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Linear and rotary friction welding review

Wenya Li^{*1}, Achilles Vairis² , Michael Preuss³ and Tiejun Ma¹

Friction welding (FW) is a high quality, nominally solid-state joining process, which produces welds of high structural integrity. Rotary friction welding (RFW) is the most commonly used form of FW, while linear friction welding (LFW) is a relatively new method being used mainly for the production of integrally bladed disc (blisk) assemblies in the aircraft engine industry. Numerous similar and dissimilar joints of structural metallic materials have been welded with RFW and LFW. In this review, the current state of understanding and development of RFW and LFW is presented. Particular emphasis is placed on the process parameters, joint microstructure, residual stresses, mechanical properties and their relationships. Finally, opportunities for further research and development of the RFW and LFW processes are identified.

Keywords: Linear friction welding, Continuous drive friction welding, Inertia friction welding, Process parameters, Microstructure, Mechanical properties

List of symbols

a	oscillation amplitude
A	weld area
C_p	specific heat capacity
D	thickness of the specimen
f	oscillation frequency
F_N	normal force
k	thermal diffusivity
P_N	axial pressure, friction pressure
q	heat generation rate or heat input to the welding interface
q_0	average heat generation rate
r	radial distance
R	radius of the specimen
t	time, welding time, friction time
T	oscillatory cycle period, temperature
T_0	initial temperature
T_m	melting temperature (solidus)
v	sliding velocity
W	width of the specimen
λ	thermal conductivity
μ	friction coefficient, nominal friction coefficient
ρ	material density
σ_y	yield stress
τ_f	friction stress or friction shear stress
τ_y	shear yield stress
ω	rotational speed in RFW, angular frequency in LFW

Introduction

Friction welding (FW) is a nominally solid-state process for joining materials by frictional heat that is generated from the movement of one component relative to another

under a normal force, which operates in a completely different mode than fusion welding which requires the formation of a molten pool of material. It shares the common characteristic of using friction as the means to generate intense localised heat, with the rotary friction welding (RFW) being the most popular variant of these technologies. In this group, linear friction welding (LFW) and friction stir welding (FSW) are relatively new solid-state bonding processes aiming at extending the application of RFW to non-axisymmetric components and thin structures. FW is now being used for metals and thermoplastics in a wide variety of aerospace and automotive industrial applications. In spite of this, much remains to be understood on the nature of the process and opportunities for further research and development should be identified.¹ Of all the FW variants in use, FSW has attracted a lot of attention since its invention in 1991 for its distinct advantages in joining sheets of lightweight alloys, and has been reviewed thoroughly.²⁻⁴ However, there are very few comprehensive reviews which focus on RFW and LFW. Research on RFW, LFW and orbital friction welding (OFW) has been reviewed by Maalekian¹ in 2007, who tried to clarify some inconsistencies in the interpretation of the FW process. Only two short reviews on LFW have been presented by Nunn⁵ and Bhamji *et al.*⁶ recently. Over the past few years, further advances have been made in RFW and LFW. It is therefore considered timely to fill this gap. The present paper provides an extensive summary of the state-of-the-art of understanding and development on both LFW and RFW.

The RFW and LFW processes

The RFW process

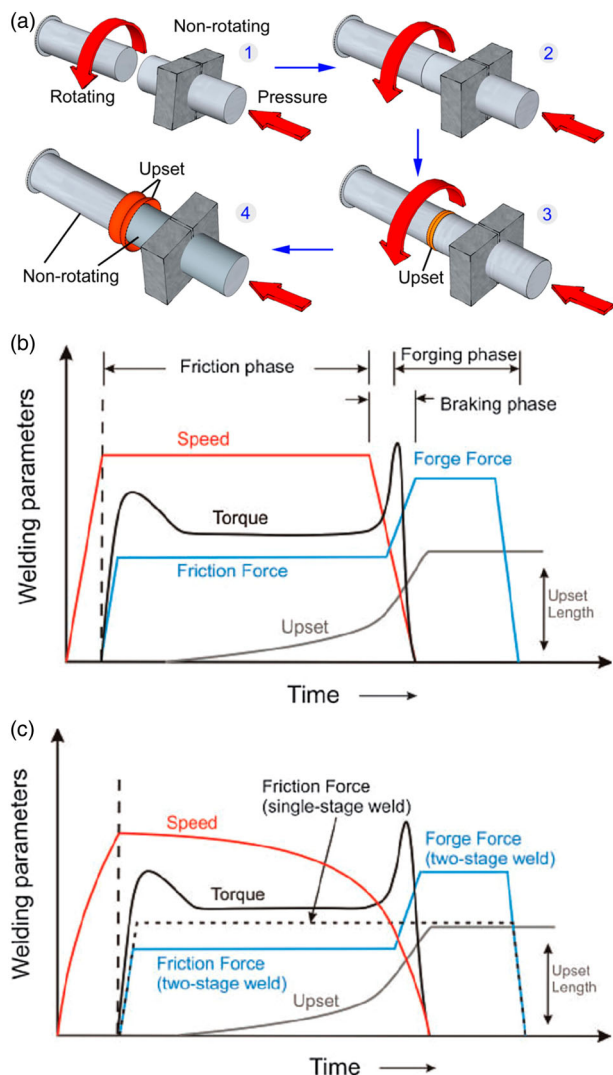
RFW is the most common form of FW. Its basic principle is illustrated in Fig. 1a⁷ where one component is held stationary while forced to rub against another rotating

