tight fit of the components to the backing is also important to control the tool plunge depth, which is a critical machine setting not to fall below a minimum value. The geometrical tolerances of the profiles are important factors that could strongly influence the weld result (extrusion tolerances are critical since no filler material is added during processing). A negative thickness offset greater than 0.1 mm for one of the work pieces will lower the mechanical properties considerably (causing a shortage of material to fully close the weld), especially with the offset on the retreating side of the weld [26]. Also straightness and flatness of the profiles must be well known and kept within narrow tolerances. Since the process uses no filler material the joints should be designed without gaps to avoid weld imperfections.
4.3.2 Corrosion performance

Only limited work on the corrosion behavior of FS welds has been reported. Tests on AA6082 showed encouraging results regarding corrosion resistance of the weld [34]. Testing on AA6013 and 6056 T6 revealed no susceptibility of the nugget area to intergranular, exfoliation, and stress corrosion cracking (in the area of the TMAZ and HAZ there was a pitting attack), whereas the base material showed attack [53]. However, studies on high strength Al alloys (2024-T3 and 7075-T6, used e.g. in ships and airplanes) have suggested that the weld and HAZ adjacent to the thermomechanically affected zone are susceptible to intergranular corrosion. It was established that Cu and Zn were depleted in the high angle grain boundaries of the recrystallized weld region. The depletion of Cu in particular tends to sensitize the grain boundaries and cause intergranular corrosion. In the high and intermediate ΔK regions fatigue crack growth rates in the weld and HAZ were up to two times higher in a 3.5% NaCl solution than those in air (7050-1T7451) [54]. Near the threshold ΔK weld and base metal growth rates were actually lower in the corrosive environment than in air. This was attributed to the corrosion product wedging phenomenon, i.e. the influence of corrosion debris in the crack tip on crack closure. In the low ΔK region, because the crack tip opening is small, this debris can wedge open the crack and thereby increase crack closure levels to “as high as 90 % of the applied maximum fatigue load”. Localized corrosion is affected by the presence of metallurgical heterogeneities such as second phase particles. One way to deal with this problem is to use surface melting to homogenize the microstructure. Therefore welds were subjected to surface laser treatment [55]. The rapid melting and resolidification of the surface layer (~10 μm) led to a refined grain structure. In addition the elements were maintained in solid solution, which also decreases the susceptibility to corrosion.

4.3.3 Fatigue properties

Several investigations agree in that the fatigue performance of friction stir welded Al alloys is as good or better as that of the commonly used fusion welding methods (MIG, Plasma-arc, Nd:YAG laser, spot welding) [5,7,8]. The fatigue of TIG welds has to a lesser extent been compared with FSW. While the latter was still quite novel a process, the two were found to be equal in fatigue strength; at 500 000 cycles the endurance limit reached 76 and 73 % of the parent material strength respectively. The TIG welded specimens were machined and polished prior to testing, while the FSW
specimens had no improvement of the surface. The fatigue testing was carried out in strain control with $R = -1$ to simulate structural constraint [56].

In early investigations it was shown that fatigue was sometimes initiated by internal defects such as lack of fusion or pores. Deterioration in fatigue life has been seen in welds containing porosity. However, with the current FSW technology the amount of porosity in the welds is very limited. James et al indicated that “onion-skin” forging-type defects (reminiscent to “kissing”-bonds in the root) can affect the fatigue life (in 5083-H321) by providing easy linking paths between initiated cracks [49]. The origin of these defects was not determined, but they tended to occur towards the end of the fatigue region or in fast fracture areas. For specimens subjected to different stress that failed at similar life, the principal cause of lower fatigue strength was found to be multiple crack initiation. Fatigue crack initiation was not associated with any specific inner defect. Instead surface tool marks served as initiation points for unpolished specimens. This notch effect of surface irregularities, like the concentric ripples and overlap weld toe on the advancing side of the weld (Fig. 13), is proclaimed to have a detrimental effect on S-N fatigue life [57].

![Image](image.png)

Fig. 13. Fatigue fracture in 6082, T6 FSW. The overlap weld toe seen in top of the specimen act as a stress concentrator (left). Fatigue is characterised by the formation of striations. The beginning of ductile overload fracture is indicated by the presence of dimples on the surface (right). A microcrack is seen in the centre, transverse to the fracture surface.
The overall good fatigue properties of friction stir welds have been attributed to a number of factors:

- The fine-grained weld structure, finer than the parent metal, which will slow down the crack growth rate.
- The smooth surface appearance at the top bed and root of the weld, giving longer time to crack initiation.
- The comparatively low amounts of residual stresses

Fatigue crack growth measurements give valuable information on the propagation characteristics of the weld and HAZ. Often a major part of the service life of welded components subjected to fatigue is related to crack propagation, whereas in laboratory S-N testing the initiation part dominates the fatigue life. Several investigations report that the crack growth rate of the FS weld nugget is either comparable [57,58], or slightly faster (2000 and 7000 series Al alloys; particularly at low ΔK, and low load ratio R) [54,59] than the base metal propagation. Tests where the weld and TMAZ/HAZ show higher resistance to crack propagation than the base metal have also been reported (6000 series alloys). In these instances the very fine-grained microstructure in the weld is given as a reason for this [8,53]. It decelerates the crack propagation rate, especially at high stress intensity ΔK values. However, the influence of two different phenomena on crack propagation curves needs to be considered; namely residual stress- and load ratio effects. Residual stress effects are important,

![Fig. 14. Crack growth rate as a function of stress intensity factor and stress ratio](image-url)
especially at low load ratios when the relative magnitude of the residual stresses is larger. This means that even small amounts of residual stresses may show an influence on fatigue crack propagation [57]. The phenomenon of crack closure is often used to explain the variations in crack propagation curves for different R-ratios (Figs. 14, 15). During the unloading portion of a load cycle there is premature contact between the crack surfaces while some tensile load is still applied. Consequently, Elber states, only the load range between the opening load and the maximum load affects the crack tip action [60]. That is, for low load ratios, the effective ΔK is lower than the applied value (see more on this in paper no. 3).

![Fatigue Crack Growth Rates](image)

Fig. 15. Various methods to correlate for the R-ratio dependence of crack growth curves (here Al 6061) [61]. All curves in the figure pertain to R=0.1. The curve furthest to the right corresponds to the applied ΔK. At a propagation rate above 10⁻⁵ mm/cycle the influence of crack closure is less than below. The ASTM 2% opening load method (marked by stars) underestimates ΔK_{eq} at near threshold growth rates, but provides the best correlation with “closure free” R=0.7 data (not in the figure) in the mid region.

Because of the high (roughness induced) crack closure levels, the fatigue crack growth rate in the HAZ of Al 7050 FSW was significantly lower and ΔK_{eq} significantly higher than those in the base metal and weld. In this case the HAZ region was severely overaged and had very large precipitates, which in conjunction with residual stresses and lower yield strength might have altered the crack-tip deformation process [54].
4.3.4 Residual stresses

Tensile residual stresses are detrimental for the mechanical properties. Residual stresses in FSW are higher in the longitudinal than in the transverse weld direction (similar to observations in fusion welds. The transverse direction is often the most critical loading direction.). In both cases there is an “M” like stress distribution across the weld. This was determined by cut compliance technique and by diffraction methods (X-rays or neutrons depending on the depth of the weld). The weld seam and the parent material adjacent to the HAZ contain small compressive residual stresses, while tensile residual peak stresses are located in the heat affected zone (6013, T4) [62,63]. Sutton et al reports of an asymmetric stress distribution with respect to the weld centerline, with the largest gradients occurring on the advancing side of the weld (possibly due to higher thermal gradients) in the proximity of the shoulder diameter. Here sharp gradients in microstructure exist. The largest effective residual stresses were located just outside the pin diameter, extending from the weld top side towards mid-thickness. The lowest nugget tensile stress, as well as the largest compressive stress component (through-thickness normal stress) was measured near the root side of the test specimen. It was found that grain size varied in the vertical direction of the weld nugget, decreasing from top to root. It is most likely due to the higher heat input near the top surface which causes additional grain growth in this region [64]. Weld root transverse residual stresses have also been measured in 6082-T6 and 5083-O using the centre-hole drilling method [42]. These were 13 and -3 MPa respectively. It was acknowledged that due to the less reliable method used, experimental error might be larger than the measured values. In any case they illustrate the generally low values for this area and direction. Zhang et al demonstrated that residual stress due to the friction stir process has little effect on fatigue crack initiation. Instead the initiation sites were determined by the stress caused by the specimen geometry [65].

The peak tensile longitudinal residual stresses in the HAZ have been measured at magnitudes of 20-50 % of parent material yield strength (in small samples, potentially larger in larger samples) or 30-60 % of weld material yield strength, i.e. higher values than were initially thought existing. The rigid clamping arrangements that impede contraction of the weld nugget and HAZ during cooling are given as reason for this. The peak stress values are only identified when a sufficiently high local resolution is used (X-rays for top and root faces and synchrotron radiation with a small gauge
volume for through-thickness measurements). Tensile residual stresses in the HAZ were found to decrease with decreasing transverse and rotational weld speed (1000-300 mm/min and 2500-1000 rev/min) [62]. This is explained by the lower cooling rates. Residual stresses are reduced with depth. Therefore peening techniques or machining of the surface can be used to diminish their influence.

4.4 Process developments

Several innovations have been introduced in recent years pertaining to the FSW technique. Perhaps the most important feature is design of the tool shoulder and probe (or pin) system. Design of the tool is critical to the formation of a good quality weld. In the case of overlap welding, a conical probe will not produce sufficient pressure at the sides of the weld, leading to lower strength. Also a conventional cylindrical threaded pin originally designed for butt welding will experience problems; most importantly sheet thinning and undisrupted oxide hooks breaking into the material. Skew stir welding™ is a generic term for a number of variants in which the axis of the tool is given a slight inclination (skew) to that of the machine spindle [66]. It enables the ratio between the swept material volume (“dynamic”) and the volume of the probe (static) to be increased. This can also be achieved by re-entrant features machined into the probe, so called flutes (Figs. 16, 17). By increasing the volume of plasticized material and improving the flow path the skew stir-techniques are beneficial for oxide disruption. The greater width of the weld nugget improves

Fig. 16. Various pin designs: I – tapered (conical) pin without flutes; II – tapered pin with shallow flutes; III – tapered pin with deep flutes; IV – cylindrical pin
mechanical strength in the case of lap and T-welds. With the so called Flared Triflute-design, three flute lands are flared out at an inverted angle from the core so as to increase the tip diameter. The flow path around and underneath the probe is enhanced. In addition, a tip profile in the shape of a three pronged whisk provides better fragmentation and dispersal of the surface oxides, Fig. 17 [25]. This is especially important when welding aluminium. Improvements in tool design also reduce the possibility of joint line remnant flaws during butt welding.

Fig. 17. (left) Another pin design based on the Triflute™ concept. (right) Flared-Triflute™ probe with tip profile for improved flow path and mixing action

Naturally there is a will within the industry to optimize the process with regard to productivity and profitability. A prospect of gains in the long-time economy often decides the willingness of manufacturers to invest in new equipment. High capital cost requirements of FSW may discourage potential users. Cost reductions associated with the FSW process would include:

- simplified pre-weld work – only degreasing of the plates is required
- no need for welding consumables, nor shielding gas or worker protection
- low energy consumption
- dimension accurate products after welding – no need for time consuming and difficult straightening work
- reduced need for repairs due to high, consistent quality (high repeatability)
Lower manufacturing costs and cycle times might justify an investment if the company’s time-schedule allows for the positive money-flow to be obtained. There are still barriers to overcome for a more general implementation though. One important limitation in FSW process development is the stationary machine design and powerful fixtures, made necessary by the large forces exerted. This restricts the field of potential applications. For minor scale aluminium workshops to fully benefit from FSW increased flexibility on production processes will be needed. This is the scope of much current research. Swedish company ESAB has presented a modular “plug&play” concept without fixtures and clampings for easy implementation.

The new Self-Reacting FSW (SR-FSW) process uses a tool with a bottom shoulder on the backside of the work-piece. The bottom shoulder is mechanically attached to the probe that penetrates through the work-piece, see Fig. 18. This concept eliminates the large stiff structure – the anvil – behind the parts to be welded. An actuator linked to the probe helps to adjust the load and position of the bottom shoulder, while a separate axis controls the load of the top shoulder. This system of independently controlled shoulders (“self-reacting”) results in an equal forging force on topside and backside of the part, the net force applied being essentially zero [67].

![Fig. 18. End view of SR-FSW pin tool set-up](image-url)
Besides elimination of a heavy back-up, advantages compared to conventional FSW include:

- increased forward speed due to the increased heat of two shoulders
- no risk for lack of penetration at the root side of the weld
- reduction of process loads

An additional benefit is the ability to weld thicker materials and tapered welds (thickness varies along the length of the weld path), since the position and effective length of the probe can be adjusted independent of the position of the shoulder. This feature also provides a mean to eliminate the key-hole at the termination of a weld, by slowly decreasing the probe length while the shoulder remains in contact with the part surface.

There is also craving for more specialized FSW equipment. One innovative example is the orbital system adapted for joining of aluminium pipes for gas insulated power transmission lines. It contains a separable ring-shaped cage with an integrated friction stir welding head and guide rings/clamping rings. The weld head is moved around rather than rotating the work piece. Development of 3-dimensional type friction stir welding equipment broadens its use beyond linear and planar joints. Free-Form FSW is an experimental, manual variant working in principal in the same way as operating a sewing-machine. The work-pieces are clamped together at one end of the seam and then hand-guided through the rotating tool by an operator. It was developed by researchers at Brigham Young University. For welding of e.g. car body members the integration with robotic units is significant. One of the main difficulties is to build in sufficient stiffness in the system based on the large force required. A spot FSW robot system has been developed for lap joints in aluminium plates [68]. It may be used as an alternative to conventional resistance spot welding (RSW) or riveting. A spot FSW gun similar to conventional RSW guns was developed and attached to a multi-articulated robot. The gun consists of a tool rotation unit and an axial loading unit. In addition to equal or superior quality welds, the FSW system showed lower energy consumption and maintenance cost compared with the RSW system. Recent developments in robot technology render higher payloads for jointed-arm robots possible, which will improve the implementation of FSW technology. In addition,
more powerful control systems allow for significant improvements in the force control system capability.

4.5 Applications

4.5.1 Recent developments

Much has happened since Norwegian shipbuilders first introduced FSW on a commercial scale in the welding of decks and hulls [e.g. 69,70]. Applications range over a broad spectrum, covering large scale welding within the transportation (shipbuilding-, aerospace-, railway-, and land transportation industries) and construction sectors as well as finer applications; e.g. wheel rims, electrical motor housings, heat exchangers, window frames, etc.

R&D activities so far have mainly been concentrated to investigation of conventional butt and lap friction stir welded joints. Several other geometries based on varieties and combinations of these two can be welded however, e.g. corner and T-section welds. For each of these joint geometries specific tool designs are required which are being further developed and optimised.

For all transport applications cost, weight and joint quality are important factors. The FSW process can potentially provide benefits in all of these areas. Therefore much research is undergoing to provide a better understanding of the advantages and limitations of friction stir welding.

4.5.2 Aerospace

In the airplane industry FSW is mainly used as a means of joining fuselage and wing structures, although it can be applied more extensively for the primary structure [59]. The majority of aircraft structures are assembled from high strength (2000 and 7000-series) Al alloys using riveted joints, which is extremely labour-intensive. (The rivet joint has benefits in the perspective of maintenance, where reparable and inspection possibility by non-destructive testing are positive features.) Welding is a more cost effective alternative, reducing production times and manufacturing costs. These high strength alloys are difficult to weld with conventional techniques (MIG, TIG and laser), but as a solid state process FSW offers property improvements relative to fusion welding.
The WA1FS project (Welding of Airframes by Friction Stir) has been initiated by main airframe manufacturers in Europe as well as various research institutes. Their main goal is to “advance the state of the art of FSW to enable the widespread adoption of the technology to primary structures in airframes” [53]. Besides the butt joint the T joint is also of interest. Joints are made in similar and dissimilar materials, in one thickness or as tailored blank. This latter joint system is extensively utilized by the automotive and airplane industries due to the optimized usage of materials. The static strength of friction stir welded alloys, relevant for airplane applications, has been in the interval of 80-90 % of base metal strength. Compared to an equivalent single-row riveted joint the FSW lap weld was found to be 2.4 times stronger. The FSW fatigue life equaled or exceeded the fatigue life of a riveted or bolted joint (in comparison with the latter the as-welded, non-machined FSW specimen had lower dynamic strength though). Also barrel pressure testing representative of the cabin was performed, in which cracks initiated in the weld nugget immediately turned out of the nugget and grew towards machined pockets and “flapped”. Flapping is desirable in that it improves structural residual strength compared to continual crack growth along the joint, and is readily detectable during visual inspections. In addition, it provides a warning as cabin pressure can not be maintained. More authentic, sub-element fatigue tests need to be conducted to enable a comparison with current solutions, particularly where stress concentrations are present near welds. [53,58,59].

A FSW process data specification standard has been produced to ensure interchangeable and reproducible data. Another object is to develop the modeling and simulation capability for FSW. It is necessary to further improve certain properties to allow full application. These include corrosion properties, residual stresses and distortion. A corrosion protection system working as a sealant to prevent interface corrosion in lap joints has been developed [59]. Microstructures can be improved by online active cooling/heating, or pro/post weld heat treatments. Thermal or mechanical tensioning is used to control residual stresses and distortion [53,71]. Reparability of FSW wing structures may be an issue, since attachment holes would need to be drilled to allow conventional reinforcing repairs to be made. Therefore the consequences of bolt holes being drilled through a FSW should be examined, to ensure no detrimental effects on the mechanical properties [58].
Fig. 19. The Eclipse Aviation’s 500 twin-engine jet is the first aircraft to make extensive use of friction stir welding as a joining method for primary structure. Locations for FSW lap joints include attachment points of ribs and stringers to wing skin, eliminating over 60% of the riveted joints. This has lowered manufacturing costs and cycle times.

4.5.3. *Overlap welds*

When it comes to overlap welding (Fig. 20), the stress situation is more complex than for butt joints, since there is a combination of shear and bending involved. Forces applied to the ends of lap joints result in eccentric loads in the connection area which can cause joint rotation [72]. In addition, tears may be formed by reorientation of the original interface between the work-pieces, Fig. 21. This is due to vertical metal flow.

Fig. 20. Conventional lap and tee joints with single or double side fillet welds
Therefore lap welds are more sensitive to the presence of surface oxides layers. Remnants of oxide film that break into the material in the form of tears may lower the mechanical properties. As a result, the shear tensile strength of the welded joints seems to be more dependent on the process parameters and tool choice than the strength of the unaffected base material [73,74].

![Schematic illustration of the cross-section of a FSW lap joint and possible tears](image)

Fig. 21. Schematic illustration of the cross-section of a FSW lap joint and possible tears

When the tool rotation is same as the probe screw direction (counter-clockwise if left-threaded screw), the metal near the probe is moved upward. According to a hypothesis about the metal flow in FSW overlap joints metal convection in the stirred and thermomechanically affected zones occurs in the opposite direction, however. The reason is that the joining zone is so tightly closed up by backing, tool shoulder and material [75]. Therefore the metal moves downward, and the interface is rolled downward. If the tool rotation and probe screw directions are opposite (clockwise if
left-threaded screw, more common) the metal moves upward, and the interface is rolled upward, Fig. 22a. The rolled interfaces lead to a reduction in the “effective sheet thickness” (EST) of the top or bottom sheet, Fig. 22b.