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1 Schematic of the RFW process ^{a7} and process phases of CDFW ^b and IFW ^{c1}

part under normal pressure. There are two process variants: continuous-drive friction welding (CDFW, also termed direct-drive friction welding) and inertia friction welding (IFW, also termed inertia-drive or inertia welding). The biggest difference between these two methods is the means of supplying energy to the welding interface. In CDFW, the rotating part is attached to a motor driven unit which keeps a constant rotation speed during the whole welding process (Fig. 1b).¹ The process runs until a braking force is applied or axial shortening (also known as burn-off or upset) is produced. In the case of IFW the rotating part is connected to a flywheel which, after being accelerated to a desired rotational speed, is disconnected from the drive motor and welding commences. Here, it is the rotating flywheel which supplies the kinetic energy to transform to frictional heat for welding at the interface, while the speed of the flywheel gradually decreases to zero (Fig. 1c).¹ There are certain cases where IFW does not use the additional application of a forging force (Fig. 1c), but in the majority of applications it requires a forging stage (Fig. 1c). It should be pointed out that the definition of single-stage and two-stage process in Fig. 1c is vague and limited to RFW. While friction pressure can remain constant throughout the process, it

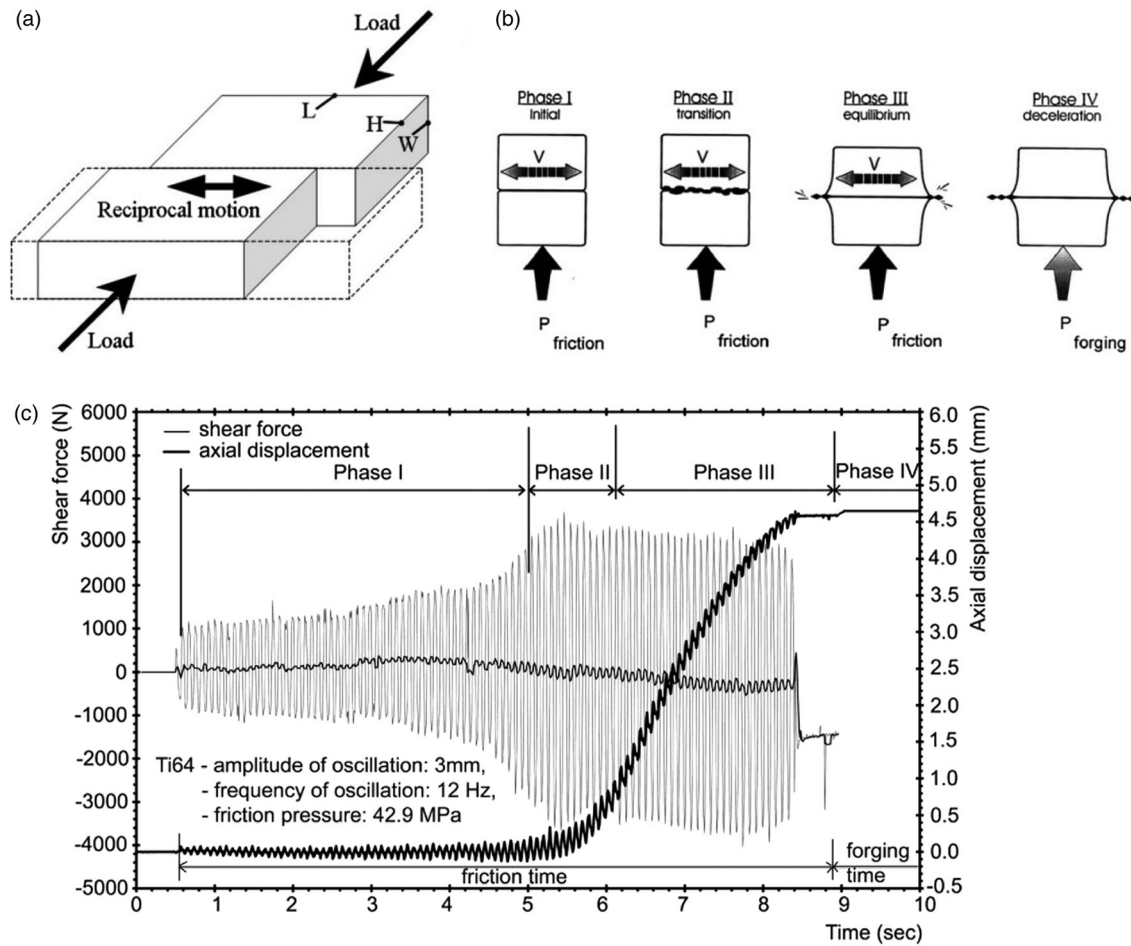
can also be applied in two or more steps. In addition, both single-stage pressure and two-stage pressure processes are followed by forging at the end of welding, without altering the nature of the phases/physical processes of RFW. In general, RFW is composed of two phases: the friction phase where the material is heated up and necessary upset is formed, and the forging phase to consolidate the weld (Fig. 1b and c). Due to the different means of energy input, IFW can fulfil much higher input energy specifications compared to CDFW, and for this reason aero-engine manufacturers use IFW and the automotive industry CDFW.