![Diagram showing metal flow and interface motion during lap FSW](image)

**Fig. 22.** a) Conception of metal flow during lap FSW [75]. b) Contrast material image showing interface motion in a FSW lap joint. There is a “pull up” on the retreating side, minimizing the effective sheet thickness (EST) of the upper sheet. On the advancing side there is an even more critical (due to geometry) interface pull down minimizing the EST of the bottom sheet. The advancing side pull up, illustrated in the previous figure, is undetectable here. c) Hammer “S” bend test to assess basic weld integrity of FSW lap weld.

Basic weld integrity can be assessed by a simple method, the so-called hammer “S” bend test, Fig. 22c. The characteristic notches leading to work-piece thinning will then be readily discerned. It has been shown that a counter-clockwise tool direction (with a Flared-Triflute™ type probe) can be used to reduce the upper work-piece thinning [76]. The shape of the edge notches was significantly improved in this case, resulting in a better weld integrity.
When welding sheets of different thicknesses, the worst situation from a mechanical point of view is associated with a tool rotation which gives a reduction of the thinner sheet. When this cannot be avoided however, it seems better to arrange the thicker sheet on top. It is supposed that the temperature around the interface then is lower, and thereby the motion of the interface, than the other way around. Mixed joints (2000-7000 Al series) of sheets of similar thickness have been evaluated in monotonic shear testing [74]. The advancing weld side was found to be subject of a majority of fractures; 68%, the others failing by shear through the nugget. Therefore the advancing side was eliminated by double pass (DP) welding; i.e. an additional weld was made in opposite direction but along the same line as the first, switching over retreating and advancing sides. By a double pass the mechanical strength is enhanced compared to a single pass overlap FSW. The weld quality in regard to oxide disruption is improved. Furthermore, in addition to the more favorable notch shape (more rounded, not as sharp) of two retreating weld sides, a wider interface weld width can be obtained by introducing a separation distance in DP-welding. This will decrease the amount of bending stresses. The static strength was increased with the transverse separation distance between subsequent weld passes, up to a distance corresponding to the width of the tool shoulder [74]. From that point a serious cavity was observed that is probably critical to fatigue properties.

As previously indicated the effective sheet thickness plays an important role in the case of FSW overlap welding. To maximize the EST, i.e. to minimize the amount of interface movement adjacent to the weld nugget, some measures have been suggested: A “cold” weld causes less vertical mixing on the retreating side. On the other hand, decreased rotational speed or increased welding speed may give coarse cavities or pores in combination with certain tools [76]. A shorter pin (given that the pin exceeds the top sheet thickness) also causes less influence on the sheet interface by vertical mixing, and hence maximizes the EST. The reason is that material flows less upwards near the bottom than at the middle of a pin [74]. Results indicate that the positive effect on the EST of a shorter pin is more important for the weld strength than the reduction in penetration depth.

4.5.3. Non-Al alloys

Although the main effort has been to optimize the FSW process for joining of aluminium and its alloys, preliminary welds have been successfully produced in a
number of other materials (e.g. copper, magnesium, titanium, and their alloys [77,78,79]). One of the interesting applications involving FSW of non-ferrous alloys is the sealing of copper canisters for long time deposition of nuclear fuel waste. Friction stir welding can be used to join the lid and bottom to the canister. The copper canister holds a cast iron insert for nuclear fuel. The severe demands on this type of cylinders; sealing and corrosion protection for 100 000 years, require a joint of extremely high integrity. For FSW to be used, a thorough understanding of the microstructure development is needed in order to avoid weld defects. Dynamic recrystallization is common in copper. By EBSD (Electron Back Scatter Diffraction) it was confirmed that the nugget had experienced deformation after recrystallization, since it had much less annealing twins than the root. The TMAZ was located 25 mm from the weld centre, where the grain misorientation was large [80]. Since the canister will have a significant thickness of ~50 mm, this requires the welding machine to be sufficiently strong and stiff to react to all process forces without excessive flexure [77].

Previously high melting temperature materials such as steels and nickel base alloys have not been friction stir weldable because of tool limitations. A steady state FSW process of these alloys, in addition requires careful management of temperature, heat flow, and flow stresses. Tests of tools made of various refractory metal alloys (containing e.g. tungsten, molybdenum, niobium and zirconium) have not been satisfactory since the test materials have lacked sufficient strength, hardness and abrasion resistance at elevated temperatures [81,82]. However, FSW of high temperature ferrous and nonferrous alloys is now feasible since the development of PCBN (polycrystalline cubic boron nitride) tools. PCBN is not new on the market. It was developed in the 1960s as a cutting tool material for use in turning and machining hardened tool steels, cast irons and super-alloys. Based on its thermal stability and superabrasive properties it was early recognized as a strong candidate for FSW at high temperatures. For example, welding of a high strength 0.29C-Mn-Si-Mo-B steel with this tool material have given encouraging results. The transverse tensile strengths of the FS weldments were greater than those of conventional gas-metal-arc weldments. Charpy V-notch toughness of the FSW stirred zone and HAZ was similar to that of the plate material, while toughness of the GMA weld metal and HAZ was generally higher [83]. While the tip of the FSW tool can be made of PCBN, the shank requires
other properties. In one instance tungsten carbide was selected for shank material due to its thermal conductivity and rigidity. A liquid cooled tool holder was added to aid in the removal of weld heat [82]. If too much heat is retained in the shank, it will soften and flex resulting in tool failure. On the other hand, if too much heat is removed, temperatures required to friction stir weld will not be achieved.

When it comes to FSW of high melting temperature materials, a comparison with the well established fusion welding techniques has to be made. Merits of FSW may include reduction in cost of postweld treatment, and advantage in the welding of advanced materials such as ultra fine grained steels or irradiated materials.
5. Summary of appended papers

Paper I. Fatigue Performance of Friction Stir Welded AlMgSi-alloy 6082

The commonly used precipitation-hardened AlMgSi-alloys (with aerospace, construction, and automotive applications) have been of main interest in the investigation. Aluminium alloy 6082 was studied with regards to weld microstructure, mechanical and fatigue properties. Two temper conditions, T6 and T4, were used. The T6 condition is obtained through artificial ageing at an elevated temperature of 170-200 °C. The naturally aged T4 condition was subjected to a post weld ageing treatment (PWAT), raising its mechanical strength to a level similar to T6 base material.

The material was fatigue tested at a stress ratio of 0.5 (70-110 MPa stress range). The T6 welds showed high and consistent fatigue strengths. So did T4 + PWAT, though slightly under T6 performance. This was somewhat unexpected since the latter material is statically stronger. There is however no strong correlation between base alloy properties and fatigue strength for aluminium welds. Fatigue data deriving from

![Image](image.png)

Fig. 23. FSW weld nugget and thermo-mechanically affected zone (TMAZ) at the sheet top surface. The nugget zone contains a very fine, recrystallised and equiaxed structure with a grain size of about 1-6 μm. The TMAZ contains mainly deformed and substructured areas, indicating that neither the deformation, nor the temperature is enough to recrystallise the material in this region. Tool marks can be seen on the surface.
a different T6 batch was also considered. The larger scatter in these results compared to the new series is believed to be due to different solution heat treatments and quenching parameters used at the two occasions.

The microstructure of fractured welds was studied in light optic and SEM microscope to analyse fracture modes and possible defects. The effect of welding heat on the distribution of precipitate particles (which can only be observed with TEM microscopy) in the weld and heat affected zone (HAZ) was determined through hardness measurements. Hardness drops in the HAZ due to dissolution and growth of the strengthening precipitates. For the T4 material dissolved particles are reprecipitated by the post weld ageing treatment. The T6 specimens failed in their weak weld/HAZ border outside the tool shoulder mark, on the advancing side of the weld. This position has also been observed for static fracture. They often received a more extended area of fatigue than T4 + PWAT, which frequently failed in the weld. SEM studies of fracture surfaces revealed fatigue striations, which occurred with a four times longer mutual distance in T6 than in T4 - PWAT. This is probably due to the lower hardness of the former. The studied fractures in T6 were transgranular with some intergranular areas. For T4 transgranular fracture was dominating. No such effects as oxides or lack of fusion was seen in the FS welds, the latter phenomena being the cause of fatigue crack initiation in some reported literature.

Fig. 24. (left) Fatigue fracture surface in 6082, showing typical initiation point at the edge of the specimen top side. The river pattern marks the successive advancement of the crack front. (right) Grain structure transverse to the fracture surface. The crack has mainly followed a transgranular path with some intergranular (along grain borders) areas.
Paper II. Influence of Welding Speed on the Fatigue of Friction Stir Welds, and Comparison with MIG and TIG

In friction stir welding there are fewer weld parameters to consider than in e.g. MIG welding. These are the most important ones: downward force, transverse welding speed, rotational (spindle) speed of the tool, and penetration depth of the tool pin. Since FSW is a fully automated process with high production potential, the influence of welding speed on weld properties is of significant interest. To avoid a “cold” weld without sufficient bondage enough heat-energy must be supplied to the material. This can be critical if the contact between tool shoulder and work-piece is not sufficient, e.g. if the welding speed is raised.

In this investigation two commercially applied welding speeds were tested with respect to mechanical and fatigue properties of the welds. A lower welding speed used in the previous investigation was also considered. In addition fatigue strengths of MIG-pulse and TIG welds were determined and compared with FSW results.

According to the results, welding speed in the tested range had only a modest influence on the properties of the FS welds; properties were not deteriorating at the highest speed. The specimens welded at extra low speed, deriving from a different batch, were somewhat thicker and had a broader weld. They showed the highest fatigue strength in the compilation. Besides the mentioned discrepancies it is possible that the increased amount of heat supplied to the weld at this low speed has enhanced its effective deformation.

All welds matched or exceeded design curves (Eurocode 9, BS 8118) for the category transverse fusion butt welds, see Fig. 25. Static strength and elongation being comparable, the TIG welds showed considerably higher fatigue strength than the MIG-pulse welds. The amount of weld defects was equally small for the two types. This shows that the geometric advantage of a finer weld bead (TIG) is more important than softening of the HAZ, which is greater in TIG welding being a slower (but more accurate) process. The FS welds had overall the best fatigue performance. The reasons for this are to be found both in macroscopic features and in the microstructures of the welds. Since the top and root surfaces are smoother in FSW than in fusion welding, the notch stress effect is less severe. The only major macroscopic stress concentrators are the concentric ripples created by the tool shoulder, and the characteristic, curl
shaped weld toes forming on the top surface. In addition, the higher heat input and melting of the fusion welds create a cast dendritic microstructure with a high degree of residual stress, which will significantly affect fatigue performance.

The softening of the alloy around the weldline was modelled. Using a model without adjustable parameters, a fair description of the hardness profiles across the weld was obtained. The softening in front of the FSW tool was also estimated. At the low and high welding speeds a full and partial softening respectively was predicted, indicating that full softening is not required to obtain a flawless weld.

![Stress range (MPa) vs. Number of cycles to fracture](image)

Fig. 25. Results of fatigue testing for FSW, MIG-pulse and TIG welded specimens. FSW data pertain to both low and high-speed tests. Design curves are given as a reference; Draft Eurocode 9 and British Standard 8118 design curves for transverse fusion butt welds (the Eurocode curve pertains to quality level B).

Paper III. Fatigue Crack Propagation in Friction Stir Welded and Parent AA 6082

A significant part of the service life of many welded components subjected to fatigue is related to crack propagation. Crack propagation characteristics of the friction stir welded Al-Mg-Si alloy 6082 were investigated. The electrical potential drop (EPD) method was used for measurements.
It is suggested that the good fatigue performance of FS welded specimens is partially due to comparatively low tensile residual stresses in the HAZ. In addition to that, the pressure under the tool shoulder gives rise to small, compressive stresses in the weld seam, which are favorable. The apparent positive influence of these can, however, be enhanced by the measuring method, especially at low load ratios R. Then the relative magnitude of the residual stresses is larger [57].

Two R values were tested: a low and a high. At low load ratio (R=0.1) and a low stress intensity ΔK the propagation rate in the weld was higher than in the parent material with a factor 3-5. At high load ratio (R=0.8) the crack propagation rate was similar for parent material and weld. The weld crack propagation rate was about the same at low and high R (somewhat faster at the high R close to fracture), see Fig. 26.

Paris law was used to describe the measured crack growth rates in the weld. In the case of the parent material, showing load ratio dependence of propagation curves, Forman’s law gave a reasonable fit. Crack closure may to some extent explain the difference in measured propagation rates in the parent material at R=0.1 and 0.8.

Fig. 26. Crack growth rate in the weld at R=0.1 and 0.8. Literature data for 6015 is given for comparison.
Paper IV. Fatigue Properties of Friction Stir Overlap Welds

Fatigue characterization of lap joints is not as easily carried out as for butt joints, since there is a combination of shear and bending involved. To avoid joint rotation, parts have to be sufficiently restrained. In friction stir welding of overlap and T-type joints the interface between the sheets is influenced by material flow around the tool pin; adjacent to the weld nugget the interface is rolled upwards. The shape of these “hooks” is of fundamental importance, especially for applications that are subject to fatigue. Design of the tool pin (or probe), which is the lower part of the tool penetrating the material, is important for the weld quality. Problems associated with conventional pins designed for butt welding pertain to hook formation and oxide disruption. If not thoroughly dissipated, streaks of Al surface oxide can break into the material by way of metal convection in the stirred zone, lowering the mechanical properties [76].

FSW lap joints of alloy 6082 in the artificially aged T6 condition were studied. Tool modifications based on the Triflute concept were used; the pin end being either convex or concave. Tool shoulder diameters of 15 and 18 mm respectively were utilised, producing four different weld series.

Static joint efficiency of the welds in relation to the strength of the base material was about 55 %, which by comparison with the literature is reasonable but not exceptional. By further improvement of the tool design or by performing a double-pass weld (a way to eliminate the dangerous advancing weld side, by switching over retreating and advancing sides in the second run) the static joint strength may be raised [74]. The fatigue strength of the overlap welds was clearly inferior to that of FSW butt welds. Their load bearing capacity was above the theoretical strength of an equivalent overlap spot weld, however. Fracture was initiated at the notches (hooks) in the highly stressed area where the weld cuts through the interface between the sheets. The notch tip on the advancing weld side experienced the highest local stress (due to severity and position).

The broadest tool shoulder with a concave pin end design gave the best fatigue performance. This is most likely due to the increased contact area tool-work piece, and due to an improved flow path provided by the hollowed out end of pin. In particular, since the tip of the concave pin was closer to the sheet interface (lower
bottom sheet penetration), the influence of vertical material flow on the interface was small. In all welds the amounts of undispersed oxides were low. After SEM analysis, the oxide disrupting capability was deemed somewhat better with the concave than with the convex pin.

The stress distribution in the test specimen was modelled. The stress intensity factor \( \Delta K \) was determined. The corresponding crack propagation rates were in fair accordance with the experimental results. It was found that a simplified approach, developed to estimate \( \Delta K \) for overlap spot welds, could be used also for friction stir overlap joints.

![Graph showing stress range vs. number of cycles to fracture](image)

Fig. 27. Fatigue of FSW overlap joints, 6082 T6, welded with concave end of pin. Compared with literature data for Al-alloy 5754

**Paper V. Fatigue of Friction Stir Welded T-joints**

Static and dynamic strengths of friction stir welded T-joints were determined. The specimens were loaded in two principally different ways. In the first mode the flanges of the T-specimen were subjected to tensile load (cross welds). In the second case a specially designed specimen holder was used to pull/bend the flanges in a direction parallel to the web. This type of loading simulates pressure changes in hermetically
closed boxes. Similar tool designs based on the Triflute concept were used as in the previous investigation.

The static strength of the T-joined cross welds was comparable with butt joints. The fatigue strength was about 50% of the static tensile strength at $10^5$ cycles, whereas over 90% for butt welds. Due to the significant bending moment, the T-joint strength in pull/bend testing mode was much lower than in cross-weld testing. In specimens welded with the concave end of pin, the notch radius between the weld nugget and the interface slit was particularly sharp on the advancing side. The convex pin end produced an interface less susceptible to failure on the advancing side, which was beneficial for the mechanical strength. For both pin designs the amounts of undispersed oxides were low. However, since the concave pin seems to enhance the material flow, it is deemed important to control the tool pressure in order to avoid excessive “hooking”.

The stress distribution in the T-specimen subjected to pull/bending was modelled. Three different methods were used to determine the stress intensity factor $\Delta K$. It was seen that a simplified approach developed for overlap spot welds could be used to estimate $\Delta K$ also for friction stir overlap type joints. The corresponding crack propagation rates were in fair accordance with the experimental results.

6. Conclusions

The friction stir process produced aluminium butt welds with high and consistent fatigue strengths, which exceeded the strengths of similar fusion welded samples. A smooth weld geometry showed to be of great importance for the fatigue performance, favouring the friction stir welds. Welding speed in a tested range of 0.35-1.4 m/min had only a modest influence on the properties of the friction stir welds; properties were not deteriorating at the highest speed. The softening of the alloy 6082 around the weldline was modelled. A fair description of the hardness profiles across the weld was obtained. At a low and high welding speed a full and partial softening respectively was predicted, indicating that full softening is not required to obtain a flawless weld. At a high load ratio ($R=0.8$) the crack propagation rate was found to be similar in 6082 parent material and FS weld.
The mechanical strength of friction stir overlap welds was clearly inferior to that of butt welds. The eccentric load and severe notch situation controls that. In this case tool design is even more important than in butt welding to secure weld quality. A broad tool shoulder with a concave pin end gave the best performance. In particular, the minimal influence on the sheet interface when welding with such a tool was beneficial for the fatigue strength. In addition, the heat energy supply was increased by the increased contact area tool-work piece. The stress distribution in overlap and T-type test specimens has been modelled. The stress intensity factors were determined. The corresponding crack propagation rates were in fair accordance with the experimental results. It was found that a simplified approach, developed to estimate $\Delta K$ for overlap spot welds, could be used also for friction stir overlap joints.

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Paper I
Fatigue performance of friction stir welded AlMgSi-alloy 6082

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Friction stir welding is a relatively new solid-state welding method, ideal to join large, thin-walled aluminium profiles. In this paper results from tensile and fatigue testing of friction stir welds of the hardenable Al-Mg-Si alloy AA6082 is presented.

For temper T4, post-weld age treated (PWAT), the fatigue strength was found to be lower than for temper T6 and with a steeper slope of the stress-life curve. The lower fatigue strength for the statically stronger T4+PWAT material was unexpected. Fracture in T6 initiated and propagated in the vicinity of the weak weld/HAZ border, directly outside the stirred up edge on the advancing (or shear-) side of the weld. For T4+PWAT many fractures occurred in the weld, often initiated on the advancing side.

Friction stir welding (FSW) is a relatively newly developed solid-state welding method suitable for joining large, thin-walled aluminium profiles. This is an alternative to large-scale extrusion or fusion welding, the latter causing distortion in thin metal sheets. There are many potential applications for friction stir welding currently attracting great interest, for example in car and aeroplane body sheets. Many components in railway wagons and trams, etc. can readily be welded by friction stir welding. The ship building industry was one of the firsts to commercially use the method on a large scale. Cooling elements and electric engines are other products currently being manufactured by friction stir welding [1].

The purpose of the present report is to determine fatigue properties of friction stir welded aluminium alloy 6082, in condition T6 (artificially aged) and in the T4 temper condition when subjected to a post weld ageing treatment (PWAT) to improve the static properties. Results from investigations in the literature on the fatigue strength of friction stir welded Al-Mg-Si alloys are summarised below.

Fatigue properties of welded AA6005, T4 have been compared for FSW, MIG, and the plasma-keyhole method [2]. The static strength of the FSW welded material was somewhat lower than that of the other two methods. However, the FSW welds had superior fatigue strength, coming close to that of the base material. The good surface finish of the welds was given as a reason for this.

Fatigue properties of FSW and TIG welded joints in AA6082 and 6063 T5, have been determined [3]. The TIG welded specimens, which were machined and polished before testing, showed fatigue strengths of 76% of the parent materials fatigue strength at 500,000 cycles. For the FSW welds fatigue strength was 73% of the parent materials, without any improvements of the surfaces. The testing was carried out as strain controlled fatigue with R = -1 to simulate structural constraint.

An investigation on fatigue properties of FSW and MIG welds in extruded plates of AA6082, T4 has been undertaken [4]. The results show an approximately 50% higher fatigue strength of transverse friction stir welds than of MIG butt welds. Crack growth data for the FSW welds indicates that crack growth rates are lower in the weld and heat affected zone than in the base material, this probably due to the more fine-grained microstructure in the weld region. All fatigue cracks in the friction stir welds had been initiated in defects located in the lower part of the welds. These defects were due to lack of fusion or appeared as pores.

Experimental

Material

Material of the aluminium alloy AA6082 (Al Si1MgMn) was friction stir welded in the T6- and T4 conditions. Chemical composition is given in tab. 1. SAPA Technology produced the welds. Alloy 6082 is developed for high strength, and contains manganese to increase ductility and toughness. The alloy is solution heat treated at 530-550°C, followed by quenching to room temperature. The solid solution then becomes supersaturated. If the material is stored at room temperature, this condition is referred to as T4 (precipitation through natural ageing). The T6 condition is obtained through artificial ageing at an elevated temperature of 170-200°C.

Parts were welded together with a speed of 350 mm/min and a rotational speed of the tool of 1000 rpm. Specimens were then cut by sawing from the welded plates perpendicular to the welding direction (cross welds). No preparation of the weld top surface was made.

The welds in the alloy with the T4 condition were further post

| Table 1: Chemical composition of alloy AA6082 (wt%) |
|----------------|-----|-----|-----|-----|-----|
|                | Si  | Fe  | Cu  | Mn  | Mg  |
| Typical        | 0.7-1.3 | ≤0.50 | ≤0.10 | 0.4-1.0 | 0.6-1.2 |
| Actual batch   | 0.91 | 0.18 | 0.01 | 0.45 | 0.60 |
| Cr             | ≤0.25 | Ni  | Zn  | Ti  | Ph  |
| Typical        |     |     |     |     |     |
| Actual batch   | <0.01 | <0.01 | 0.01 | 0.01 | <0.01 |
weld heat treated (PWAT), to enhance yield- and tensile properties to basically those of the base material in the T6 condition. This post weld heat treatment consisted of artificially ageing at 185°C for 5 hours, which gives reprecipitation of the hardening particles. About 90% HAZ strength recovery can be achieved in this way, resulting in a considerable increase in the strength of the T4 material.

Specimens cut transverse to the weld were first polished to study constituent particles, and then anodized to reveal the remainder of the microstructure.

Results

Hardness

Vickers hardness (load 10 kg) was measured on a transverse section of the weld along a line of approx. 100 mm, with its middle in the weld centre.

Tensile properties

Tensile properties were determined. The dimensions of the gauge length were 50x12.5x5.8 mm (length, width, thickness). Elongation was determined from measuring the relative increase in the length of the fractured specimens. A gauge length of 50 mm was used.

Fatigue testing

Fatigue testing was carried out in a servo-hydraulic testing machine equipped with an actuator of 250 kN load capacity. The dimensions of the pieces tested were 260x70x5.8 mm (length, width, thickness). A sinusoidal curve function was used and the stress ratio was set to 0.5. Average stresses in the interval of 105 to 165 MPa were used.

Microscopy

In collaboration with SAPA Technology, microscopy studies have been conducted on the base material and the welds, and later, transverse to the fracture surfaces of pieces which failed in different modes. Fracture surfaces have also been studied with SEM.

Fig. 1: Vickers hardness of AA6062 T6, series 2, friction stir welded. Advancing side to the left.

Fig. 2: Vickers hardness of AA6062 T4, friction stir welded + PWAT. Advancing side to the left.

Fig. 3: Mean stress vs. life for AA6062 T6, friction stir welded. Cross weld specimens. R=0.5.

Fatigue strength

The number of cycles to failure for the T6 condition versus stress approximately follows a straight line in the stress-life log diagram, see fig. 3. At 500,000 cycles the average fatigue strength is about 140 MPa for series 1, and 150 MPa for series 2. The scatter in data is small for series 2, but noticeably larger for series 1. This difference may be due to different solution heat treatment and quenching parameters at the two occasions.

For the post weld heat treated T4 specimens (involving artificially ageing with precipitation) the number of cycles to failure was below T6 values (fig. 4). This was an
unexpected result since the T4 + PWAT condition has higher yield and tensile strengths. At 500,000 cycles the average fatigue strength is about 140 MPa.

**Microstructural observations**

When looking at merely polished tests of the weld area in light optical microscope, two different types of particles can be observed. Apart from the large constituent particles there are smaller, irregularly shaped dispersoids. These are about 0.1-0.4 µm in diameter and are located in the centre of the weld. The dispersoids are enriched with Fe and Mn, and do not contribute to the strength of the material [6].

For the T6 material all samples went to fracture at the side of the weld which contains the rougher welding edge (fig 5), resulting from the rotating action of the tool [7]. This is the advancing (or shear-) side of the weld, where the relative difference in velocity between tool and workpiece is the largest and thereby also the stress. Fracture is at the tool outer shoulder diameter. As seen by the hardness graph (fig. 1) this is the softest area in the T6 material. The fatigue cracks propagated in many cases perpendicular to the loading direction along the weld/HAZ borderline. The cracks started at one edge of the specimen and then propagated into the material. This can be seen from the fatigue striations and river pattern. Ductile overload fracture has then occurred as characterised by the presence of dimples on the fracture surfaces.

For T4 + PWAT, fractures for a majority of the specimens are located in the weld area. As for T6, they are on the advancing side of the weld. In a few cases fracture was outside the thermomechanically affected zone. As opposed to T6, fatigue has started from the edges of the specimens only in half of the cases.

Fractured surfaces of T6 and T4 + PWAT were examined by SEM. The differences between specimens with fracture in weld and weld/HAZ respectively were studied. The T4 tests with fracture in the weld showed no obvious signs of fatigue (12X magnification), other than a few large diagonal verges which were also apparent to the unaided eye. At 5,000X magnification however, striations are visible (fig. 6). At even larger magnification one can observe microcracks in the material, transverse to the propagating fatigue crack.

For T6, the fatigued areas of the failed specimens have a more extended, and at 12X magnification, more characteristic fatigue structure. In some cases beach marks can be seen. Examples have been studied where fracture has grown diagonally into the material from the edge of the top and root surface, respectively. It is striking that the distance between the striations is considerably larger, approx. 4 times, for T6 than for T4. This ought to depend on the difference in hardness of the weld and HAZ for the two temper conditions. Microcracks are present on the surface, as in condition T4 + PWAT.

Transverse sections of fracture surfaces were examined in light optical microscope, to study the influence of microstructure on cracks and their propagation. All fractures in T6 were mainly transgranular with some intergranular parts. For T6 subjected to high load, it seems that initiation has taken place slightly inside the finely stirred area at the edge of the top surface and then propagated diagonally out. It rapidly reaches the weak thermomechanically affected area and then continues to the root.

For T4 + PWAT, many fractures took place in the fracture stirred area. It can be seen that the fracture pattern is mainly transgranular.

**Discussion**

Friction stir welds of aluminium alloy AA6082 have been examined with regard to tensile and fatigue properties. The results of the fatigue testing show, that the statically stronger, post weld heat treated T4 aluminium material has lower fatigue strength than the T6 material. This was not an expected result. In general, materials with higher static strength also have higher high cycle fatigue strength.

Results in the literature on the fatigue performance of friction stir welded Al-Mg-Si, as well as for other aluminium alloys, indicate that this is in general quite good. The fine properties are due to the mashing, stirring and forging action of the processing tool in FSW, which produces a weld metal with
finer grain structure than the parent metal. This will slow the crack growth rate, resulting in good fatigue strength of the weldment. Furthermore, the surface appearance at the top bed of the weld is smoother than for other commercial welds, giving longer time to crack initiation.

In the case of AA6082 friction stir welded in the T6 temper condition, the combined effect of the weak area in the HAZ, the edge of the test specimen and the stirred up welding edge can have served as a critical stress concentrator. This failure position has been seen for many of the specimens tested. Fracture is on the advancing side of the weld. On the retreating side the hardness decrease is slightly lower.

For T4 + PWAT the most common fracture is close to the centre of the weld. Since the crack in T6 propagates in softer material this may delay the process of crack propagation through blunting.

Conclusions

- AA6082, post weld heat treated (PWHT) in the T4 condition had lower fatigue strength than statistically weaker T6 friction stir welded material, which was unexpected.
- Fractures in T6 initiated and propagated in the vicinity of the weak weld/HAZ border, at the stirred up edge on the shear side of the weld. For T4 + PWAT many fractures were in the weld.
- For a majority of cases (all in T6) fracture started at the edge of the top or root surface and then diagonally propagated into the material. Fatigue striations were seen at high magnification both for T6 and T4. The distance between the striations was considerably larger, approx. 4 times, for T6 than for T4. This is probably due to the lower hardness in the T6 condition than for T4.
- The studied fractures in T6 were transgranular with some intergranular areas. For T4 transgranular fracture was dominating.

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References


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Paper II
Influence of welding speed on the fatigue of friction stir welds, and comparison with MIG and TIG

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Abstract

The objective of this investigation was to determine whether the fatigue strength of friction stir (FS) welds is influenced by the welding speed, and also to compare the fatigue results with results for conventional arc-welding methods: MIG-pulse and TIG. The Al-Mg-Si alloy 6082 was FS welded in the T6 and T4 temper conditions, and MIG-pulse and TIG welded in T6. The T4-welded material was subjected to a post-weld ageing treatment. According to the results, welding speed in the tested range, representing low and high commercial welding speed, has no major influence on the mechanical and fatigue properties of the FS welds. At a significantly lower welding speed, however, the fatigue performance was improved possibly due to the increased amount of heat supplied to the weld per unit length. The MIG-pulse and TIG welds showed lower static and dynamic strength than the FS welds. This is in accordance with previous comparative examinations in the literature on the fatigue strength of fusion (MIG) and FS welds. The TIG welds had better fatigue performance than the MIG-pulse welds.

The softening of the alloy around the weldline has been modelled. Using a model without adjustable parameters, a fair description of the hardness profiles across the weld as a function of welding speed was obtained. The softening in front of the Friction Stir Welding tool was also estimated. At the low and high welding speeds a full and partial softening is predicted, respectively.

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Keywords: Friction stir welding; Aluminium alloy; Welding speed; Fatigue; Tensile properties; Microstructure; MIG-pulse and TIG welding

1. Introduction

Friction stir welding (FSW) is a solid state welding method developed by TWI in the 1990s, and now being increasingly used in the welding of aluminium [1,2]. Application ranges from the production of small-scale components, such as cooling elements and electric engines, to welding of large panels, e.g. in ship building, in train wagons and trains, in offshore structures, and in bridge constructions.

FSW has several advantages over the commonly used fusion welding techniques. Following from its relatively low process temperature, below the melting point, the method is suited for joining thin or difficult to weld materials [3]. With no melting, the cast microstructure formed during conventional fusion welding is avoided as well as the weld zone shrink from solidification. Furthermore, there is limited risk for porosity in the weld zone, which is common in fusion welds. The FSW joint is created by friction heating with simultaneous severe plastic deformation of the weld zone material. The stirring of the tool minimises the risk of having excessive local amounts of inclusions, resulting in a homogenous and void-free weld. Since the amount of heat supplied is smaller than during fusion welding, heat distortions are reduced and thereby the amount of residual stresses. The deformation control is therefore easier [4].

Design of the FSW machine fixture and backing is of utmost importance since the forces exerted by the tool are large. The operation in itself is simple, once the right process parameters have been established, including most importantly downward force, welding speed, rotation speed, and penetration depth of the tool. To produce the best weld quality, these parameters have to be
2. Experimental

2.1. Material

The aluminium alloy 6082 (AlSi1MgMn) was investigated. Chemical composition is given in Table 1. The material was produced as extruded flat profile. Alloy 6082 is a high strength Al-Mg-Si alloy that contains manganese to increase ductility and toughness. The alloy was solution heat treated at 530–550 °C, followed by quenching to room temperature. The solid solution then becomes supersaturated. If the material is stored at room temperature natural ageing takes place. This condition is referred to as T4. The T6 condition is obtained through artificial ageing at a temperature of 170–200 °C. The welds in the alloy with the T4 condition were further post-weld aged (PWAT). This enhances yield and tensile properties to a level corresponding to that of the T6 temper. During this ageing reprecipitation of the hardening particles that have been dissolved during the welding occurs. The post-weld heat treatment consisted of artificially ageing at 180 °C for 7 h.

<table>
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<th>Cu</th>
<th>Mn</th>
<th>Mg</th>
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<th>Pb</th>
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<td>≤0.20</td>
<td>≤0.10</td>
<td></td>
<td></td>
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<tr>
<td>Actual batch</td>
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<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.01</td>
<td>&lt;0.01</td>
</tr>
</tbody>
</table>

2.2. Welding

The FS welds were produced by Sapa. The parts were joined with a welding speed of 700 or 1400 mm/min. The optimal welding speed depends on several factors: alloy type, penetration depth, and joint type being the most important [4]. Therefore the speed has to be determined individually for each new part to be welded. The probe diameter of the tool measured 6 mm, and the shoulder 14 mm. The penetration depth was adapted to fully penetrated butt joint in a material of 4 mm thickness. The rotation speed of the tool (spindle speed) was 2500 rpm for the high speed and 2200 rpm for the low-speed tests.

From each welded plate, one tensile specimen and three fatigue specimens were cut out. Prior to that, 50 mm at the beginning and 25 mm at the end of the plate were removed to exclude possible deviation from steady state during start and stop. It was checked that no such deviations were present in the test material. The cut specimens were milled at the edges, but no weld surface preparation was made.

2.3. Fatigue testing

The fatigue tests were carried out in a servo-hydraulic testing machine equipped with an actuator of 250 kN load capacity. Specimen dimensions were chosen according to the standard SS-EN 288-4. The dimensions of the fatigue specimens were 260 × 70 × 4.0 mm³ (length, width, thickness). The weld was transverse to the stress axis in the S-N specimen (cross-weld). A sinusoidal load-time function was used, with the stress ratio R set to 0.5. Average stresses in the interval of 105–165 MPa were tested. The oscillation frequency was in the interval of 9–15 Hz.

3. Results

3.1. Hardness

Vickers hardness was measured with a 10 kg load indenter (see Fig. 1). For FSW, T6 hardness drops in the heat affected zone (HAZ) due to dissolution (and growth) of precipitates. A minimum 70 HV is reached around the weld/HAZ border for the high-speed specimens. The width and maximum depth of the HAZ are
somewhat larger for the low-speed specimens (Fig. 1b)). Base material hardness is about 90–95 HV for the T6 and T4 + PWAT conditions. The T4 condition does not show a decrease in hardness in the weld due to the post-weld ageing treatment. The MIG-pulse welded specimens have a larger hardness decrease in the weld than FSW, T6 ones do. The minimum hardness is just below 60 HV, around the weld centre. The hardness decreases at the beginning of the HAZ followed by a small increase and after that another decrease in the area closest to the joint line (Fig. 1). The hardening effect in the zone within the HAZ bordering the fusion zone is due to age hardening, since hardness measurements were made some time after the production. In the case of TIG welded specimens a HAZ with a slightly larger width than that of MIG was produced. TIG welding has a higher heat input into the material per unit length.

3.2. Tensile properties

Tensile properties for base material and cross-welds are given in Table 2. For the MIG-pulse welded material, average yield and tensile strengths were 147 and 221 MPa, respectively, and for TIG 145 and 219 MPa. The elongation (A50) was 5.2% for MIG-pulse, and 5.4% for TIG. Tensile MIG and TIG specimens fractured in the HAZ.

For FSW, the welding speed had no obvious influence on tensile properties. The only significant difference was that for T4, FSW + PWAT; the elongation was 3.6% for the low- and 2.0% for the high-speed welds (see Table 2). Fracture was in the weld or in the weld/HAZ borderline.

3.3. Fatigue strength

3.3.1. MIG and TIG

The TIG welds showed considerably better fatigue performance than the MIG-pulse welds. At 500,000 cycles to failure the stress range was about 60 MPa for the MIG, and 70 MPa for the TIG specimens (Fig. 2). In the case of the MIG welds this indicates normal to high values [5,6]. The endurance limit for TIG is raised due to individual high values (and high values at a stress range of 90 MPa).

3.3.2. FSW

The FSW welds had longer fatigue lives than the fusion welds (see Fig. 2). For both temper conditions there are no major differences in fatigue performance between low and high welding speed specimens. Specimens welded with an extra low welding speed of 350 mm/min [8], show an average higher strength.

For FSW T6, at 500,000 cycles to failure, the stress range was about 90 MPa both for the low and high welding speed specimens (slightly higher for the high speed), see Fig. 3. For the extra low welding speed [8], the stress range is around 100 MPa. The scatter is generally small, with somewhat greater scatter for individual high-speed specimens. The results show that the fatigue performance does not decline with increased welding speed in the tested interval. In the present investigation, the low welding speed was beneficial only at high stresses. Regarding the fatigue results at the much lower welding speed investigated previously, see Section 5. It should be mentioned that at 80 MPa stress range two high-speed tests failed at the fixture and not in the weld, indicating higher values for the weld. Since no waist was used for the fatigue specimens the likelihood for failure at the fixture was increased.