The LFW process

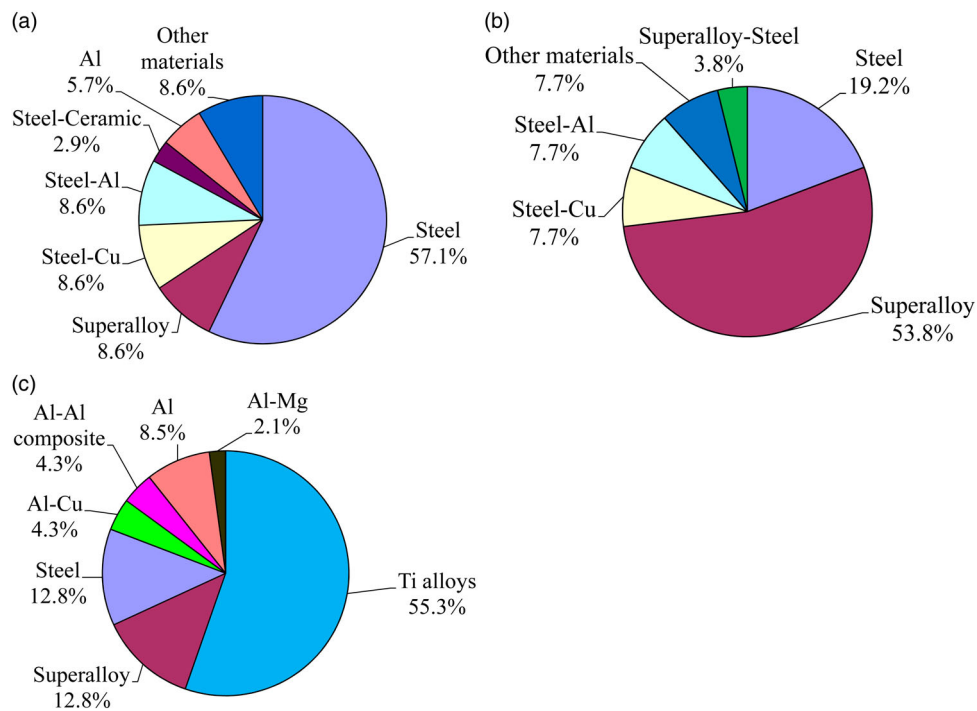
Linear friction welding is a process where frictional heat is produced as one component is moved in a direct reciprocating mode relative to another under normal pressure (Fig. 2a). This process is divided into four distinct phases,⁸⁻¹⁰ which include the initial phase, the transition phase, the equilibrium phase and the deceleration (or forging) phase (Fig. 2b). The detailed changes of process variables during LFW are illustrated in Fig. 2c. The process description is based on Ti-6Al-4V (Ti64) experiments and it is assumed that all materials which are ductile like metals and their alloys do follow it. Following the first two very short phases, there is phase III when axial shortening of the workpieces starts as a result of plasticity and material is being ejected from the weld interface. When the desired upset has reached a preset limit the two objects are brought to rest very rapidly (less than 0.1 s) and forging pressure may remain to preserve the bond. The last phase is considered important in the FW industry to consolidate the weld and may affect residual stresses developed in the joints.^{8,9} When comparing to Fig. 1b and c, it should be pointed out that the predominant phase definition of RFW may be vague and limited as it also includes the distinct initial, transition and equilibrium phases in the RFW process. As expected, the physical processes that the joint material undergoes in RFW and LFW are quite similar. The torque (in the case of RFW) and shear force (in the case of LFW) attest to the correctness of the four-phase definition.

Historical background and applications of RFW and LFW

Rotary friction welding is the oldest friction joining process dating back to 1891 when a patent by Bevington¹¹ revealed a process for welding together the ends of metal wire, rods of metal, and other like pieces of solid metal. A rapidly revolving die was used within which the joined ends of the rods were held. CDFW has been used commercially since the 1940s and IFW in the early 1960s.^{1,12} Since then, they have been widely applied to join various steels, superalloys, aluminium alloys, titanium alloys, copper alloys and dissimilar materials combinations (Fig. 3a and b), and have found a wide range of applications in manufacturing from tiny daily necessity items to complex and expensive aero-engine components, for the most common pure metals to the most advanced alloys. In recent years, dissimilar material combinations (e.g. dissimilar steel alloys, steel-aluminium, steel-copper, superalloy-steel, steel-ceramic) and superalloys (e.g. IN718, RR1000 and U720Li) have been widely studied.



2 Schematic of the LFW process *a*, its process phases *b*, and process variables during LFW *c*^{8,9}



3 Materials experimentally studied in published research of *a* CDFW and *b* IFW since 2007, and *c* LFW since 1992. (Data from Web of Science®, Thomson Reuters)

The first patent that described the LFW process appeared in 1929.⁵ The Caterpillar Tractor Company (Peoria, IL, USA) was next in 1969,¹³ describing a machine to produce the linear reciprocating motion. The process was first studied in a systematic way in the UK in the 1980s at The Welding Institute (TWI), the University of Bristol and Rolls-Royce plc. In spite of the limited base of research machines available and the high capital requirements for industrial machines, compared to other forms of FW, a small number of industrial countries have research groups which produce an increasing number of research papers.

In particular, in Europe, the UK is very active with TWI and Rolls-Royce plc playing an important role in research and information dissemination on LFW, through university based research at Birmingham, Manchester, Cambridge, Bristol, Heriot-Watt and Cranfield carrying out research which ranges from mathematical modelling to microstructural studies. The commercial firm Thompson designs and produces LFW machines for industrial uses. In the Russian Federation the commercial firm UMPO and the Ufa State Aviation Technical University have been active in research in recent years. In Italy, the University of Bologna and that of Palermo are also active, while the latter operates its own research machine. In France, the commercial firm ACB (Cyrill Bath) designs and manufactures LFW machines for research and production.

In Asia, the Northwestern Polytechnical University (NPU) in China is the most active university in the last decade having built and operating its own research machines. In addition to this, the Beijing Aeronautical Manufacturing and Technology Research Institute (BAMTRI) actively performs research on the process.

In Canada, the Advanced Manufacturing Technology Research Centre, which operates a research LFW machine, together with the University of Manitoba and the McGill University has produced a number of research papers in the last few years. In the USA, there are companies which design and built LFW machinery, like the Manufacturing Technology, Inc. (MTI) and

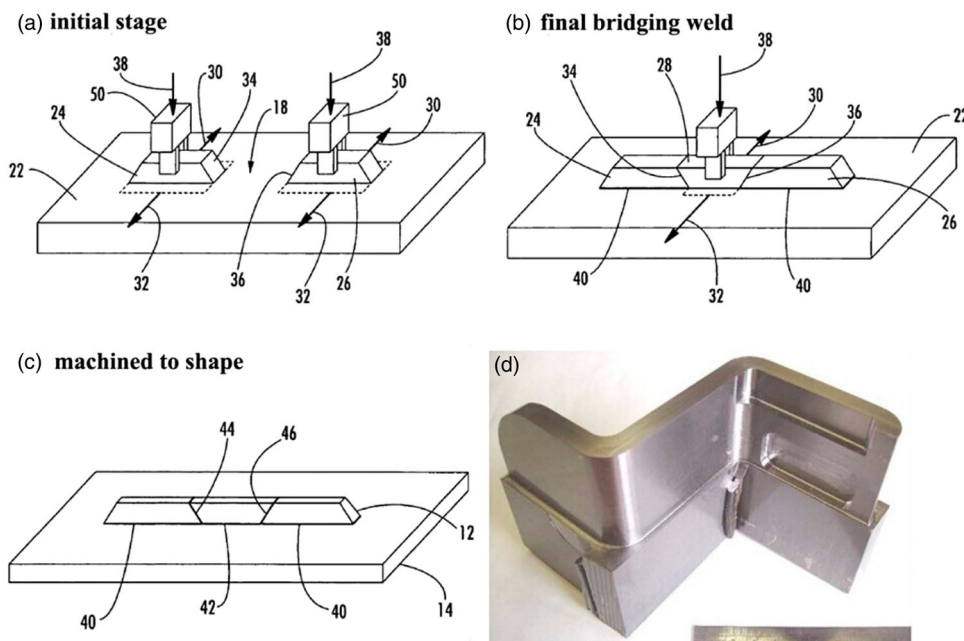
the ACPI Company (ACPI LLC). Industrial firms that employ this process include Pratt & Whitney and Boeing.

While the first report¹⁴ on the LFW process appeared in 1987, the first scientific publication in a peer-reviewed scientific journal was written by Koo and Bae-slack III¹⁵ only in 1992 focusing on the microstructure and mechanical properties of LFW'd 8009 aluminium alloy rods. In 1998, Vairis and Frost¹⁰ systematically reported on the LFW process with experiments performed with Ti64. However, the process is not limited to these applications and welding of other materials and dissimilar metal combinations have started to be explored (Fig. 3c).

Although the LFW process has been available for three decades, the majority of its commercial applications are in titanium alloy blisks for aerospace engines. The blisk arrangement provides considerable weight savings over conventional 'fir-tree' blade-disc assemblies.¹⁶ The high value-added cost of the blisk therefore justifies the initial cost of an LFW machine.¹⁷ In such cases, the most convenient method to repair blisks is also with LFW.¹⁸ Expanding the area of applications, a concept for near-net-shape component manufacture using LFW has been proposed by Boeing Company,¹⁹ where a number of structures can be assembled from simple shaped parts (Fig. 4). If machined from solid parts, a lot of machining would be required where near-net-shape forgings would be very expensive for most applications. LFW could produce material cost savings of up to 90% compared to traditional machining operations and low volume near-net-shape component manufacturing becomes the next promising application of LFW.²⁰

Advantages and disadvantages of RFW and LFW

Compared to conventional fusion welding methods, RFW and LFW have a number of distinct and important advantages.



4 Near-net-shape components manufactured with LFW a-c¹⁹ and an example of a preform d²⁰

- (i) the processes do not form a molten pool hence solidification defects (e.g. hot cracking, porosity, segregation, etc.) are reduced significantly²¹
- (ii) the processes are especially suitable to weld dissimilar metal joints, thus allowing the combination of different materials which can provide uniquely tailored solutions to engineering problems
- (iii) in the majority of joints, localised intense deformation and high temperatures result in recrystallised microstructures, producing an over-match in tensile properties to parent material (PM)
- (iv) there is no requirement in any of these processes for shielding gas or filler metals
- (v) the processes possess a self-cleaning ability due to wearing action of the relative movement of parts. The wear debris is continuously removed during the initial phases of the process by the relative movement itself into the flash
- (vi) the inherent reliability and repeatability of the processes are very good, which translates to minimal non-destructive testing time and expenses
- (vii) as all the process parameters are set before welding, operator errors are minimal
- (viii) furthermore, the process cycle is very fast, of the order of a few seconds, reducing a bottleneck from the production line²²
- (ix) finally, most RFW and LFW joints possess high integrity characteristics with the area of bonding being approximately 100% of the cross-sectional area of the joined parts.

Like other welding processes, RFW and LFW also have their limitations.

- (i) the two processes have different movement modes, with RFW being more suitable to join axisymmetric components, like tubes, shafts and discs, and LFW without this limitation
- (ii) welding of parts with large cross-sectional areas with CDFW is difficult due to the limited available power input of the spindle motor, making IFW the preferred option
- (iii) welding of thin-wall structures like tubes and plates with RFW and LFW, respectively, remains a challenge
- (iv) for the processes to weld successfully it is necessary for the material of at least one of the components to be plastically deformable. For example, difficult to weld alumina can be joined to aluminium, but cannot be joined to itself, as aluminium can deform more easily
- (v) compared to the more widely used RFW, LFW is limited by the high capital cost of its welding equipment, given the same welding area requirements. This is due to the fact that LFW machines need to be strongly built to contain parasitic vibrations, which originate from the oscillatory movement of large metal masses. Such design limitations do not allow the development of cheap and portable machines, but require large and expensive fixed installations
- (vi) the overall welding energy that can be produced during LFW is limited compared to RFW. Thus LFW has remained a niche manufacturing technology for the aircraft engine manufacturing
- (vii) as the processes form a flash which is an essential by-product, it limits the applications to the ones where the flash can be removed. If there are geometric constraints in the design then the FW processes cannot be used.

Process parameters

The appropriate process parameter selection governs the process development and its outcome in a different way for each FW variant. As FW is a process which depends on localised material deformation, heat generation plays a very important role and is discussed separately in this section.