The corresponding results for FSW T4 + PWAT are shown in Fig. 4. The extra low speed gives the best results but the difference between the low and the high speed is insignificant. The number of cycles to failure
Table 2
Mechanical properties of investigated materials

<table>
<thead>
<tr>
<th>Material</th>
<th>Yield strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation A5 (%)</th>
<th>References</th>
</tr>
</thead>
<tbody>
<tr>
<td>T6, base material</td>
<td>291</td>
<td>317</td>
<td>11.3</td>
<td>Present study</td>
</tr>
<tr>
<td>T6 (sheet and plate 0.35–10 mm)</td>
<td>260</td>
<td>310</td>
<td>10</td>
<td>Aluselvet (minimum values)</td>
</tr>
<tr>
<td>T4 (sheet 3–20 mm)</td>
<td>110</td>
<td>205</td>
<td>14</td>
<td>Present study</td>
</tr>
<tr>
<td>MRG-pulse</td>
<td>147</td>
<td>221</td>
<td>5.2</td>
<td>Present study</td>
</tr>
<tr>
<td>TIG</td>
<td>145</td>
<td>219</td>
<td>5.4</td>
<td>Present study</td>
</tr>
<tr>
<td>T6, FS welded low</td>
<td>150</td>
<td>245</td>
<td>5.7</td>
<td>Present study</td>
</tr>
<tr>
<td>T6, FS welded high</td>
<td>150</td>
<td>245</td>
<td>5.1</td>
<td>Present study</td>
</tr>
<tr>
<td>T4, FSW + aged low</td>
<td>255</td>
<td>310</td>
<td>3.6</td>
<td>Present study</td>
</tr>
<tr>
<td>T6, FS welded*</td>
<td>132</td>
<td>221</td>
<td>7.0</td>
<td>[8]</td>
</tr>
<tr>
<td>T4, FSW + aged*</td>
<td>260</td>
<td>289</td>
<td>7.3</td>
<td>[8]</td>
</tr>
</tbody>
</table>

* Six millimetre extrusions FS welded with a welding speed of 350 mm/min [8].

Fig. 2. Results of fatigue testing for MRG-pulse and TIG specimens. Comparison with FSW low- and high-speed tests. T6. Stress range.

Fig. 4. Influence of welding speed on results on fatigue life of FSW specimens in the temper T4 + PWAT.

Fig. 3. Influence of welding speed on fatigue life of FSW specimens in the temper T6.

Fig. 5. Influence of welding speed on results on fatigue life of FSW specimens in the temper T4 + PWAT. The fatigue life is given as mean stress versus number of cycle to fracture.

shows similar behaviour whether it is plotted versus stress range or mean stress, see Figs. 4 and 5.

The influence of temper on the fatigue life is illustrated in Figs. 6 and 7. For FSW T4 + PWAT, at 500,000 cycles to failure, the stress range is similar to that of T6, i.e. about 90 MPa both for the low and high welding speed specimens. For FSW T4 + PWAT the drawn mean logarithmic curves are almost identical. The low-speed tests show three to four specimens deviating from an
otherwise small scatter. The same applies to the high-speed tests. Scatter in fatigue data has shown to be generally small for FSW specimens [5,6,8]. This is due to the high repeatability of the process.

The comparison between the T6 and T4 + PWAT test series at high welding speed (Fig. 7) shows that T4 + PWAT has somewhat higher fatigue strength at high stresses while T6 is better at lower stresses (< 80 MPa stress range). Thus there is a steeper slope of the Wöhler curve for the statically stronger T4 + PWAT. It is possible that the higher ductility of T6 is advantageous at low stresses and long lives, since the main part of the fatigue life is related to initiation.

3.4. Fracture surfaces and microstructure

For the MIG-pulse and TIG specimens, fracture during fatigue loading propagated in the weld metal near the toe of the weld. Fracture surfaces after fatigue are transverse to the loading direction. Fatigue is often initiated at multiple positions in the root, see Fig. 8. Striations are observed in this area, see Fig. 9. In some instances initiation has been controlled by large inclusions. Fracture is mixed trans/intergranular. No signs of hot cracks were seen. Hot cracks can form in Al–Mg–Si fusion welds when the low melting Mg–Si eutectic in the grain boundaries remains liquid during weld solidification and subsequent material shrinkage. Both the MIG-pulse and TIG specimens show a dendritic cast microstructure in the weld with solute gradients near the dendrite boundaries. In FSW the weld nugget is composed of a superplasticised and recrystallised structure, which makes it very fine-grained see Fig. 10.

The high-speed T6 welds often fracture in the weld, either on the advancing or retreating side, no preference could be seen. In a few cases fracture is in the base material. T4 + PWAT, both low- and high-speed tests, show a similar pattern. In the same way as for MIG and TIG, the fatigue fracture pattern is mainly transgranular with intergranular parts. For many of the low-speed T6
FSW specimens fatigue initiated in the parent material or in the outer HAZ. Natural ageing has strengthened the weld and HAZ, since these test series were run a few weeks after production. When ductile fracture occurred, it propagated to the weld/HAZ border and followed it.

4. Modelling of softening

It has been proposed that the maximum welding speed is controlled by the softening of the material in front of the tool [9]. If no softening has taken place the forces on the tool and the surrounding material would be excessively high, easily giving rise to voids and other defects. This can reduce the fatigue strength. In this section a model for the softening around the weld is presented.

The generation of heat around the rotating tool is believed to be controlled by the friction between the shoulder and the work material [9,10]. The generation rate \( q \) can be expressed as [11]

\[
q = \frac{\pi}{3} \sigma \omega R_0^3
\]

where \( \sigma \) is the flow stress, \( \omega \) the angular frequency and \( R_0 \) is the radius of the tool shoulder. The softening of 6082 at different temperatures has been measured by Myhr [12] and Myhr and Grong [13]. They introduced a temperature compensated time \( t_{\text{comp}} \) to model the time dependence of the hardness

\[
t_{\text{comp}} = t_1 \exp \left( - \frac{Q_{\text{comp}}}{R} \left( \frac{1}{T} - \frac{1}{T_1} \right) \right)
\]

where \( Q_{\text{comp}} \) is an activation energy, \( R \) the gas constant, \( T \) the temperature and \( t_1 \) and \( T_1 \) are the reference values. By proper selection of \( Q_{\text{comp}} \) the hardness versus time curves at different temperatures can be brought together to a single master curve, if they are plotted versus an effective time \( t_{\text{eff}} \), see Fig. 11

\[
t_{\text{eff}} = \int_{t_{\text{comp}}}^{T} \left( \frac{dr}{t_{\text{comp}}(T)} \right)
\]

For a weld, the integration is performed over the thermal cycle of each element. The effective time has no dimension. The temperature field around the tool is assumed to be described by the modified Rosenthal equation in the so-called medium plate solution [14,15]. To take into account the limited amount of heat transport through the upper and lower plate surface mirror image sources are introduced

\[
T - T_0 = \frac{q}{2\pi l} \exp \left( \frac{\alpha l}{2d} \sum_{i} \frac{1}{R_i} \exp \left( \frac{\alpha R_i}{2d} \right) \right)
\]

Fig. 11. Softening curves for AA6082. Hardness versus effective time (Eq. (3)) at 250–375 °C.
where $\lambda$ is the heat conductivity, $a$ the thermal diffusivity, $v$ the welding speed, $x$ the coordinate in the welding direction and $R_1$ is the distance from the source ($i = 0$) or from one of the mirror sources ($i \neq 0$). The constant $T_o$ is taken as room temperature. The solution (4) is believed to describe the temperature field outside the shoulder in the HAZ in a reasonable way, but is more questionable closer to the tool [9,10].

The flow stress $\sigma$ is strongly temperature dependent, see Fig. 12 [16]. In addition the magnitude of $\sigma$ has a strong influence on the heat input and thereby also on the temperature distribution, cf. Eq. (1). The temperature below the shoulder versus the friction stress is also shown in Fig. 12. A balanced situation is obtained when the two curves meet. The stress at this point has been chosen for $\sigma$, Eq. (1). The maximum temperature below the shoulder and consequently also the flow stress are about the same for the extra low and the low speed, but the temperature is lower and the flow stress higher at the high speed.

Using the temperature distribution in Eq. (4) and the hardness versus the effective time curve, see Fig. 11, the softening of the alloy around the tool can be modelled. The result is given in Fig. 13. A cut off is introduced at the outer radius of the shoulder to take into account the fairly constant conditions below the shoulder. The model values do not give a fully precise description of the hardness profile which is to be expected for a model with no fitting parameters. The model however, clearly shows how the hardness decreases with increasing heat input per unit length in the sequence high, low and extra low speed. Some increase in the hardness takes place due to natural ageing after the welding in the partially solution treated area in the centre of the weld, in particular at the extra low welding speed. This effect is not included in the model.

To find the softening in front of the tool, the temperature cycle of these elements has been integrated and the effective time, Eq. (3), has been derived. For the extra low and low welding speeds a reduction in hardness of HV 51 and HV 54, respectively, is predicted corresponding more or less to full softening. However, at the higher welding speed the softening is only partial and the hardness reduction is HV 22 in front of the tool. Thus, complete softening is evidently not necessary to obtain a satisfactory welding procedure. It should also be pointed out that no defects as lack of fusion or pores were detected in the high-speed FS welds. At the same time the wide softening area of the tool at the extra low speed is likely to increase the metal flow and thereby avoid disturbances in the weld. This should contribute to the excellent fatigue properties at the extra low speed.

5. Discussion

The endurance limit for high-speed FS welds was at least equivalent to that of low-speed welds (Figs. 3 and 4). This has great practical consequences since the higher FS welding speed can be utilised without sacrificing the fatigue properties of AA6082. In a former investigation on the same material [3], specimens were FS welded with a transverse speed of the tool of 350 mm/min and a rotation speed of 1000 rpm. The tool and shoulder diameters were 10 and 20 mm, respectively. With a lower welding speed and a broader tool, the amount of heat supplied to the material is greater and therefore the metal flow and cooling of the weld might have been more effective. This could explain the relatively high fatigue strengths for the former series as compared to present data. The fatigue performance is relatively unaffected by the tensile strength of the material.
since a majority of the fatigue life of welded joints is spent in propagation of cracks to a critical size, and crack propagation rates are relatively insensitive to changes in alloy composition, heat treatment, or temper [17]. In these tests statically stronger T4 + PWAT material showed fatigue strengths at best equal to T6, which was unexpected. For S-N tests, the main part of the fatigue life is related to initiation. An increased ductility makes the material less sensitive to stress concentration and can then have a beneficial effect, which was seen for the extra-low speed and for T6 at high speed (Table 2; Figs. 3, 4, and 7). Furthermore, fractured specimens showed that fatigue propagation was generally more extensive in the HAZ than in the weld. High stress fracture often took place in the weld, whereas at low stresses and long lives it appeared in the HAZ or base material. Long fatigue lives can be associated with HAZ propagation, which was more common in T6. If a marked overlap weld toe is present, fatigue fracture can initiate and propagate on the advancing side of the FS weld [8], where it is most severe. This is particularly true if a softened area is also present (T6), introducing an area of maximum local stress and minimum local strength. In the present investigation this was not observed probably due to the absence of a marked weld toe.

A major problem at very fast welding speeds is to supply the material with enough heat energy to avoid a cold weld without sufficient bonding. Furthermore, relatively high rotation speed and thereby high strength of the spindle bearing are necessary. If the welding speed is increased so is typically the rotation speed of the tool, in order to provide enough heat per unit length along the weldline. A larger shoulder diameter of the tool will enhance the heat generated by increased friction area. At very large shoulders, however, local melting or excessive softening may result in difficulties keeping the plasticized volume under the shoulder. Instead, a smaller tool will be more efficient since it allows a greater increase in the spindle speed. Very high spindle speeds do not necessarily imply a problem, as the reaction force will decrease at high temperature [10].

The fatigue strength of the FS welds is clearly higher than that of the MIG-pulse and TIG welds. The TIG welds show higher fatigue strength than MIG, which, since the amounts of weld defects were equal, should in particular be attributed to its more narrow joint geometry. The height of the weld bed on the topside was about 1 mm for TIG welds as compared to 2–3 mm for MIG. As defects, both the MIG and TIG welds contained a limited amount of porosity (mainly occasional pores with a size less than 0.5 mm). Porosity occurs in aluminium alloy fusion welds mainly due to the rejection of hydrogen during weld solidification, and excessive porosity in the fusion zone reduces weld static strength and fatigue properties [17]. Some of the TIG welds contained large inclusions, which controlled fatigue initiation.

6. Conclusions

- The fatigue strength of FS welded Al–Mg–Si alloy 6082 is higher than that of MIG-pulse and TIG welds of the same material. The TIG welds show better fatigue performance than MIG, which is primarily accredited to the narrower geometry of the TIG joints.
- Mechanical and fatigue properties of the FS welds are relatively independent of welding speed in the range of low to high commercial welding speed in this alloy. An extra low speed gave improved properties.
- Welds in the artificially aged condition T6 showed at least equivalent fatigue properties in comparison to the slightly stronger naturally aged T4 plus PWA condition, which was unexpected. Contributing factors to this behaviour are the higher ductility in the T6 condition, which reduces the influence of stress concentrations and that crack propagation in the T6 condition mainly takes place in the HAZ (which was a comparatively slow process) whereas for T4, the crack propagation was in the weld.
- The softening behaviour around FS weldments was modeled and a fair representation of the measured hardness profiles as a function of welding speed was found.

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References


Fatigue Crack Propagation in Friction Stir Welded and Parent AA6082

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Abstract
The fatigue crack propagation characteristics of a friction stir welded Al-Mg-Si alloy, 6082, have been investigated. The electrical potential drop method was used for measurements. Two load ratio (R) levels were tested: a low and a high. At low load ratio (R=0.1) and a low stress intensity ΔK the propagation rate in the weld was higher than in the parent material by a factor 3-5. However, the propagation rates were approaching each other close to fracture. At high load ratio (R=0.8) the propagation rate was similar in parent material and weld. The weld crack growth rate was about the same at low and high R (except close to fracture), while the parent material growth rate increased at high R. Paris law was used to describe the measured crack propagation rates in the weld. In the case of the parent material, showing an R-dependence, Forman’s law was used.

Keywords: Friction stir welding, Al-Mg-Si alloy, fatigue crack propagation, load ratio, residual stress

1. Introduction
Previous examinations regarding the fatigue of friction stir welded (FSW) Al-Mg-Si alloys have shown that these welds have higher fatigue strengths than conventional fusion welds [1, 2]. Furthermore, fatigue initiation and propagation in FSW (butt welds) are not intimately associated with the weld nugget. Fatigue failure quite frequently occurs in the heat-affected zone (HAZ) or in the base material, indicating that the weld strength is close to parent material strength [2].

The major part of the service life of welded components subjected to fatigue is often related to crack propagation. To study the propagation rate of the friction stir weld, HAZ and parent material is therefore essential.
It has been suggested that the good fatigue performance in FSW is partially due to the comparatively low tensile residual stresses in the HAZ [3, 4]. This is due to the fact that welding takes place in the solid state with a comparatively low process temperature. The maximum temperature will reach about 500°C close to the weld centre in Al-Mg-Si alloys [5]. Another reason is the fine-grained weld nugget, where the grain size is finer than in the parent material. In addition to that, friction stir welds have small, favourable compressive residual stresses in the weld seam, both in the transverse and longitudinal direction [3, 4, 6]. The apparent positive influence of these can, however, be enhanced by the measuring method. This effect was evaluated by Dalle Donne et al. [4]. Measuring fatigue crack propagation (FCP) in FSW and base material, they showed that residual stress effects could be important, especially at low load ratios R. Then the relative magnitude of the residual stresses is larger. The small compressive stresses generated by FSW were found to lower fatigue crack propagation rate at low R, since they lowered the mean stress of the applied loading. As a result they concluded that residual stress effects should not be ignored, since they would give nonconservative estimates in FCP testing. By including the residual stress contribution in the applied load ratio this effect was eliminated, and propagation rate of the weld was seen to fall in the same scatter band as that of the parent material [4]. An earlier investigation, which does not take into account the effect of residual stress, reports crack growth rates that are somewhat lower in the weld and HAZ than in the base material. This is said to be due to the smaller grain size in these regions [1].

The aim of the present report is to evaluate the fatigue crack propagation of the friction stir welded Al-Mg-Si alloy 6082. In order to examine the effect of load ratio on propagation curves, two different R-values were used. These were 0.1 and 0.8.

2. Experimental

2.1 Material

The Al-Mg-Si alloy 6082 was friction stir welded in the naturally aged T4 temper condition, and thereafter post weld heat-treated. Mechanical data and composition are given in Tables 1, 2. Sapa performed the welding. A welding speed of 700 mm/min and a rotation speed of 2200 revolutions per minute were utilized.
Table 1. Mechanical properties (cross weld) of the material 6082.

<table>
<thead>
<tr>
<th>Material Condition</th>
<th>Thickness (mm)</th>
<th>Yield Strength (MPa)</th>
<th>Tensile Strength (MPa)</th>
<th>Elongation A5 (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>T4, FSW+PWAT weld speed: 700 mm/min</td>
<td>4</td>
<td>285</td>
<td>310</td>
<td>3.6</td>
</tr>
</tbody>
</table>

Table 2. Chemical composition of the 6082 alloy (wt%).

<table>
<thead>
<tr>
<th>Reference</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Mg</th>
</tr>
</thead>
<tbody>
<tr>
<td>Typical</td>
<td>0.7-1.3</td>
<td>≤0.50</td>
<td>≤0.10</td>
<td>0.4-1.0</td>
<td>0.6-1.2</td>
</tr>
<tr>
<td>Actual batch</td>
<td>0.96</td>
<td>0.19</td>
<td>&lt;0.01</td>
<td>0.45</td>
<td>0.59</td>
</tr>
<tr>
<td>Reference</td>
<td>Cr</td>
<td>Ni</td>
<td>Zn</td>
<td>Ti</td>
<td>Pb</td>
</tr>
<tr>
<td>Typical</td>
<td>≤0.25</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>≤0.10</td>
<td>&lt;0.01</td>
</tr>
<tr>
<td>Actual batch</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.01</td>
<td>&lt;0.01</td>
</tr>
</tbody>
</table>

2.2 Test procedure

Centre cracked (CCT) specimens were used in the crack propagation tests. The test specimens had a dimension of 230*70*4 mm (length, width, thickness). Electrical discharge machining was used to produce fully penetrated notches in the test specimens. These notches measured 10 mm in length and were symmetric about the specimen centreline, parallel to the welding direction.

The ASTM standard E647 regulating the procedure was followed. No precracking was performed but the notch edges were sharpened with a scalpel. Conditions of linear elastic fracture mechanics are required, affecting the material characteristics, specimen size, crack length, and applied load. For CCT-specimens these conditions can be expressed as

\[ W - 2a \geq 1.25P_{\text{max}}/B\sigma_{YS} \] (1)

Here \( W - 2a \) is the minimum uncracked ligament of the specimen width for linear elastic fracture mechanics to be valid, \( P_{\text{max}} \) is the maximum load, \( B \) the specimen thickness, and \( \sigma_{YS} \) the yield strength of the material. At the low load the allowable crack length (\( a=30 \) mm) was not exceeded. In the case of \( R=0.8 \), the uncracked ligament of the specimen width was required to be at least 42.5 mm, i.e. a crack
length \( a \) of max 13.75 mm was allowed. As is seen in Fig. 3, the final cracks in base material and HAZ slightly exceed this length. However, since the general trend is already manifested, it should be tolerable.

The Electrical Potential Drop method was used for measurements of the fatigue crack propagation rate. The principle is to apply an electric field through a cracked specimen. This field is a function of the specimen geometry and especially the crack length. If the current flow is constant, the electric potential drop across the crack plane will increase with increased crack length due to modification of the electrical field and associated disturbance of the current streamlines. The change in potential can be related to crack size through analytical or experimental calibration relations [7]. In this case an analytical expression was used to determine the crack length, eq. (2):

\[
a = \frac{W}{\pi} \cos^{-1} \left[ \frac{\cosh \left( \frac{\pi}{W} \right) \times Y_0}{\cosh \left( \frac{V}{V_r} \times \cosh^{-1} \left( \frac{\cosh \left( \frac{\pi}{W} \times Y_0 \right)}{\cos \left( \frac{\pi}{W} \times a_r \right)} \right) \right)} \right]
\]

where \( a \) is half the total crack length. \( V \) is the measured EPD voltage, while \( V_r \) is the measured voltage corresponding to a reference crack size \( a_r \), which was the original notch size. \( Y_0 \) is the distance from the lead for voltage measurement to the crack plane.

The propagation rate of the crack \( da/dN \) was plotted against the stress intensity factor \( \Delta K \). The propagation rate was determined by the secant method, that is

\[
\frac{da}{dN} = \frac{(a_{i+1} - a_i)}{(N_{i+1} - N_i)}
\]

where \( N_i \) is the number of cycles at the crack length \( a_i \). The stress intensity factor was calculated by the linear elastic handbook solution for a central crack in a finite plate with prescribed tensile stresses [8], i.e.

\[
\Delta K = \Delta \sigma \sqrt{\pi} \cdot a \cdot f_1 \left( \frac{a}{W}, \frac{h}{W} \right)
\]

and the geometry factor
\[ f_r\left(\frac{a}{W}, \frac{h}{W}\right) = \frac{1}{\cos\left(\frac{a}{W}\right)} \left(1 - 0.025 \cdot (2a/W)^2 + 0.06 \cdot (2a/W)^4\right) \]  

(5)

The initial value of \( \Delta K \), and thereby the applied stress range, was taken from a crack growth measurement on a 7000-series friction stir welded Al alloy in the literature [9]. It was chosen so that a reasonable growth rate of approximately \( 10^{-3} \) mm/cycle would be obtained. The value chosen for \( \Delta K_0 \) was 4 MPa\( \sqrt{\text{m}} \), giving a load range \( \Delta P \) of 8.8 kN. A \( \Delta K_0 \) of 3 MPa\( \sqrt{\text{m}} \) was also used for one specimen.

3. Results

3.1 Propagation data measurements

The results of the propagation tests, plotted as crack growth rate \( da/dN \) (mm/cycle) versus stress intensity factor \( \Delta K \) (MPa\( \sqrt{\text{m}} \)), are shown in Figs. 1 and 2 for \( R=0.1 \) and 0.8, respectively. The average stress was about 20 MPa at \( R=0.1 \) and 120 MPa at \( R=0.8 \). The potential drop voltage was registered with an interval of 0.5 min between readings. The number of data points for each series has been reduced by taking average values of \( da/dN \) and \( \Delta K \) over intervals of 10 to 50 points (depending on the

![Fig 1. Crack propagation rate at \( R=0.1 \) in weld, HAZ and base material of AA6082. \( \Delta K_0 \) was set to 4 MPa\( \sqrt{\text{m}} \).](image-url)
total number of cycles). The number of cycles to failure for each series is given in Table 3. In Fig. 3, half the total crack length \( a \) is plotted against the number of cycles for the different specimens. This plot clearly illustrates the distinction in fatigue life at \( R=0.1 \) and 0.8, and is useful as a supplement to Figs. 1 and 2.

Table 3. Results of the fatigue testing on pre-cracked specimens. \( \Delta K_o = 4 \text{ MPa m}^{1/2} \), \( \Delta P=8.8 \) kN.

<table>
<thead>
<tr>
<th>Crack growth area</th>
<th>R-value</th>
<th>Number of cycles to failure</th>
</tr>
</thead>
<tbody>
<tr>
<td>Weld 1</td>
<td>0.1</td>
<td>379 595</td>
</tr>
<tr>
<td>Weld 2</td>
<td>0.1</td>
<td>275 366</td>
</tr>
<tr>
<td>HAZ 1</td>
<td>0.1</td>
<td>1 030 105 *</td>
</tr>
<tr>
<td>HAZ 2</td>
<td>0.1</td>
<td>500 532</td>
</tr>
<tr>
<td>Base 1</td>
<td>0.1</td>
<td>1 124 147</td>
</tr>
<tr>
<td>Base 2</td>
<td>0.1</td>
<td>1 048 249</td>
</tr>
<tr>
<td>Weld 1</td>
<td>0.8</td>
<td>258 320</td>
</tr>
<tr>
<td>Weld 2</td>
<td>0.8</td>
<td>282 138</td>
</tr>
<tr>
<td>HAZ 1</td>
<td>0.8</td>
<td>321 115</td>
</tr>
<tr>
<td>Base 1</td>
<td>0.8</td>
<td>312 620</td>
</tr>
<tr>
<td>Base 2</td>
<td>0.8</td>
<td>295 889</td>
</tr>
</tbody>
</table>

*test ended after data recording accidentally interrupted (at 700 000 cycles)
As can be seen in Fig 1, at a low load ratio $R=0.1$, growth rate in the weld area is higher than in the parent material (measured about 25 mm from the tool shoulder line) with up to a factor of 3 to 5. The difference is reduced towards higher $\Delta K$ and at fracture the crack growth rate is about the same for weld and parent material. The length of the crack $2a$ at final fracture is about the same for the specimens, approximately 50 mm. The total number of cycles is clearly higher for the parent material than for the weld, see Table 3. Crack growth rate of the HAZ (measured about 1 mm from the tool shoulder line) is intermediate of the two and relatively low close to fracture. For the test designated HAZ 1, the EPD data collection was accidentally interrupted at $7\times10^5$ cycles. The mechanical test was allowed to continue, however, until the loading was interrupted above 1 million cycles. It is interesting to note that the HAZ crack propagation rate here is similar to that in parent material, see Fig. 3.
At high load ratio $R=0.8$, the difference in crack propagation rate between weld, HAZ, and parent material is much lower, and so is the difference in number of cycles to failure, see Fig. 2 and Table 3.

The weld crack growth rate remained rather unchanged between low and high load ratio, while the parent material growth rate increased to a level equal to that of the

![Graph a)](image)

**Fig. 4.** Crack propagation rate in the weld (a) and base material (b) at $R=0.1$ and 0.8.
weld at high R, see Figs. 4a and b. In the next segment it is discussed how to correlate for the R-dependence of the crack growth rate in the parent material.

Propagation data from the literature for alloy 6061 [10] is plotted together with 6082 in Fig 5. Parent 6061 nearly matches 6082 in the tested interval. When comparing the FSW plots with results in the investigation of Dalle Donne et al. (6013 T6, using their data obtained when the residual stress contribution was included in the applied load ratio) [4], they agree quite well, Figs. 6, 7. The difference is that the base material in the present investigation fails at a lower stress intensity ratio ΔK.

![Base material](image)

**Fig. 5.** Crack propagation rate in the base material at R=0.1 and 0.8. Literature data for 6061 [10] is given for comparison.

### 3.2 Analysis of data

The crack propagation curves were described with either Paris law, or Forman’s law in the case of an R-dependence. Regression analysis was used to obtain the best data fit curves. For crack propagation in the weld the data points followed a straight line according to Paris law:

\[
\frac{da}{dN} = CΔK^n
\]  

(6)
Fig. 6. Crack growth rate in 6082 and 6013 [4] FSW at R=0.1. Curve approx. by Paris law. Regression analysis for best data fit: \( C=1.09 \cdot 10^{-7}, n=3.50 \) (6082); \( C=1.94 \cdot 10^{-7}, n=3.47 \) (6013).

Fig. 7. R=0.8. Curve approx. by Paris law. Regression analysis was made for best data fit; \( C=3.29 \cdot 10^{-8}, n=4.35 \) (6082); \( C=5.54 \cdot 10^{-8}, n=3.70 \) (6013).

At R=0.1, the constants were evaluated to \( C=1.09 \cdot 10^{-7} \) [m/(MPa·m\(^{1/2}\))] and \( n=3.50 \) (Fig. 6). At R=0.8 the slope of the propagation curve was steeper and \( C=3.29 \cdot 10^{-8} \) and
was used to give a fair fit (Fig. 7). In this case, however, there was a greater uncertainty in determining the slope of the curve. The final data points before fracture diverge from a generally straight line, which is typical for the last regime of crack growth. However, since the data originates from samples with different $\Delta K$ start values ($\Delta K_0=3$ and 4 respectively), and due to the fact that the beginning of “steady state” propagation is difficult to determine, also the initial parts of the subsequent curves may be regarded as offset.

No attempt was made to evaluate the threshold $\Delta K$ value. However, since the data points follow the generally straight line just above $\Delta K=3$, and hence are in the Paris regime, the threshold for crack propagation is below that value.

Since the 6082 parent material exhibited a significant $R$ dependence of $da/dN$, Forman’s law, incorporating the load ratio in the formula, was used. It takes the form:

$$\frac{da}{dN} = \frac{C_f(\Delta K)^n}{(1-R)K_f - \Delta K}$$

where $C_f$ and $n$ are Paris-type material constants. $K_f$ is the fracture toughness of the

---

**Fig. 8.** Crack growth rates in 6082 base material at $R=0.1$ and 0.8. Curve approx. by Forman’s law: $da/dN = C_f(\Delta K)^n / (1-R)K_f - \Delta K$. Regression analysis for best data fit curves; $C_f=2.8 \times 10^4$, $n=3$. 

material. It was not determined for 6082, but was estimated to about 55 from the $K_t$-value for 6061, T6 sheet, which is given in the literature as 60.1 [11]. Forman’s law provided a reasonable data fit when the constants were set to $C_i = 2.8 \times 10^{-5}$ and $n=3$ (Fig. 8). The curve at $R=0.8$ slightly illustrates the sigmoidal shape, pertaining to escalation of crack growth in the final failure regime.

4. Discussion
The supposed lowering of the weld growth rate (below parent material growth rate) at low $R$, due to residual stress effects in the weld seam was not observed during this investigation (see the introduction, ref [4]; small compressive stresses generated by the FSW tool shoulder were found to lower crack propagation rate at low $R$, since they lowered the mean stress of the applied loading). Considering a relatively small test specimen, the effect of residual stress is thought to be limited. However, magnitudes of longitudinal tensile stresses as high as 20-50 % of parent material yield strength have been measured in small FSW samples. A probable explanation for this was given, namely that the effect of rigid clamping arrangements impeded contraction of the weld nugget and HAZ during cooling [6]. It should be noted, however, that the amount of residual stress is lower in the centre-cracked specimen than in the compact tension (CT) specimen. In addition, the magnitude of the stress is low perpendicular to the welding direction, being the stress state which affects a crack growing parallel to the weld.

In general, a higher $R$-value gives a higher propagation rate, which is seen here for the parent material. Many semi-empirical and empirical models have been proposed to account for the load ratio dependence of $d\alpha/dN$. For example, the Walker equation yields a series of parallel lines on a log-log plot whose separation is a function of the stress ratio [12]. The approach by Forman, Kearney and Eagle (eq. 7) additionally considers the deviation of crack growth behavior in the near-threshold and final failure regimes from that described by the Paris law [13]. To correlate and predict the crack propagation rate for different $R$-ratios, which is important for design engineers, several concepts have been adopted. The mechanism perhaps most commonly referred to is crack closure; during the unloading portion of a load cycle there is premature contact between the crack surfaces while some tensile load is still applied. Elber has proposed that only the load range between the opening load and the maximum load affects the crack tip action [14]. That is, for low load ratios, the
effective $\Delta K$ is lower than the applied value (at a high $R$ the crack closure effect is more or less non-existent). By using e.g. the ASTM 2% opening load method, crack propagation curves at different $R$-values are seen to coincide for some Al alloys (the method underestimates $\Delta K_{eff}$ at near threshold growth rates) [15]. The fact that crack closure phenomena have not been taken into account here, may to some extent explain the difference in growth rate for parent material at $R=0.1$ and 0.8. The long fatigue lives of specimens with cracks initiated in the HAZ (compare table 3) are likely to depend on roughness induced crack closure. It has been suggested as an explanation for low fatigue crack growth rates in the HAZ of FS welded aluminium alloy 7075 [9]. As shown in Fig. 9 the fracture surface morphology in the HAZ exhibits more tortuous paths than that in the weld. The rougher structure will complicate advancement of the crack.

![Fracture surfaces in 6082](image1)

*Fig. 10. Fracture surfaces in 6082 a) weld and b) HAZ. c) Close up of fracture surface in the border area between weld and HAZ showing rough morphology.*
There has been some debate regarding the validity of crack closure data, since the crack-opening load depends on the measurement location and the technique employed. Furthermore it has been shown that a significant contribution to fatigue damage can occur below the opening load. Therefore some new approaches to R-ratio effects on fatigue crack growth have been introduced. The so called ACR and CWI methods measure the change in displacement at minimum load due to closure, a quantity that is less subject to variability [15]. Another method simply calculates the driving force parameter as a geometric mean of the positive part of the applied stress intensity factor range $\Delta K^*$ and the corresponding maximum value of $K$ [12]:

$$\Delta K^* = (\Delta K^* \cdot K_{max})^{0.5}$$  \hspace{1cm} (8)

where $K_{max}$ is $\Delta K/(1-R)$. The parameter $\Delta K^*$ yielded a fairly good correlation of the load ratio effect for a number of Al alloys, among them 6013-T651 [16]. As opposed to the opening load methods, here the crack growth curve for the low R (0.1) becomes more or less a fixed curve against which the curves for higher load ratios are correlated. In this the crack closure data are not used. Compared to Elber’s opening load approach the $\Delta K^*$-method gave an improved correlation in the near threshold region [16]. However, the fatigue crack growth data for 6082, R=0.8 is largely overcompensated by eq. (8) and the formula is not applicable in the present case.

Conclusions

- The crack propagation rate in the friction sti: weld, 6082, matches that of the parent material at high load ratio $R$.
- The propagation rate in the weld is about the same at a low and a high R-value, while that in the parent material is increased at high R. This may to some extent be attributed to the phenomenon of crack closure, which is known to occur in some Al alloys at low load ratio.
- The propagation rate in the HAZ is intermediate between weld and parent material, and is the lowest one at high $\Delta K$. The long fatigue lives of specimens with cracks initiated in the HAZ are likely to depend on roughness induced crack closure.
• Paris law was used to approximate the measured crack propagation rates in the friction stir welds. Forman’s law, incorporating the R-dependence, was used in case of the parent material for which it gave a reasonable fit.

Acknowledgements

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Paper IV
Fatigue properties of friction stir overlap welds

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Abstract
For many applications involving lap or T-joints friction stir welding (FSW) is presently used, e.g. hermetically closed boxes such as cooling elements and heat exchangers. The frequent pressure changes in these make them susceptible to fatigue. The fatigue characterization of lap joints involves a combination of shear and bending. Forces applied to the ends of lap joints result in non-axial stresses in the connection area.

FSW lap joints of the Al-Mg-Si alloy 6082 in the artificially aged condition T6 were studied. A pin (probe) based on the Triflute™ concept was used. Two modifications were made to the pin, the pin end being either convex or concave. Tool shoulders of 15 and 18 mm respectively were utilized, producing four different weld series. Fracture was initiated in the highly stressed area where the weld cuts through the interface between the two sheets. The cracks typically propagated through the weld in the upper sheet (tool side). The broadest tool shoulder with a concave end of pin design gave the best fatigue performance. This was due to an improved flow path provided by the hollowed out end of pin; the material flow around the pin resulted in minimal “hooking” of the sheet interface adjacent to the weld nugget. In addition an increased amount of heat energy was supplied by the enhanced contact area tool work piece. In all welds the amount of porosity was low. The stress distribution in the test specimen was modelled. The stress intensity factor ΔK was determined. It was found that a simplified approach, developed to estimate ΔK for overlap spot welds, could be used also for friction stir overlap joints. The corresponding crack propagation rates were in fair accordance with the experimental results.

Keywords: Friction Stir Welding, overlap joints, tool design, fatigue, mechanical properties, stress intensity factor
1. Introduction

Solid-state friction stir welding (FSW) is by now well established for aluminium butt joints. The technique is also used in many applications involving lap joints, or joint-types which can be described as a combination of lap and butt joints. For example, hermetically closed boxes such as cooling elements and heat changers are friction stir welded. When welding aluminium in particular, the ability to join at a low temperature is critical, since the material is very sensitive to thermal distortion. In the car industry there is potentially an increased use of FSW. Current applications include engines, wheel rims and lap joints in car back supports, see Fig. 1. Traditional resistance spot welding implies the same problem with distortion as other fusion welding methods, and the scope of FSW would be larger if body members were to be made of aluminium. High-strength aluminium structures in airplanes (often of lap- or t-joint type), traditionally viewed upon as unweldable and fastened by rivets, can be friction stir welded. This will reduce production times and manufacturing costs [1,2]. On the basis of static strength, the FSW lap joint has been shown to be fully comparable with resistance spot welding and riveting. The shear tensile strength of FSW joints was found to be 2.4 times that of single row riveted joints [2,3].

Fig. 1. FSW overlap joint example, back support in a car seat.
Fatigue characterization of lap joints is not as easily carried out as for butt joints, since there is a combination of shear and bending involved. Forces applied to the ends of lap joints result in eccentric loads in the connection area. This can cause joint rotation. To avoid this, parts have to be sufficiently restrained. If there is no external restraint, two rows of welds can be used (in the case of fillet welds) [4]. For FSW lap joints, the width of the weld in the interface between the sheets is significant. A wider weld nugget causes less bending.

The rotation of the tool results in material flow around the pin. When rotating clockwise (if left-threaded pin) the metal is moved upwards towards the tool shoulder [5]. The interface between the sheets is pulled up adjacent to the nugget. The shape of these interface line upturns or tears is of fundamental importance, especially for applications that are subject to fatigue. One result is that the “effective sheet thickness” (EST) on the retreating side of the weld is reduced, making it more prone to fracture. Even more important is the notch effect introduced by the shape of the interface line on the advancing side of the weld, see Fig. 2.

![Fig. 2. Etched specimen cross-section, tool 1 (T-joint, same tools used as in the present investigation). Influence of vertical material flow on the interface between the sheets. Both on advancing and retreating side the interface is pulled up adjacent to the weld nugget, as a result of material being forced into the top sheet from below. On the advancing side the interface forms a sharp notch with the nugget.](image)

In each sheet, the side which experiences maximum tensile stress can be defined as “loaded”, Fig. 3. Cederqvist et al. have shown that minimum tensile strength is attained when the retreating side is loaded in the top, weld sheet (in a mixed 2000-
Fig. 3. Theoretical stress distribution in an ideal shear test specimen. The tensile stress varies in the top and bottom sheet according to the figure. The shear stress is approximately constant along the interface weld width. Observed fracture paths during the mechanical testing are marked. Advancing weld side is to the left.

7000 series FSW lap joint) [3]. Fracture was then mainly through the bottom sheet, advancing side, since an interface pull down was created there that was more critical than the pull up on the retreating side (compare Fig. 2). The pull down reduced the effective sheet thickness of the loaded side in the bottom sheet. For this reason the authors recommend the weld direction to be chosen so that in the top sheet the advancing weld side becomes “loaded” (mainly for single pass welds, in the case of double pass welding the critical side can be eliminated by switching over advancing and retreating sides in the subsequent weld run).