Process parameters of RFW

As RFW is a well-established process there is a wealth of publications on it. Hence, only a brief description of its parameters is given in this review. The process parameters of CDFW and IFW are essentially the same with the exception of the mode of energy supply. In the case of IFW it is the flywheel moment of inertia which is available from the start of the run and is consumed through friction and plastic work during the process. It is therefore a simpler process with only three input parameters (flywheel moment of inertia, rotational speed and friction pressure) to control heat generation and release of the kinetic energy of the flywheel, and thus they are used as the process control variables. In the case of CDFW, energy input depends on the driving motor, which makes the friction time, axial shortening as additional process control variables besides rotational speed and friction pressure. Additional limitations of IFW are the fact that heat input rate cannot be controlled during the process as it relies on the flywheel to supply the necessary energy and large heat input requirements demand very large flywheels.

Process parameters of LFW

There are a large number of process parameters to control when welding with LFW (Fig. 2c),^{6,8-10} whose presence, sequence and value can have a significant impact on the process as well as on the finished joint properties. These are: (1) oscillation frequency, (2) oscillation amplitude, (3) friction pressure, (4) oscillation direction, (5) forging pressure, (6) friction time, (7) forging time, (8) welding time, (9) axial shortening (or burn-off or upset) and (10) axial shortening rate. Among them the first three parameters are the most important factors which influence the heat generation rate. The last three parameters in the list above are dependent on the first seven. Industry uses axial shortening as a marker for a successful joint and production of a part to specific dimensions.

Heat generation during RFW

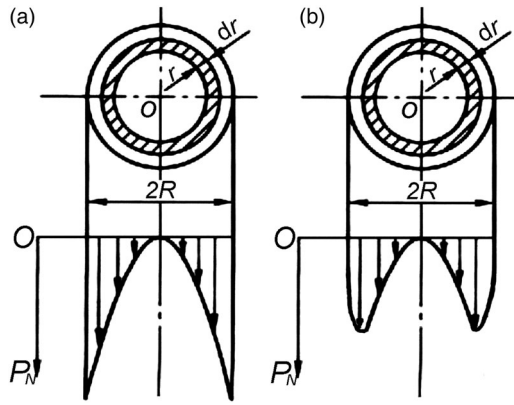
Heat generation during RFW is usually empirically represented as a frictional heating process. Given that axial pressure applied is P_N , $\omega(t)$ is the instantaneous rotational speed, the instantaneous heat generation rate per unit interfacial area at a radial distance r from the rotating axis can be expressed as:

$$dq = \mu P_N r \omega(t), \quad (1)$$

where μ is the friction coefficient. As the rotational speed ω for CDFW is constant, average heat input q for welding a rod is expressed as:

$$q = \frac{2}{3} \pi R^3 \mu P_N \omega. \quad (2)$$

According to equation (2), maximum heat generation will appear in the circumference of the interface as



5 Theoretical heat generation across the interface in RFW a^{23} and heat generation in experiment b

shown in Fig. 5a. However, in practice the applied friction pressure is not evenly distributed to the interface and is lower at the edges with the maximum heat generation being produced at somewhere between $1/2R$ and $2/3R$ (Fig. 5b), as experiments have shown.^{23,24} This effect reduces the theoretical unevenness of heat generation in RFW. In the case of IFW, rotational speed ω decreases to zero during the process nonlinearly due to friction.

In addition, for a more accurate representation of the process, friction coefficient should be expressed as a nonlinear function of temperature,²⁵ pressure²⁶ and/or of process parameters.²⁷ To this effect, the experimental power-travel method has been used to calculate heat input to the interface and ignore the friction coefficient estimation problem.²⁸ Another alternative method is to assume that friction stress (τ_f) is equal to the shear yield stress (τ_y) at the interface.²⁹ In such a case employing the von Mises yield stress, it can be expressed as:

$$\tau_f = \tau_y = \frac{\sigma_y}{\sqrt{3}}. \quad (3)$$

Hence the nominal friction coefficient can be:

$$\begin{aligned} \mu &= \frac{\sigma_y}{\sqrt{3}P_N} \text{ or } \mu \\ &= \frac{1}{\sqrt{3}} \text{ if one assumes that normal pressure equals yield stress.} \end{aligned} \quad (4)$$

Heat generation during LFW

The main difference between LFW and RFW is the mode of movement which affects heat generation and as a consequence the process phases as well. In one of the first papers published on LFW,⁸ a heat input model of the process was proposed with the heat generation rate per cycle per unit interfacial area q being:

$$\begin{aligned} q &= \tau_f v = \mu F_N v = \frac{\mu F_N}{A} a \omega \cos(\omega t) \\ &= \frac{\mu F_N a \omega \cos(\omega t)}{D[W - a \sin(\omega t)]}, \end{aligned} \quad (5)$$

where v is the sliding velocity, F_N is the normal force, a is the oscillation amplitude, ω is the angular frequency ($\omega = 2\pi f$), f is oscillation frequency, t is the time, D is the

thickness and W is the width of the specimen (weld area $A = D \cdot W$). By integrating the heat generation rate as expressed in equation (5), one can obtain the average heat input rate (q_0):

$$q_0 = \frac{2\mu F_N \omega}{\pi D} \ln \frac{W}{W-a} = \frac{4\mu F_N f}{D} \ln \frac{W}{W-a}. \quad (6)$$

Omitting the $a \sin(\omega t)$ term in equation (5), one can get the average heat generation:

$$\begin{aligned} q_0 &= \frac{1}{T} \int_0^T q dt = \frac{4}{T} \int_0^{T/4} q dt \\ &= \frac{4}{T} \int_0^{T/4} \mu P_N a \omega \cos(\omega t) dt = \frac{2\mu P_N a \omega}{\pi} \\ &= 4\mu P_N a f, \end{aligned} \quad (7)$$

where T is the oscillatory cycle period.⁸

The average heat generation rate depends mainly on friction pressure, amplitude and frequency of oscillation. Such a relationship would mean that an increase in frequency would allow the reduction of axial forces to weld delicate components, such as turbine blades, without buckling. This heat generation analytical model of the process does not take into account material properties such as strain-rate hardening effects.³⁰ In practice, the heat generation rate required to produce sound welds increases with the frequency of oscillation for strain hardening materials such as Ti alloys.³⁰