Design of the tool pin (or probe), which is the lower part of the tool penetrating the material, is an important feature for the weld quality. It has been shown that a conventional, conical pin will produce deficient lap joints due to incomplete pressure at the sides of the weld [6]. If not thoroughly dissipated, streaks of surface oxide can break into the material by way of metal convection in the stirred zone, lowering the mechanical properties. To improve the flow path around and underneath the probe reentrant features can be machined into its core, so called flutes, thereby increasing the difference between the swept material volume and the static volume of the probe.

The tool shoulder diameter also affects the final weld result. A larger shoulder diameter will enhance the heat generated by increased friction area. At very large shoulders, however, local melting or excessive softening may result in difficulties keeping the plasticized volume under the shoulder.
2. Experimental

2.1 Material and welding

Sheets of AlMgSi alloy 6082, 500 mm long and 80 mm wide, were lap welded by friction stir welding. They were thereafter cut into test specimens. Sapa performed the welding. The feeding speed of the material was 300 mm/min, the rotating speed of the tool 1500 rpm. The material had the temper condition T6. Material composition is given in Table 1. The effective length of the welded specimens was 130 mm. The width was 40 mm, and the material thickness 4 mm. The overlap length was 30 mm (As a general guideline the overlap should be at least five times the thickness of the thinner part (min. 25 mm) [4]).

Table 1. Chemical composition of the 6082 alloy (wt%).

<table>
<thead>
<tr>
<th>Reference</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Mg</th>
</tr>
</thead>
<tbody>
<tr>
<td>Typical</td>
<td>0.7-1.3</td>
<td>≤0.50</td>
<td>≤0.10</td>
<td>0.4-1.0</td>
<td>0.6-1.2</td>
</tr>
<tr>
<td>Actual batch</td>
<td>1.0</td>
<td>0.20</td>
<td>&lt;0.02</td>
<td>0.48</td>
<td>0.65</td>
</tr>
<tr>
<td>Reference</td>
<td>Cr</td>
<td>Ni</td>
<td>Zn</td>
<td>Ti</td>
<td>Pb</td>
</tr>
<tr>
<td>Typical</td>
<td>≤0.25</td>
<td>≤0.20</td>
<td>≤0.10</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Actual batch</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.01</td>
<td>0.02</td>
<td>&lt;0.01</td>
</tr>
</tbody>
</table>

A preliminary series was tested. Thereafter, four test series were produced. Pins based on the Triflute<sup>TM</sup> concept were used in all cases. Three flutes were machined into the cylindrical, threaded core. The length of the unmodified pin was 5.5 mm. Two slight modifications were tested; the end being either concave (hollowed out) or convex (protruding). In the latter case the max length of the pin was increased with 1 mm, giving an effective bottom sheet penetration of approximately 2 mm (50%). The probe diameter was 6 mm. Tool shoulders of 15 and 18 mm were utilized, producing four different tools (see Fig. 4). In order to obtain a tensile/shear test, edge pieces corresponding to the material thickness were glued to the specimen ends (Fig. 3). It was not possible, however, to fully eliminate a degree of bending moment. At a static load corresponding to 30 MPa, the gap width between the overlap sheets was determined to be approx. 70 µm. At 60 MPa a slit of 200 µm was measured. The deformation attained by the static loading was plastic. After fatigue the gap width was measured to ~35 µm. This value was used in modelling the slit in the test specimen as well as for the diameter of the notch.
2.2 Test procedure

The static and dynamic strengths of the welds were determined. Testing was carried out in a servo-hydraulic machine equipped with an actuator of 250 kN load capacity. The static tensile testing was performed in load control at a rate of 10 kN/min. In fatigue a sinusoidal load-time function was used, with the stress ratio R set to 0.1. Here maximum net section stresses from 25 up to 60 MPa were tested (22.5 to 54 MPa stress range). The oscillation frequency was set to 12 Hz. The net section stress was defined as the force divided by the material thickness (4 mm) times the specimen width (40 mm).

2.3 Macro and micro studies

Macro studies on weld cross sections were performed to evaluate weld formation, interface tears, and fracture paths. Back scattered SEM was used to analyse the presence of oxides in conjunction with the interface and weld.
3. Results

3.1 Mechanical testing

The designation of the tools is given in Fig 4. The tensile shear strength of the welds was 160 MPa. It was determined as the mean of eight tests with all tools. In general fracture propagated through the weld in the upper sheet, advancing side (which coincided with the side of the top sheet experiencing maximum tensile load, see Fig. 3). For one specimen, tool 2, fracture was through the bottom sheet (retreating side). Only in one instance fracture was by shear through the interface. This specimen had a lower strength than the average of about 130 MPa. The joint efficiency in relation to static strength of T6 base material of the same dimensions was about 55%. FSW butt joint efficiency for 6082 has previously been determined to 80% [8]. A few literature references are available, giving lap joint efficiencies in the range of around 40% for mixed joints (2024-7075, single pass welds, efficiency relative to 2024 base metal) up to ~65-85% for AlMg0.5Si1 T4 (at the instance of fracture occurring in the weld or HAZ, otherwise more than 98%) [3,5]. Although process parameters and material thickness combinations have differed between the investigations, the general trend of lower joint efficiencies for lap welds compared to butt welds is manifested (double pass welding has been shown to increase joint efficiencies of lap welds substantially. Reasons for this are the increased interface weld width, enhanced oxide disruption, and the possibility to eliminate the critical advancing weld side.). Results of the fatigue testing are plotted in Figs. 5 and 6. Individual specimen data along with information on fracture location is given in Table 3. The fatigue data for the specimens welded with the smaller tool shoulder: concave and convex end of pin, nearly matches (tools 1 and 3). However, the convex pin is not as beneficial as the concave pin at an increased shoulder diameter (tools 4 and 2, respectively). The tool with broader shoulder and concave end of pin, tool 2, produced the welds with best fatigue resistance. Friction stir welded overlap joints have not yet been extensively investigated with regard to fatigue. The results are in agreement with one set of previously published data though [7]. This reference dealt with FS overlap welded sheet EN AW-5754-H14 (AlMg1) of 3 mm thickness (as well as mixed joints which had lower strength). The tool shoulder diameter was 15 mm, and the penetration depth of the pin through the bottom plate 1.8 mm (60%). Static joint efficiency was ~65-70 %.
Fig. 5. Fatigue of FSW overlap joints, AA6082 T6, welded with concave end of pin. Compared with literature data [7].

Fig. 6. Fatigue of FSW overlap joints, AA6082 T6, welded with convex end of pin.
Table 3. Data for fatigue tested specimens (preliminary series, tool 1, not included)

<table>
<thead>
<tr>
<th>Tool</th>
<th>Shoulder diam. (mm)</th>
<th>Max. stress (MPa)</th>
<th>Number of cycles</th>
<th>Fracture: Sheer, Distance from outer tool shoulder mark</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>15</td>
<td>60</td>
<td>21452</td>
<td>top sheet, 4-5 mm</td>
</tr>
<tr>
<td>1</td>
<td>15</td>
<td>55</td>
<td>22887</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>15</td>
<td>50</td>
<td>35673, 9309, 28882, 11012</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>15</td>
<td>45</td>
<td>10351, 66601, 71456, 40710</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>15</td>
<td>40</td>
<td>16307, 22433, 83511, 66141, 21956, 455544</td>
<td></td>
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<td>15</td>
<td>35</td>
<td>47640, 26496</td>
<td></td>
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<tr>
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<td>78559, 66485, 87436, 109911</td>
<td></td>
</tr>
<tr>
<td>1</td>
<td>15</td>
<td>25</td>
<td>107288, 165863</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>18</td>
<td>60</td>
<td>23480</td>
<td>shear through interface</td>
</tr>
<tr>
<td>2</td>
<td>18</td>
<td>55</td>
<td>31086, 21851</td>
<td>top, 5-6,5 mm</td>
</tr>
<tr>
<td>2</td>
<td>18</td>
<td>50</td>
<td>30512, 222695, 36380</td>
<td></td>
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<tr>
<td>2</td>
<td>18</td>
<td>45</td>
<td>85037</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>18</td>
<td>40</td>
<td>74363, 110981, 98161, 531464</td>
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<td>18</td>
<td>40</td>
<td>153742, 323899</td>
<td>bottom, 4-5 mm</td>
</tr>
<tr>
<td>2</td>
<td>18</td>
<td>35,6</td>
<td>1830000</td>
<td>test interrupted</td>
</tr>
<tr>
<td>2</td>
<td>18</td>
<td>30</td>
<td>528529, 1436796</td>
<td>bottom, 4-5 mm</td>
</tr>
<tr>
<td>2</td>
<td>18</td>
<td>30</td>
<td>2150000</td>
<td>test interrupted</td>
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<tr>
<td>3</td>
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<td>55</td>
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<td>4</td>
<td>18</td>
<td>55</td>
<td>8662</td>
<td>top, 5-6 mm</td>
</tr>
<tr>
<td>4</td>
<td>18</td>
<td>40</td>
<td>39311, 32186</td>
<td></td>
</tr>
<tr>
<td>4</td>
<td>18</td>
<td>35</td>
<td>91344</td>
<td></td>
</tr>
<tr>
<td>4</td>
<td>18</td>
<td>30</td>
<td>63647, 311750, 174599, 195884</td>
<td></td>
</tr>
</tbody>
</table>

The results were also compared with FSW butt joint data [8], Fig. 7. The new data corresponds to not fully penetrated butt joints. In these a notch of about 0.5 mm remained in the root surface (in addition to that the new specimens were thinner and had a waist with 80 mm² cross section area, as compared to 280 mm² without a waist for the previous). As expected, the dynamic strength of the overlap welds, in relation to butt welds, was found to be small. The eccentric load and severe notch situation controls that. The max fatigue stress at 100 000 cycles was about 20-30 % of the static overlap strength depending on tool selection (similar results were reported for
the 5754-alloy referred to above). A corresponding value for butt joints is about 90%.

Fig. 7 shows that the strongest overlaps (tool 2) has a fatigue strength up to 
$\sim2\times10^6$ cycles of about 70-80% of that of the partially penetrated butt welds ($R=0.1$).
At $2\times10^6$ cycles the same overlaps has a fatigue strength of only 38% of the fully penetrated butt welds ($R=0.5$). In the graph the Eurocode 9 design curve for fusion fillet welds has been added. Cracking is initiated at the weld and propagated through the base material perpendicular to the load direction. The same detail category pertains to partially penetrated fusion butt welds (with cracking originating in the root) [9]. It is seen that the mean curves for partially penetrated FSW butt joints and FSW overlaps exceeds the design curve towards longer fatigue lives.

3.2 Macroscopic observations

By macro etching of fractured specimen cross sections, the fatigue crack path in relation to the weld was revealed. A caustic etch was used. The weld penetration was smaller for the concave than for the convex end of pin. For the former it was in the range of approximately 0.8 to 1.5 mm or 20-38% penetration through the bottom
Fig. 8. Macro etched over lap cross sections. It can be seen that the concave end of pin (left) gives about 1 mm or 25% weld penetration through the bottom sheet as compared to 50% with the convex pin (right).

sheet, as compared to ~2 mm or 50% for convex pin, Fig. 8. The penetration was not as constant for concave as for convex pin which may have given some variations in mechanical data.

Fracture is initiated in the highly stressed area where the weld cuts through the interface between the sheets, the stress state being aggravated by the notch. This was seen for all specimens. The transverse width of the weld nugget at the interface was determined. This interface weld width was ~6-7 mm, i.e. about the diameter of the pin. As expected, fracture coincided with the “loaded” side of the weld (top sheet). This side has the highest structural stress (Fig. 3). In welding it has been the advancing side of the tool, which is most prone to failure in monotonic shear testing (due to the interface notches, see Fig. 2 and ref. [3]).

Fig. 9. Fracture patterns. (a) Advancing side failure through weld in top sheet. (b) Retreating side failure by way of base material in bottom sheet (rare).

In the majority of cases fatigue cracks propagated nearly vertically through the weld nugget/TMAZ in the upper, fully penetrated plate (path A, Figs. 3, 9a). In four cases
only, tool 2 tests at low load, failure was through the bottom plate (path B, Figs. 3, 9a). In these instances fracture is on the retreating side of the weld. Cracking initiated at the interface and propagated along the edge of the weld nugget in the bottom plate, continuing down through the HAZ and unaffected base material. These tests managed comparatively high number of cycles before final fracture.

In Fig. 10 the effect of tool design on the sheet interface movement is seen. It is obvious that tool 2 (18 mm shoulder, concave end of pin) has given very little influence either on the advancing or retreating side of the weld. The bottom sheet penetration is small (~1 mm) compared with specimens welded with convex pins.

![Fig. 10. Effect of pin design on the sheet interface movement. Concave pin, 18 mm shoulder (tool 2); interface line adjacent to the weld nugget on a) retreating side, b) advancing side. Convex pin, 18 mm shoulder (tool 4); c) retreating side, d) advancing side.](image-url)
3.3 Numerical simulation of deformation in FSW overlap joints

3.3.1 Stress intensity factor in the symmetric case

To determine the stress distribution in the overlap test specimen a finite element model was used. The geometric feature which affects the stress in the model is foremost the notch radius between the overlap sheets. A larger sheet interface width and notch radius lower the maximum stress. For the ideal case, when there is no pull up of the interface (compare Fig. 10a-b), the interface line was drawn straight to the nugget (case 1).

Fig. 11 plots the numerically calculated stress in the specimen at an applied stress $\sigma_{\text{max}}$ of 40 MPa during dynamic loading. The maximum stress is around the notch tip at the interception between top and bottom sheet and the weld. This mesh area constitutes a stress singularity in the model, and shall not be used for calculations. In the symmetric case 1, the stress distribution is the same on the advancing and retreating weld side.

On the advancing side the stress peak lies in the weld, while on the retreating side it is in the bottom sheet. The variation within the weld of the stress in the x-direction $\sigma_x$, y-direction $\sigma_y$, and the shear stress $\tau$ can be expressed as:

\[
\sigma_x = \frac{1}{\sqrt{2\pi r}} \left[ K_t \cos \left( \frac{\theta}{2} \left( 1 - \sin \frac{\theta}{2} \sin \frac{3\theta}{2} \right) \right) - K_i \left( \frac{\rho}{2r} \right) \cos \frac{3\theta}{2} \right] \left( 1 \right)
\]

\[
\sigma_y = \frac{1}{\sqrt{2\pi r}} \left[ K_t \cos \left( \frac{\theta}{2} \left( 1 + \sin \frac{\theta}{2} \sin \frac{3\theta}{2} \right) \right) + K_i \left( \frac{\rho}{2r} \right) \cos \frac{3\theta}{2} \right] \left( 2 \right)
\]

\[
\tau = \frac{1}{\sqrt{2\pi r}} \left[ K_{\tau} \sin \left( \frac{\theta}{2} \left( 1 - \sin \frac{\theta}{2} \sin \frac{3\theta}{2} \right) \right) - K_{\tau} \left( \frac{\rho}{2r} \right) \sin \frac{3\theta}{2} \right] \left( 3 \right)
\]

where $\theta$ is the angle of deviation from the slit centreline, and $\rho$ is the notch tip radius [10]. $K_t$ is the stress intensity factor corresponding to a crack opening load perpendicular to the plane of the crack. $K_{\tau}$ pertains to the shear mode of crack opening. The resulting stress intensity factor $\Delta K$ can then be evaluated for $\theta=0^\circ$ as
\[ K_1 = \frac{\sigma_x \cdot \sqrt{2\pi \cdot r}}{(1 - \frac{\rho}{2r})} = \frac{\sigma_y \cdot \sqrt{2\pi \cdot r}}{(1 + \frac{\rho}{2r})} \quad \text{and} \quad K_\perp = \frac{\sigma \cdot \sqrt{2\pi \cdot r}}{(1 - \frac{\rho}{2r})} \]

\[ K_{eq} = \sqrt{K_1^2 + K_\perp^2} \quad (4) \]

\[ \Delta K = (1 - R)K_{eq} \quad (5) \]

The stress in the x-direction reaches 164 MPa at a maximum in the weld (Fig. 11a, adv. side shown). This is 25-30 µm from the notch tip. In the y-direction the stress is about 225 MPa at a maximum in the weld, and close to zero along the surface of the slit (Fig. 11b). \( \Delta K \) is plotted as a function of the distance from the notch tip, Fig. 12. It is seen that at a distance exceeding the notch tip radius 17.5 µm, \( \Delta K \) becomes constant at about 3.2 MPa/√m. This result was obtained by combining eqs. (2) and (3).

Fig. 11. Stresses in the overlap test specimen around the interface ends adjacent to the weld nugget. Case 1: Stresses in a) x- and b) y-direction on the advancing side at a dynamic loading \( \sigma_{max} \) of 40 MPa.
Fig. 12. The variation of the stress intensity factor $\Delta K$ (within the weld nugget) as a function of the distance from the notch tip according to eqs. (1-5). The notch tip radius is 17.5 μm.

Radaj and Zhang [e.g. 11] showed that if the structural stresses, i.e. $\sigma_{sh}$, $\sigma_{ho}$, $\sigma_{lt}$, $\sigma_{lo}$, $\tau_{qy}$, and $\tau_{ql}$ around a spot weld are available, the stress intensity factors can be determined according to:
where $\sigma$ refers to stress in the x-direction and $\tau$ to stress in the y-direction. The prefixes denote: $u =$ upper sheet, $l =$ lower sheet, $i =$ inner side of sheet and $o =$ outer side of sheet (Fig. 13). The equations were used to estimate $\Delta K$ along the notch where the crack starts to propagate. The structural stresses were determined from the FE-model. The stresses should not contain the singularity caused by the slit tip between the welded sheets. 1-2 mm from the tip, the stress in the x-direction takes on quite a constant value around 160 MPa, measured on the surface of the slit. The corresponding $\Delta K$ lies around 3 MPa/m, see Table 4.

![Fig. 13. Structural stresses around a spot weld [11.]](image)

Table 4. Structural stresses and stress intensity factors in the overlap test specimen; $\sigma$ refers to stress in the x-direction and $\tau$ to stress in the y-direction. Stresses at an applied fatigue load of 6.4 kN (40 MPa) for case 1. The stresses were determined at the upper ($\sigma_{u\tau}$, $\tau_{u\tau}$) and lower ($\sigma_{o\tau}$, $\tau_{o\tau}$) surfaces of the sheet interface at a distance $r$ from the notch tip, and on the top ($\sigma_{on}$) and bottom ($\sigma_{on}$) sheet surfaces. $K_{I}$ and $K_{II}$ were determined by eqs. (6-7).