Similarly to RFW, the interfacial heat generated can be related to the shear yield stress, as described by the von Mises yield stress as shown in equation (3). Therefore, equation (7) can be rewritten as:

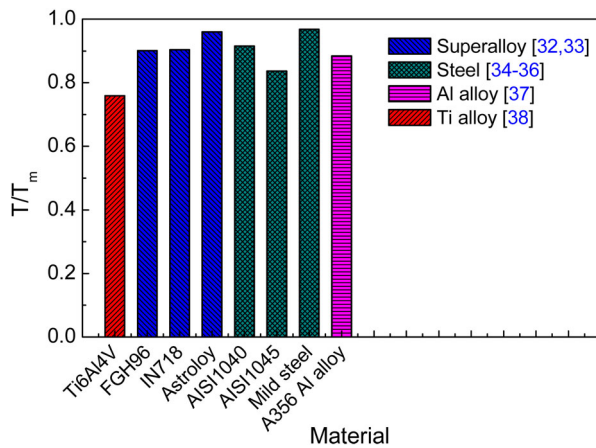
$$q_0 = \begin{cases} \frac{4P_N a f}{\sqrt{3}} & T \leq T_s \\ \frac{4\sigma_y a f}{\sqrt{3}} & T > T_s \end{cases}. \quad (8)$$

Here, T_s is the temperature when flow stress is equal to normal pressure. Heat input will initially depend on normal pressure, and once the interface temperature is high enough for the flow stress to be lower than the normal pressure, the heat input will depend on the temperature-dependent flow stress.

Interface temperature

In FW the interface temperature is possibly the most important factor which affects material deformation during welding and microstructure formed at the end. Having estimated the interface heat generation rate q , the temperature distribution for an one-dimensional case of RFW or LFW can be easily calculated.³¹

The actual maximum interface temperature reached in the steady-state phase of the process is quite high but below the T_m (see Fig. 6, between 0.75 and 0.95 of T_m (solidus)).³²⁻³⁸ Experiments show that the material type has little effect on the normalised interface temperature with the exception of Ti alloys. It is possible that a lower thermal diffusivity together with a pronounced decrease in strength at elevated temperatures are affecting the relatively low interface temperatures recorded for Ti alloys.



6 Normalised interface temperature during FW (T_m refers to solidus)

Effect of process parameters on the welding process

Effect of rotational speed in CDFW

It is expected that the intensity of heat generation is proportional to, and that the duration of the process is inversely proportional to, rotational speed. For example, it was reported by Li and Wang³⁹ that numerical analyses showed increasing rotational speed causes the weld interface to reach a quasi-stable temperature quicker, which starts material extrusion earlier and increases the upsetting rate remarkably.

Contrary to this, Vill²³ had found that when rotational speed is relatively high (from 800 to 3000 rpm), the welding process time is instead proportional to rotational speed. Upset was affected by the relationship of friction coefficient to speed. This difference could be related to material behaviour for different deformation and heating rates.

Effect of axial pressure in CDFW

In a similar fashion to rotational speed, axial pressure affects the temperature gradient, the required weld power and the burn-off rate. Numerical analyses by Li

and Wang³⁹ identified a proportional relationship between axial pressure with burn-off.

However, the selected rotational speed or axial pressure should be within an appropriate range for each material to be joined, and they should not be selected independent of each other. Their combined effects on the weld shape and width (Fig. 7) are common for all materials. It is usual in industry to select an average value for pressure and speed from the allowable window of process parameters, for a sound weld and small heat-affected zone (HAZ).¹²

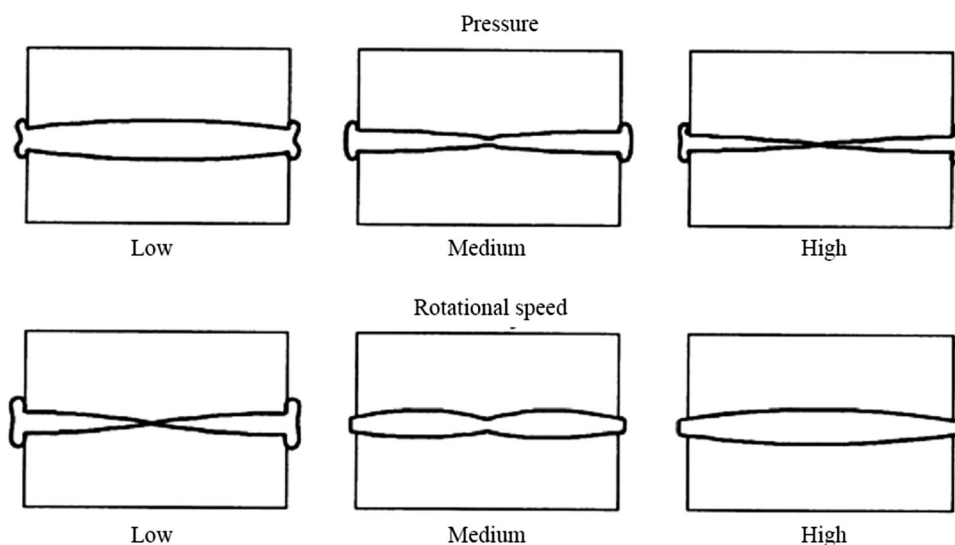
Effect of process parameters in IFW

The effects of welding pressure and rotational speed on the welds in IFW are similar to those in CDFW (Fig. 7). Ates *et al.*⁴⁰ reported that the HAZ width of MA956 alloy joints was proportional up to a certain value to friction pressure. Numerical modelling of Ni-based superalloy by Grant *et al.*⁴¹ showed an association of maximum interface temperature to welding pressure. But they pointed out that low friction pressure welds produce a wider high temperature region with flash extrusion being more difficult to achieve.