<table>
<thead>
<tr>
<th>$r$ [mm]</th>
<th>$\sigma_{u\tau}$ [MPa]</th>
<th>$\sigma_{o\tau}$ [MPa]</th>
<th>$\sigma_{u\tau}$ [MPa]</th>
<th>$\sigma_{o\tau}$ [MPa]</th>
<th>$\tau_{u\tau}$ [MPa]</th>
<th>$\tau_{o\tau}$ [MPa]</th>
<th>$K_{I}$ [MPa/m]</th>
<th>$K_{II}$ [MPa/m]</th>
<th>$\Delta K$ [MPa/m]</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.5</td>
<td>9.0</td>
<td>-66</td>
<td>-360</td>
<td>-9.6</td>
<td>9.0</td>
<td>-5.4</td>
<td>6.8</td>
<td>20.1</td>
<td>19.1</td>
</tr>
<tr>
<td>0.25</td>
<td>207</td>
<td>-69.3</td>
<td>-53.8</td>
<td>-8.62</td>
<td>-0.058</td>
<td>0.023</td>
<td>2.1</td>
<td>4.1</td>
<td>4.2</td>
</tr>
<tr>
<td>0.5</td>
<td>179</td>
<td>-72.4</td>
<td>-28.8</td>
<td>-7.5</td>
<td>-0.0042</td>
<td>0.0058</td>
<td>2.1</td>
<td>3.3</td>
<td>3.5</td>
</tr>
<tr>
<td>1.0</td>
<td>164</td>
<td>-76.9</td>
<td>-11.6</td>
<td>-5.3</td>
<td>-0.0048</td>
<td>0.0017</td>
<td>2.1</td>
<td>2.8</td>
<td>3.2</td>
</tr>
<tr>
<td>2.0</td>
<td>159</td>
<td>-80.9</td>
<td>-0.35</td>
<td>-0.37</td>
<td>-26.10^6</td>
<td>29.0^7</td>
<td>2.2</td>
<td>2.5</td>
<td>3.0</td>
</tr>
</tbody>
</table>
3.3.2 Stress intensity factor in case of an interface pull-up

The interface line upturn was also considered. The stress in the model is raised when the interface deviates from the horizontal centreline of the specimen (Fig. 14a, stress in x-direction). On the advancing weld side, as a typical appearance, the interface was pulled up 0.5 mm from the nugget, intercepting the nugget border approximately at a right angle (case 2). The interface on the retreating side, however, was not altered. This was motivated considering the much softer back bend of the line upturn, and

Fig. 14. Case 2: Stresses in x-direction at a dynamic loading $\sigma_{\text{max}}$ of 40 MPa on a) advancing and b) retreating side.
since the affected interface is partially enclosed by the nugget. In addition, the EST of the “loaded” bottom sheet on the retreating side is increased, making it less susceptible to fracture.

\( \Delta K \) was determined by eqs. (1-5) and plotted as a function of the distance from the notch tip on the advancing side, Fig. 12. \( \theta \) was set to 40°, which implied crack growth perpendicular to the load direction. \( K_I \) was about 5 MPa√m, while \( K_{II} \) was close to 2. Eqs. (4) and (5) give a \( \Delta K \)-value of 5 MPa√m, see Fig. 12. The two different case 2-curves were derived by combining eqs. (1) and (2), and (2) and (3) respectively. The simplified model by Zhang et al. was not used in this case since the model assumes a straight interface slit.

The crack growth rate in the weld was determined from the curve for 6082 T4, FSW+PWAT, previously examined [12]. A \( \Delta K \)-value of 3.2 was taken into consideration in case 1. This corresponds to a growth rate of the order of 7.5×10⁻⁵ mm/cycle, see Fig. 15. For the crack to propagate vertically from the interface through the thickness of the top sheet (4 mm), it would require about 500 000 fatigue cycles. As seen in Table 3, fatigue lives in this range have been obtained by tool 2-samples at the applied max stress of 40 MPa (samples welded by tool 2 are considered since their

![Graph](image)

Fig. 15. Crack propagation rate in the weld at \( R=0.1 \) [12]. The growth rates which correspond to stress intensity factors \( \Delta K \) of 3.2 and 5 MPa√m (case 1 and 2 respectively) are marked.
structure most resemble the ideal case, when there is no pull up of the sheet interface. In the case of an interface pull up, $\Delta K = 5$ gives a crack growth rate of the order of $3.3 \times 10^{-5}$ mm/cycle. The resulting number of cycles, around 100 000, corresponds to the range of fatigue lives of the strongest samples at the given stress level. Due to the notch situation, crack initiation is assumed to take place relatively early during the testing.

4. Discussion

Considering concave and convex pin designs separately, the tool with the broader shoulder produced specimens with the best fatigue performance. The reason is most likely the increased heat energy supplied by a larger contact area tool-work piece. As previously pointed out, neither the shoulder diameter, nor the different end of pin-designs seems to have influenced the interface weld width, which is important for the strength of lap joints. The extent of the heat affected zones varies somewhat between the two shoulder sizes as seen in Fig. 16. Hardness is measured transverse to the weld, showing the drop due to dissolution of strengthening precipitates. The minimum hardness coincides with the TMAZ/HAZ border at the end of the tool shoulder. The base metal hardness may vary in the same batch material, which seems

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Fig. 16. Vickers hardness transverse to the weld, measured 2 mm through thickness on top (weld) sheet and on top surface.
to have been the case here. The relative hardness decrease is quite similar, just below 40 HV. Since the specimens produced by tool 1 showed lower base metal hardness, it was measured on additional (fractured) samples. It became apparent that top and bottom sheets of both the same and different hardness had been joined. In the latter case the hardness was approx. 100-106 HV in the top (weld) sheet, and 115-117 in the bottom sheet. No clear trend could be read out from this disparity however; samples that have been cut from the same welded plate have managed both long and short fatigue lives.

At the start of friction stir welding, the rotating tool is plunged down into the material and held for a few seconds, heating it up before it starts to transverse. As shown in Fig. 17, the hardness is quite constant along the weld: 75-80 HV (measured on top surface within 1 mm from the weld centreline). It is plotted as a function of the travelling distance from weld start to stop.

![Graph](image)

Fig. 17. Surface hardness as a function of distance from weld start for an overlap welded plate (500 mm length, weld stop at 482 mm), tool 1. Specimens have been cut from these plates transverse to the weld (cross welds). Hardness is measured on the top surface within 1 mm from the weld centreline.

The stirred up edges (metal flash) at the end of the tool shoulder, shear side in particular, tended to be rougher for the 15 mm-tools, concave and convex, than for the 18 mm-tools. This was especially the case in the middle of the weld run, not very much so close to start and stop. Disparity in the amount of flash can depend on different tool pressure which is manually adjusted. However, its influence as a stress
raiser is probably less important, since fracture is initiated in the interface between the sheets and not at the surface. The tool 2-series (concave pin, 18 mm shoulder) had very little surface metal flash, indicating a comparatively low tool pressure.

As is seen in Table 3 and Fig. 5, the fatigue strength of tool 1-specimens (concave pin, 15 mm shoulder) vary; some specimens have much higher strength than the mean. Also for tool 2, the strength of some individual tests is comparably high, whereas tools 3 and 4 (convex pin) produced little diffusion in data. Some possible reasons for this have already been mentioned. The one thing that differs tool 1 from 2 is the slightly smaller shoulder diameter. The 15 (15.1) mm, concave tool has produced a transverse weld width on the top surface of approx. 15.5-16 rather than 15 mm as opposed to tool 3. This may partly be the illusion of a coarser weld “beard”. It can also have been the result of the tool being slightly uncentered, which was not verified however. In any case, since the 18 mm-tool with concave end of pin gave quite outstanding fatigue performance compared to 18 mm convex tool, some of this credit has to lie in the pin design. The concave end of pin probably gives a material flow that breaks up the surface oxides more effectively [7]. In addition to the flutes, the hollowed out end (max 1 mm) increases the difference between the swept volume and the static volume of the probe. This effect was not easily evaluated, since both pins produced welds that appear to be free of pores. Back scattered SEM was used to analyse precipitates at the interface line between the sheets and in the weld nugget. It was seen that the interface adjacent to the nugget contained silicon oxides, see Fig. 18a. Both for specimens welded with convex (tool 3) and concave (tool 2) pin a few continuous streaks containing oxides were observed in conjunction with the stirred area. These were coarser in the specimens welded with the convex tool, see Figs. 18b, c. It is however difficult to judge to which extent, if at all, these streaks have influenced the mechanical properties.

It is probable that thinning of the overlap sheets (EST) by interface pull up, and in particular the creation of sharp notches, is more important for the joint strength than the weld penetration depth. According to Cederqvist et al [3], a shorter pin will result in less pull-up on the retreating side of the weld, since material flows less upwards near the bottom than at the middle of the pin. For concave pin the interface is closer to the bottom of the pin which, by comparison of etched specimen cross sections, has given less pull up (Fig. 10).
Fig. 18. a) The interface adjacent to the stirred nugget was found to contain silicon oxides. Continuous streaks containing oxides in b) concave-tool 2 and c) convex-tool 3 specimen.

With eqs. (1-5) the stress intensity factor $\Delta K$ can be estimated. The corresponding crack propagation rates are in fair accordance with the experimental results. As has been shown, the simplified approach developed by Zhang et al for overlap spot welds can also be used to estimate $\Delta K$ for friction stir overlap type joints.

Conclusions

- Joint efficiency of the FSW overlap in relation to static strength of T6 base material of the same dimensions was about 55%.

- The fatigue strength of the overlap weld is clearly inferior to that of the butt weld. At $10^5$ cycles it is about 20-30% of static overlap strength depending on tool selection (the corresponding value for butt joints is over 90%).

- Fracture is initiated in the highly stressed area where the weld cuts through the interface between the sheets. The stress state is further aggravated by notch formation due to the vertical material flow outside the pin.
• A broader shoulder with a concave end of pin design (tool 2) gave the best fatigue performance. This is most likely due to the increased contact area tool-work piece, and due to an improved flow path provided by the hollowed out end of pin. In particular, the minimal influence on the sheet interface when welding with tool 2 seems to have been beneficial for the fatigue strength of these specimens.

• The ability to produce a material flow that breaks up and disperses the surface oxides seems to be somewhat better for the concave than for the convex pin. For both pin ends the amounts of undispersed oxides were low though.

• The stress distribution in the test specimen was modelled. The stress intensity factor $\Delta K$ was determined. It was found that a simplified approach, developed to estimate $\Delta K$ for overlap spot welds, could be used also for friction stir overlap joints. The corresponding crack propagation rates were in fair accordance with the experimental results.

Acknowledgements

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References


6. Lars Mohlkert, Anders Norlin, Sapa Technology, internal correspondence
12. M. Ericsson, R. Sandström, *Fatigue crack propagation in friction stir welded and parent AA6082*, to be submitted
Fatigue of Friction Stir Welded T-joints

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Abstract
Friction stir welded T-joints of Al-Mg-Si alloy 6082 have been investigated. Static and dynamic strengths of the joints were determined. The specimens were loaded in two principally different ways. In the first mode the flanges of the T-specimen were subjected to tensile load (cross welds). In the second case a specially designed specimen holder was used to pull/bend the flanges in a direction parallel to the web. The static strength of the T-joined cross welds was comparable with butt joints, whereas the fatigue strength was reduced a considerable amount. It was found that a pin design with a convex end produced an interface less susceptible to fatigue on the advancing weld side, which was advantageous from a mechanical point of view. The tool pressure, manually adjusted, is likely to influence the material flow and thereby the interface line upturns which was seen for the concave pin. The occurrence of undispersed oxides in the weld was small both for specimens produced with concave and convex pin-tools, and was not likely to influence the fatigue properties.

The stress distribution in the test specimen was modelled. Three different methods were used to determine the stress intensity factor ΔK. It was seen that a simplified approach developed for overlap spot welds could be used to estimate ΔK also for friction stir overlap type joints. The corresponding crack propagation rate was in fair accordance with the experimental results.

Keywords: Friction Stir Welding, T-joints, tool design, fatigue, mechanical properties, stress intensity factor

1. Introduction
When friction stir welding (FSW) was introduced in the 1990’s it was mainly used for butt weld applications. Developments in the technology have made FSW a flexible welding process. Several geometries based on combinations of butt and lap joints are
today commercially applied, e.g. corner and T-section welds. As an example friction stir welding is used to seal the lid in hermetically closed boxes, such as cooling elements and heat exchangers. The frequent pressure changes in these make them susceptible to fatigue. In terms of joint integrity, FSW provides an excellent sealing. Another common area for T-type joints, potentially FS welded, are in airplanes, e.g. as a means of connecting stiffeners to the skin.

Special features are required by the FSW tool pin when it comes to welding of lap and T-type joints. The pin design is important to control the material flow at the interface between the sheets. Problems are due to interface “hook” formation, and insufficient

Fig. 1. Effect of vertical material flow on the interface between the sheets. On the retreating weld side (left), the “effective sheet thickness” (EST) of the top sheet is reduced. On the advancing side the interface forms a notch with the weld nugget. b) Joint line remnants (oxides) in weld nugget marked by arrows.
disruption and dispersion of Al surface oxides, see Fig. 1 [1-4]. The most dangerous situation is on the advancing weld side. Here, as on the retreating side, the interface is rolled upwards by vertical material flow outside the pin. The interface on the retreating side makes a greater bend, whereas on the advancing side it forms a sharp notch just outside the weld nugget raising the stress intensity factor to a higher degree. Conventional pins, appropriate for butt welding, have been shown to produce deficient quality in lap welding. To aid the material flow reentrant cuts or flutes can be machined into the pin. It was concluded that a so called Triflute-pin with a hollowed out (concave) end design gave better fatigue performance for overlap FS welds [3]. In particular, the influence on the sheet interface was minimal when welding with this pin and a broad shoulder (O=18 mm). If fracture is by shear through the interface, a wider interface weld width is beneficial. This may be achieved by double pass (DP) welding by introducing a separation distance between the subsequent weld runs. In addition the oxide disruption becomes more effective, both effects resulting in stronger welds compared to single pass FSW [1,2]. Without altering the static volume of the pin, the Flared Triflute™ design increases the width of the joining zone; the flute lands are flared out at an inverted angle to the core so as to increase the tip diameter. The Flared Triflute™ probe also improves oxide disruption; however the interface line upturns still exist. Mishina and Norlin showed that a counter-clockwise tool rotation direction lead to a reduced thinning of the upper work-piece (for a cylindrically shaped Flared Triflute™ probe) [2]. The flared out flute lands may not be preferred in practice since a start hole has to be drilled in the work piece.

2. Experimental
2.1 Material and welding
The Al-Mg-Si alloy 6082 in temper condition T6 (artificially aged) was used. The composition of the material is given in Table 1. One sheet: 500 mm long, 80 mm wide and 4 mm thick was friction stir (FS) welded along the centre of its width to the long edge of another sheet: 500 mm long, 74 mm wide and 15 mm thick. Test specimens were cut out so that the width of the flanges was 40 mm. Sapa performed the welding. The feeding speed of the material was 300 mm/min, the rotating speed of the tool 1500 rpm.
Table 1. Chemical composition of the 6082 alloy (wt%).

<table>
<thead>
<tr>
<th>Reference</th>
<th>Si</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Mg</th>
</tr>
</thead>
<tbody>
<tr>
<td>Typical</td>
<td>0.7-1.3</td>
<td>≤0.50</td>
<td>≤0.10</td>
<td>0.4-1.0</td>
<td>0.6-1.2</td>
</tr>
<tr>
<td>Actual batch</td>
<td>1.0</td>
<td>0.20</td>
<td>&lt;0.02</td>
<td>0.48</td>
<td>0.65</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Reference</th>
<th>Cr</th>
<th>Ni</th>
<th>Zn</th>
<th>Ti</th>
<th>Pb</th>
</tr>
</thead>
<tbody>
<tr>
<td>Typical</td>
<td>≤0.25</td>
<td>≤0.20</td>
<td>≤0.10</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Actual batch</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.01</td>
<td>0.02</td>
<td>&lt;0.01</td>
</tr>
</tbody>
</table>

Four test series were made. Two modifications based on a Triflute™ probe were used: the end being either concave (hollowed out) or convex (protruding). The length of the unmodified probe or pin was 6 mm. The probe diameter was 6 mm. In the convex case the maximum length of the pin was increased with 1 mm, while it was reduced with 1 mm in the concave case (Fig. 2). The second tool variable was the shoulder diameter; being either 15 or 18 mm. The designation of the tools is given in Table 2. The same tools and welding parameters were used to produce FSW lap joints in a previous investigation [3].

Fig. 2. FSW tool outline. The shoulder has an elevated edge of 1 mm. The concave pin is hollowed out about 1 mm at the most. The convex pin has a 1 mm top (right).

2.2 Test procedure

Static and dynamic strengths of the welds were determined. The testing was carried out in a servo-hydraulic machine equipped with an actuator of 250 kN load capacity. The specimens were loaded in two principally different ways. In the first case the
Table 2. Designation of the various tools

<table>
<thead>
<tr>
<th>Tool</th>
<th>Pin end</th>
<th>Shoulder diameter, mm</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Concave</td>
<td>15</td>
</tr>
<tr>
<td>2</td>
<td>Concave</td>
<td>18</td>
</tr>
<tr>
<td>3</td>
<td>Convex</td>
<td>15</td>
</tr>
<tr>
<td>4</td>
<td>Convex</td>
<td>18</td>
</tr>
</tbody>
</table>

Flanges of the T-specimen were subjected to tensile load (cross welds), Fig. 3. In fatigue a sinusoidal load-time function was used. The load ratio R was set to 0.5. Maximum forces of 16 to 22.4 kN were tested (100-140 MPa max net section stress). The oscillation frequency was 15 Hz.

![T-welded test specimen cross section](image)

Fig. 3. T-welded test specimen cross section (a) and from above (b).

In the second case a specially designed specimen holder was fastened in the upper grip of the testing machine. The flanges rested against the bottom of the holder, while
the web of the specimen (B, Fig. 3) extended through a central opening and was fastened in the lower grip of the machine, see Fig. 4. This device created a bending moment on the flanges from below. A low frequency of 5 Hz was set in order to eliminate resonance during fatigue testing. The load ratio was set to 0.1 and maximum moments of 76 to 160 Nm were tested (corresponding to applied loads of 3.8-8 kN or 12-25 MPa net section stress. In calculating the net section stress, the area of the flange cross section times two was used since the flanges were subject to separate pressure.). During testing the behaviour of the load wave form was checked by oscilloscope.

Macro studies of etched weld cross sections were performed to evaluate weld formation and interface line upturns (“hooking”).

3. Results

3.1 Mechanical testing

The first case involved loading of the specimen flanges. The situation is similar to testing of butt cross welds. The static strength in this mode was 223-254 MPa depending on tool; tool 3 gave the highest, tool 2 the lowest strength. The fatigue
results are plotted in Fig. 5. The individual results for each specimen are collected in Table 3. Regarding the difference in tool design, the data for tools 1, 3 and 4 fall in the same scatter band, and their average trendlines coincide. The convex pin seems to produce better results, in particular at the higher stress levels. The extrapolated fatigue strength at $2 \times 10^6$ load cycles is between 20-38 MPa. Tool 2 (18 mm, concave end of pin) produced specimens with the lowest average fatigue strength.

In the second case the flanges were pulled in a direction parallel to the web and a bending moment was inflicted on the flanges. This type of loading simulates pressure changes in hermetically closed boxes. In applications where a FSW joint is used to seal boxes, e.g. in cooling elements, the pressure changes can be in the range of 10-20 bar, corresponding to stresses of 1-2 MPa. Since the load was normal to the plane of the flanges, these gradually deflected from the web outside the weld connection area.