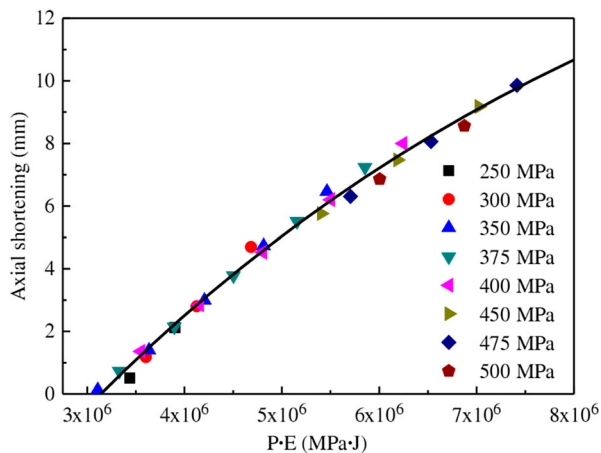
The selection of the appropriate process parameters enables this self-regulated process to reach the conditions for a sound weld with the required characteristics (e.g. strength and residual stresses). In this context, Wang *et al.*⁴² investigated the IFW of superalloys and proposed that mechanical energy depends on initial rotational speed, where high axial pressures increase the conversion efficiency from mechanical energy to welding heat. Axial shortening was found to be nonlinearly proportional to initial rotational speed and axial pressure (Fig. 8). The main parameter for a sound weld is the flywheel kinetic energy or axial pressure, which has to be above a limit to produce the necessary axial shortening.⁴²

Effect of process parameters in LFW

The analytical description of the LFW process (equation (7)) shows that increasing individual parameters affects heat generation at the weld interface. As reported by Wanjara and Jahazi,²² the welding time to produce a



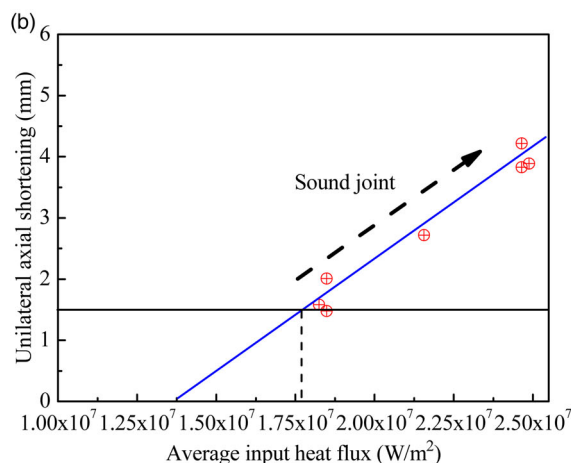
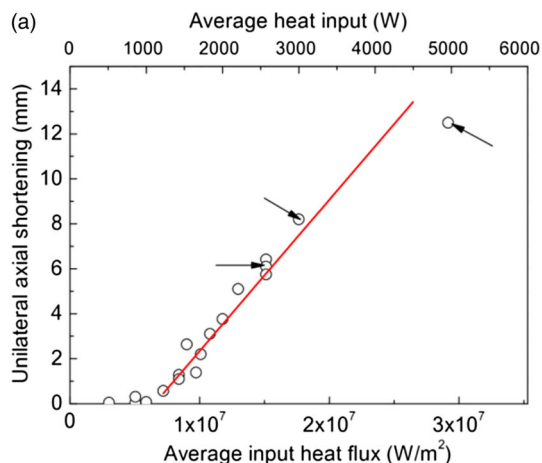
7 Effects of welding pressure and rotational speed on HAZ shape and width¹²



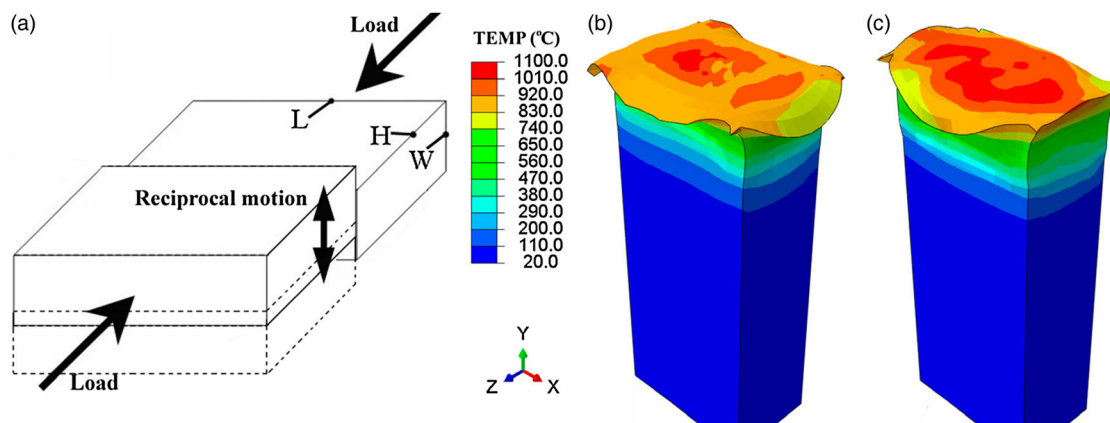
8 Combined effect of axial pressure (P) and kinetic energy (E) on axial shortening⁴²

joint having the equal shortening (2 mm) is inversely proportional to each process parameter and to power input. Similarly to RFW, there are critical process conditions for joining successfully Ti64, a power input larger than 2.4 kW and axial shortening at least 2 mm.²²

Using numerical analyses, Li *et al.*^{43,44} studied the relationship of average heat input to axial shortening



9 Relationship between average heat input and axial shortening for Ti64 *a*⁴³ and mild steel *b*⁴⁴ obtained by numerical modelling



10 Oscillation direction effect in LFW depend on aspect ratio of workpiece cross-section *a* compared to Fig. 2a. Estimated shapes and temperature contours of medium carbon steel at 4 s, friction pressure 70 MPa, amplitude 3.25 mm and frequency 35 Hz, *b* oscillation along X-axis and *c* along Z-axis

(Fig. 9). These work also identified a heat input limit for Ti64 (Fig. 9a) and mild steel (Fig. 9b) as experiments showed.^{10,45} Another process parameter, the welding pressure at high enough values affects material expulsion rate from the interface. Bhamji *et al.*²¹ found that it is proportional to burn-off rate, and inversely proportional to the required friction time once a limit has been passed.