The static strength of the T-joints in the “bending” mode was 48 MPa (average of all tools, 7 tests). The results of the fatigue testing are collected in Table 4 and plotted in Fig. 5. At a max moment of 76 Nm (12 MPa stress) the welds managed from 7700 to
Table 3. Fatigue data for T-welds: cross weld loading (R=0.5)

<table>
<thead>
<tr>
<th>Tool</th>
<th>Max Load (MPa)</th>
<th>Cycles</th>
<th>Fracture side of weld</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>140</td>
<td>116616, 53626</td>
<td>advancing</td>
</tr>
<tr>
<td>1</td>
<td>120</td>
<td>83489, 66360</td>
<td>adv</td>
</tr>
<tr>
<td>1</td>
<td>100</td>
<td>125141, 504135, 204558</td>
<td>adv</td>
</tr>
<tr>
<td>2</td>
<td>140</td>
<td>37392, 47376</td>
<td>adv</td>
</tr>
<tr>
<td>2</td>
<td>120</td>
<td>99652, 84700</td>
<td>adv</td>
</tr>
<tr>
<td>2</td>
<td>100</td>
<td>212676, 136184, 219093, 134365</td>
<td>adv</td>
</tr>
<tr>
<td>3</td>
<td>140</td>
<td>154316, 84611</td>
<td>retreating</td>
</tr>
<tr>
<td>3</td>
<td>120</td>
<td>249357, 113529</td>
<td>retr</td>
</tr>
<tr>
<td>3</td>
<td>100</td>
<td>173832, 163822, 367537, 125060</td>
<td>adv, adv, retr, adv</td>
</tr>
<tr>
<td>4</td>
<td>140</td>
<td>80467, 57719</td>
<td>retr, adv</td>
</tr>
<tr>
<td>4</td>
<td>120</td>
<td>180038, 288220</td>
<td>adv, retr</td>
</tr>
<tr>
<td>4</td>
<td>100</td>
<td>636431, 157053, 173780, 185655</td>
<td>retr, retr, retr, retr</td>
</tr>
</tbody>
</table>

Table 4. Fatigue data for T-welds: pull/bending of the flanges (R=0.1)

<table>
<thead>
<tr>
<th>Tool</th>
<th>Max Load (MPa)</th>
<th>Cycles to 90° yielding of the flanges</th>
<th>Most severed side of weld</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>25</td>
<td>1840</td>
<td>adv</td>
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<tr>
<td>1</td>
<td>17.5</td>
<td>3709, 4001</td>
<td>adv</td>
</tr>
<tr>
<td>1</td>
<td>12</td>
<td>8834, 7699, 11412</td>
<td>adv</td>
</tr>
<tr>
<td>2</td>
<td>25</td>
<td>2057</td>
<td>adv</td>
</tr>
<tr>
<td>2</td>
<td>17.5</td>
<td>10748, 3918</td>
<td>retr, similar</td>
</tr>
<tr>
<td>2</td>
<td>12</td>
<td>36766, 9186, 17424</td>
<td>retr, similar, similar</td>
</tr>
<tr>
<td>3</td>
<td>25</td>
<td>3049</td>
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</tr>
<tr>
<td>3</td>
<td>17.5</td>
<td>5535, 5307</td>
<td>retr, adv</td>
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<td>12</td>
<td>13025, 10742, 15036</td>
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</tr>
<tr>
<td>4</td>
<td>25</td>
<td>2734</td>
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</tr>
<tr>
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<td>17.5</td>
<td>10989, 6380</td>
<td>similar, adv</td>
</tr>
<tr>
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<td>29747, 15923, 14389</td>
<td>adv, adv, similar</td>
</tr>
</tbody>
</table>

36700 fatigue cycles. The results of the four tools are different from the mode 1 testing, since the concave pin, 18 mm shoulder (tool 2) produced the welds which often exhibited the best strength. The scatter in the tool 2-data, however, is greater than for the two convex tools. The concave pin, 15 mm shoulder (tool 1), had the weakest welds in testing mode 2.
For none of the “bending” tests final fracture was the result of the crack propagating all the way to the top surface. Propagation started at the interface ends bordering the weld nugget. The cracks grew into the weld on both sides, approximately 2.5-3 mm, until the load bearing capacity of the flanges was reduced sufficiently to make them yield a full 90°. This is where the tests were interrupted. For 2/3 of the specimens the flange on the advancing weld side was closest to a total break-trough. This did not necessary mean that it had the longest propagated crack, depending on the angle of propagation; cracks on the advancing side propagated nearly vertically while retreating side cracks propagated more diagonally.

The T-joined cross welds were compared with previous data for FSW butt and lap joints of the same material [5,3], Fig. 6. Similarity aside, the T-joint fatigue strength is well below that of butt joined specimens (tested at the same load ratio of 0.5). The static strength, however, is comparable being in the range of 221-245 MPa for T6 butt welds. The results imply that the fatigue strength (max stress) of the former at $10^5$

![Graph showing fatigue testing results](image)

Fig. 6. Results of fatigue testing of FSW T-joints compared with previous data for overlap and butt welds [3,5]. In all cases the load was applied transverse to the welds.
cycles is over 90% of static tensile strength while that of the latter is only about 50% (regardless of tool). As is seen in the figure, the T-joint cross weld fatigue strength is in excess of that of the lap welds. The latter were tested at a lower load ratio of R=0.1.

3.2 Macro studies

The weld has penetrated about 1 or 2 mm into the lower vertical plate depending on the end of pin design: hollowed out (concave) or protruding (convex). The weld width in the joining zone at the interface between the plates did not vary much, this width being 6 to 7 mm (the pin diameter being 6 mm).

Two different shoulder diameters were used: 15 and 18 mm. As shown in Fig. 7, the width of the top surface shoulder mark varies somewhat for concave and convex tools of the same shoulder diameter. “Concave” specimens (tools 1 and 2) have the broadest shoulder indentations. This can depend on different tool pressure since it was manually adjusted.

In Figs. 8-9 the effect of different pin design on the interface movement can be discerned. On the retreating weld side, the “effective sheet thickness” (EST) of the top

![Image](image-url)  

Fig. 7. Weld appearance of T-joints produced by the four different tools. The difference between the tools is in shoulder diameter; 15 or 18 mm, and end of pin design; concave or convex. From left to right is: 15 (15.1) mm concave tool, 15 mm convex, 18 mm concave, and 18 mm convex. Advancing weld side is up in the picture.
sheet was about the same for concave and convex pin. However, when welded with concave pin, the rolled up interface on the advancing side formed a sharper radius with the nugget than for convex pin. For concave pin the apex of this notch pointed nearly vertically towards the top surface, see Fig. 9a. Compare convex tool 4, which gave the least extreme interface movement. The effect of notch sharpness on fatigue life could be clearly seen in some instances. For example, at 12 MPa max stress a specimen welded with convex tool 4 managed 29747 fatigue cycles whereas a specimen welded with concave tool 1 endured only 8834 cycles, Table 4. In the
Fig. 9. The interface line upturns on the advancing side are shown enlarged for a) tool 2 and b) tool 4. Figs. c-d show the effect in FSW overlaps (same tools used as in the present investigation), c) tool 2 and d) tool 4 [3].

former case crack propagation was at an angle about 45° to the plane of the top sheet, in the latter close to 90°.

Remnants of the oxide film in the weld zone may lower the mechanical properties. The concave end of pin is thought to give a material flow that breaks up and disperses the surface oxides more effectively. The hollowed out end increases the difference between the swept volume and the static volume of the pin, thereby improving the flow path. In the investigated samples there was no occurrence of pores or extended cavities.

3.2 Numerical simulation of stresses

The stress distribution in the test specimen was described by a finite element model. The geometric feature which affects the stress in the model is foremost the root radius in the gap between the overlap sheets. A larger radius lowers the maximum stress.
The stress is raised when the interface deviates from the horizontal centreline of the specimen. On the advancing weld side a pull-up of the interface was modelled 0.5 mm from the nugget, based on the real appearance (compare Fig. 9a). The root radius was typically about half the width (~35 μm) of the interface slit. These values were used in the finite element model. The notch is intercepting the nugget border approximately at a right angle, Fig. 10a. On the retreating weld side the interface was approximated with a straight column all the way to the nugget. This was motivated considering the much softer back bend of the interface line upturn on the retreating side. In addition the affected interface is partially enclosed by the nugget on this side, see Fig. 1.

Fig. 10 plots the numerically calculated stress in the T-specimen for an applied fatigue load of 8 kN in mode 2-testing. The locations are at the interface adjacent to the weld nugget on the advancing (a) and retreating side (b, c). The extremely high stresses at the notches in the elastic model are strongly reduced in reality due to plastic deformation. The elastic stresses around the notches are however useful to estimate the stress intensity factors. The variation of the stress in the y-direction $\sigma_y$ due to notch can be expressed as [6]:

$$\sigma_y = \frac{K_I}{\sqrt{2\pi r}} \cdot \cos \left( \frac{\theta}{2} \right) \left( 1 - \frac{2}{3} \sin \frac{\theta}{2} \cdot \sin \frac{3\theta}{2} \right)$$  \hspace{1cm} (1)

Fig. 10a. Stress distribution in x-direction near the notch end on the advancing weld side. The max stress is within the stress singularity.
Fig. 10 continuing. Stress distribution in (b) x- and (c) y-direction at the notch end on the retreating weld side. The max stress is within the stress singularity.

where $K_I$ is the stress intensity factor, $r$ the distance from the notch, and $\theta$ the angle of deviation from the slit centreline. $K_I$ evaluated from (1) is plotted in Fig. 11 for $\theta=0^\circ$.

The stress was determined from the FE-model on the retreating weld side ($K$ may also be evaluated on the advancing side with an upturn, see Ref [7]). It is seen that $K$ increases rapidly up to a distance about 10 $\mu$m from the notch tip. At a distance corresponding to the root radius (17.5 $\mu$m) $K$ is 17.4 MPa\sqrt{m}, and $\Delta K$ thereby 15.7.
Fig. 11. The variation of the stress intensity factor $K$ with the distance from the notch tip according to eq. (1).

At the fatigue loads tested crack propagation was initiated almost immediately from start. This manifested itself in that the unwelded column between upper and lower sheet became clearly visible, the crack originating from the notch. As a mean value of all bending tests at 10.8 MPa stress range, the number of fatigue cycles was slightly above 10,000. That implies a crack propagation rate in the range of $3 \times 10^4$ mm/cycle average. When comparing with a weld propagation curve for 6082 T4+PWAT, Fig 12, it gives a stress intensity factor $\Delta K$ of $\sim$10 MPa$\sqrt{\text{m}}$ [8]. At 25 MPa stress (8 kN max load), the propagation rate would be around $1 \times 10^3$ mm/cycle and $\Delta K$ thereby $\sim$14 MPa$\sqrt{\text{m}}$.

Radjia and Zhang [e.g. 9] showed that if the structural stresses, i.e. $\sigma_{uu}$, $\sigma_{uu}$, $\sigma_{uu}$, $\sigma_{uu}$, $\tau_{uw}$ and $\tau_{uw}$ around an overlap spot weld are available, the stress intensity factors can be determined according to:

$$K_I = \frac{1}{6} \left[ \frac{\sqrt{3}}{2} \left( \sigma_{uu} - \sigma_{uu} + \sigma_{uu} - \sigma_{uu} \right) + 5\sqrt{2} (\tau_{uw} - \tau_{uw}) \right] \sqrt{I}$$

(2)

$$K_{II} = \left[ \frac{1}{4} (\sigma_{uu} - \sigma_{uu}) + \frac{2}{3\sqrt{5}} (\tau_{uw} + \tau_{uw}) \right] \sqrt{I}$$

(3)
Fig. 12. Crack propagation rate in the weld at R=0.1 [8]. The growth rate corresponding to a stress intensity factor $\Delta K$ of 14 is marked in the figure.

$$K_{eq} = \sqrt{K_I^2 + K_H^2}$$

and

$$\Delta K = (1 - R)K_{eq}$$

where $\sigma$ refers to stress in the x-direction and $\tau$ : stress in the y-direction. The prefixes denote: $u =$ upper sheet, $l =$ lower sheet, $i =$ inner side of sheet and $o =$ outer side of sheet. The equations were used to estimate $\Delta K$ along the notch (on the retreating side, since Zhang’s model [9] assumes a straight interface slit) where the crack starts to propagate. By ignoring $K_H$ which is reasonable in the pull/bending mode, $\Delta K = K_{eq}$ (1-R). The structural stresses were taken from the FE-model at various distances $r$ from the notch tip. The values are read at the surface of the interface between the welded sheets. The elastic calculations give stresses around 700 MPa on the retreating weld side, see Table 5. It is seen that at a sufficient distance from the tip, $K$ is quite constant along the interface at 12.3-12.5 MPa/$\sqrt{m}$. The corresponding $\Delta K$ is around 11 MPa/$\sqrt{m}$, which is somewhat less than the computation above.
Table 5. Structural stresses and stress intensity factors in the T-specimen; \( \sigma \) refers to stress in the x-direction and \( \tau \) to stress in the y-direction. Stresses on the retreating weld side at an applied fatigue load of 8 kN in mode 2-testing. The stresses were determined at the upper \((\sigma_{uu}, \tau_{uu})\) and lower \((\sigma_{ul}, \tau_{ul})\) surfaces of the sheet interface at a distance \( r \) from the notch tip, on the top sheet surface \((\sigma_{ub})\), and 4 mm below the interface \((\sigma_{bl})\). \( K_I \) was determined by eq. (2).

<table>
<thead>
<tr>
<th>( r ) (mm)</th>
<th>( \sigma_{uu} ) (MPa)</th>
<th>( \sigma_{ul} ) (MPa)</th>
<th>( \sigma_{ub} ) (MPa)</th>
<th>( \sigma_{bl} ) (MPa)</th>
<th>( \tau_{uu} ) (MPa)</th>
<th>( \tau_{ul} ) (MPa)</th>
<th>( K_I ) (MPa(\sqrt{m}))</th>
<th>( \Delta K ) (MPa(\sqrt{m}))</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.5</td>
<td>835</td>
<td>-716</td>
<td>-198</td>
<td>5.4</td>
<td>-0.077</td>
<td>0.048</td>
<td>12.3</td>
<td>11.1</td>
</tr>
<tr>
<td>1.0</td>
<td>739</td>
<td>-723</td>
<td>-95.6</td>
<td>4.2</td>
<td>-0.022</td>
<td>0.017</td>
<td>12.4</td>
<td>11.2</td>
</tr>
<tr>
<td>1.5</td>
<td>703</td>
<td>-718</td>
<td>-50</td>
<td>3.4</td>
<td>-0.004</td>
<td>0.003</td>
<td>12.5</td>
<td>11.2</td>
</tr>
<tr>
<td>2.0</td>
<td>685</td>
<td>-703</td>
<td>-22</td>
<td>2.1</td>
<td>-0.002</td>
<td>0.002</td>
<td>12.5</td>
<td>11.2</td>
</tr>
</tbody>
</table>

An upper limit of \( K_I \) can be evaluated by determining the bending moment on the T-specimen flanges. Formulas were used pertaining to an edge crack in a strip [10]:

\[
K_I = \sigma_0 \sqrt{\pi \cdot a \cdot f_\theta} \left( \frac{a}{W} \right)
\]

\[
\sigma_0 = \frac{6M}{BW^2}
\]

where \( B \) is the thickness, \( a \) is the crack length, and \( W \) is the specimen width. This approach gave a \( \Delta K \) of 23 MPa\(\sqrt{m}\).

**Discussion**

The stress intensity factor is smaller at the interface/weld conjunction on the retreating weld side than on the advancing side. This means that when fracture was on the retreating side, or there were similar amounts of propagation on advancing and retreating side, the notch situation was probably not as critical. Many specimens that have failed on the retreating weld side have endured a high number of cycles. It may be the reason why welding with convex pin resulted in comparatively good mechanical performance for T-welds (compare Table 3). For lap welds in the previous investigation, the advancing tool side was chosen to coincide with that of highest total stress concentration (due to geometry) [3]. Therefore fatigue fracture was bound to occur on advancing side regardless of the tool used. In that investigation concave tool 2-specimens had higher fatigue strengths than the “convex” specimens. The concave pin gave much greater interface hooking in the present T-joint welding; compare the
interface line upturns in T- and lap welds made by the same tools in Figs. 9a-d respectively. It is likely that the comparably low mechanical strength of the “concave” T-specimens is due to a high tool pressure which has produced two unfavourable side effects. First, the sheared up (adv. side) weld toe forming on the top surface is of significant size for tool 2-specimens (Fig. 13). It has previously been shown that the weld toe influences crack initiation for FSW butt welds [5]. More importantly, an increased material flow and thereby “hook” formation is seen at the interface, Fig. 1a. Especially for the broad tool shoulder the increased pressure has been detrimental, since the web (15 mm in width) does not support the entire shoulder diameter. This has lead to some excess material originating from the top sheets bulging out at the sides of the web, see Fig. 13.

![Etched specimen cross section, tool 2 (advancing side to the right). The result of the tool shoulder being wider than the web is seen; excess material has leaked out at the sides of the web.](image)

The T-joint strength in pull/bend testing mode was much lower than in cross-weld testing. This is easily realised when considering the bending moment introduced by the specimen holder, and the fact that the two parts of the T-specimen were not joined over their entire interface. No standardised procedure has been established for the testing of T-joints, which makes evaluation of data somewhat arbitrary. Erbslöeh et al have conducted tensile tests on FSW T-joints in which a stringer (web) is pulled off a “skin” [11]. Cross weld testing of the skin was also performed. The weld penetrated the joint interface close to the edges of the web. For this case the backing bar edges have to lie tight on to the corners of the T-joint, to prevent material from leaking out. However, since a high notch factor was expected, a fillet radius was introduced by rounding off the backing bar edges. In the mentioned investigation joint efficiencies
relative to base material strength (6013-T4) were about equal for the two testing modes at 78% and 74% respectively. The tensile strength of the web was much more sensitive to processing parameters however, which was seen at an increased welding speed over 300 mm/min. For that occasion an interface line defect evolved due to material flow in the fillet radius. In conclusion the authors recommend T-joints with sharp corners, which can be welded over a broader parameter range and with simpler tools. Due to the strict backing bar requirements, however, it is often not practical to produce T-joints in which the weld connects the entire interface. Therefore, in practice, unwelded columns will often exist on both sides of the weld.

With eq. (1) the stress intensity factor $K_I$ has been estimated. The corresponding crack propagation rate is in fair accordance with the experimental results. As has been shown, the simplified approach developed by Zhang et al for overlap spot welds can also be used to estimate $\Delta K$ for friction stir overlap type joints. By using a formula which incorporates the bending moment, an upper limit for $K_I$ was estimated.

Conclusions

- The static strength of 6082 T-joined cross welds is comparable with butt joints, being in the range of 223-254 MPa. The fatigue strength of the former is about 50% of the static tensile strength at $10^7$ cycles (whereas over 90% for butt welds).

- Due to material flow outside the pin, the interface between the sheets is pulled up. In specimens welded with the concave end of pin, the notch radius between the weld nugget and the interface slit was particularly sharp on the advancing side. The convex pin produced an interface less susceptible to failure on the advancing side (mainly in cross weld testing), which was beneficial for the mechanical strength.

- The scatter in mechanical data was larger for specimens welded with the concave pin.

- For both pin designs the amounts of undispersed oxides were low.

- Since the concave pin enhances the material flow it seems important to control the tool pressure in order to avoid excessive “hooking”.
• The stress distribution in the test specimen was modelled. Three different methods were used to determine the stress intensity factor $\Delta K$. It was seen that a simplified approach developed for overlap spot welds could be used to estimate $\Delta K$ also for friction stir overlap type joints. The corresponding crack propagation rate was in fair accordance with the experimental results.

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