Finally, there are two oscillation directions for LFW that can be used (Fig. 10a and b), which do have an effect on heat generation (see equation (6)) and material flow, especially for large aspect ratios. This geometric dependence affects the temperature distribution at the interface and flash shape in turn (Fig. 10c and d).

Insufficient heat input case

It may be difficult to provide the necessary energy during CDFW and LFW when the size of the weld cross-section is at the extreme, i.e. very large or very small. Vill²³ suggested the use of preheating the interface with gas burners. It has also been suggested to use Joule heating with electric resistance or high frequency currents during the process.²³

The purpose of preheating is to shorten the initial friction phase and immediately comes the extrusion phase. The use of electric current to assist CDFW of medium carbon steel⁴⁶ did result in lowering the required torque

and shortened the total process time. The produced joint had a more uniformly formed plasticised zone compared to conventional CDFW. This modification has been used to weld successfully thin valve steel rods 21-4N (austenitic stainless steel) and 4Cr9Si2 (martensitic stainless steel).⁴⁷ The use of electric current has also been applied to IFW for a Ni-based superalloy K418 and alloy steel 42CrMo dissimilar joint.⁴⁸

Effect of process parameters on joint strength

This section summarises the major published papers on joint strength while providing pointers to their use. Microstructure evolution and mechanical properties will be discussed in detail in sections 'Weld microstructure' and 'Properties of joints', respectively.

As CDFW has been studied for many years, only typical results are summarised. In the case of IFW and LFW papers are fewer. From the analysis of the reported results (for examples, Figs. 11–14), it can be concluded that:

- (i) for similar material combinations, joints have a relatively high tensile strength, usually comparable to or higher than that of the parent metal. Failure usually occurs in the parent metal or within the HAZ or the thermomechanically affected zone (TMAZ) depending on the material type. Failure at the interface indicates wrong selection of welding parameters or inadequate heat input (e.g. superalloys). The allowable process parameters have a range of values from which to choose with little effect on joint strength provided the power input exceeds the minimum limit. In the case of materials undergoing phase transformation or significant grain coarsening at high temperatures, like carbon steels, friction time is of importance. Longer friction times increase heat input at the expense of joint strength
- (ii) for dissimilar material combinations, even for materials with large differences in their thermal and mechanical properties, or even for those which form intermetallics, the joints could reach a comparable strength to fusion welds. Fracture may occur at the softer material side of the joint, or in the HAZ, or at the interface when intermetallics have formed. The process parameters usually have limited influence on joint strength provided they are within the appropriate range

(iii) for CDFW, as it is usually the case, tensile strength is proportional to friction pressure up to a maximum value, and decreases with further increase of pressure.^{40,50,51} When friction pressure is: (i) low, leaves unbonded regions at the interface (ii) at the appropriate level, material at the interface deforms and yields to form a sound joint and (iii) excessive, leads to excess frictional heat resulting in grain coarsening and reduction in weld strength. The reason for the reported contradictory results where the strength of joints increases with friction pressure,⁵² while it is expected to be inversely proportional, is due to the different optimal ranges of friction pressure for each material^{53,54} (iv) titanium alloys²² and steels²¹ can be readily welded by LFW and the tensile strength of the joints is less sensitive to process parameters. However, superalloys⁵⁵ need careful selection of parameters to satisfy the passmark of axial shortening, or the formation of a continuous flash from all corners of the interface.

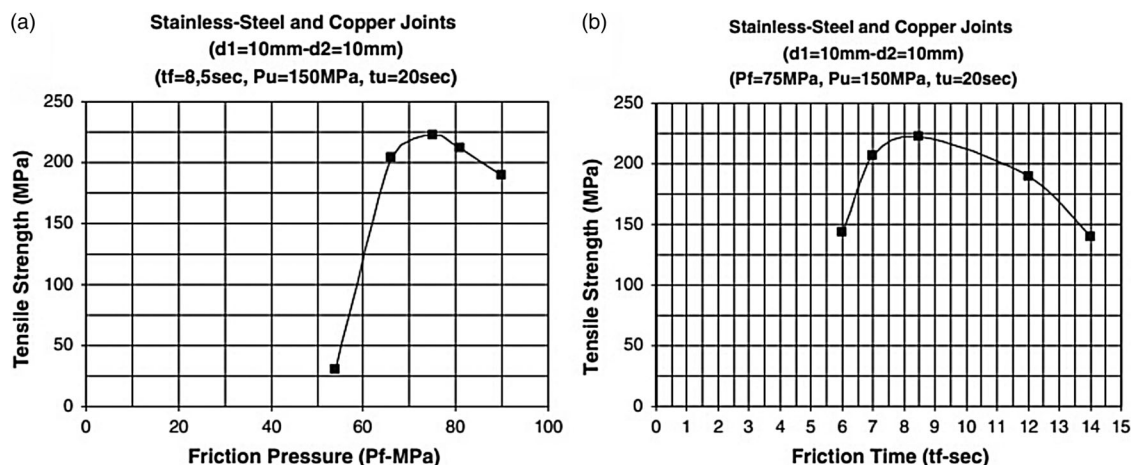
However, all published works have focused on a limited range of welding parameters and materials, without producing comprehensive guidelines on a wide range of materials.

Weld microstructure

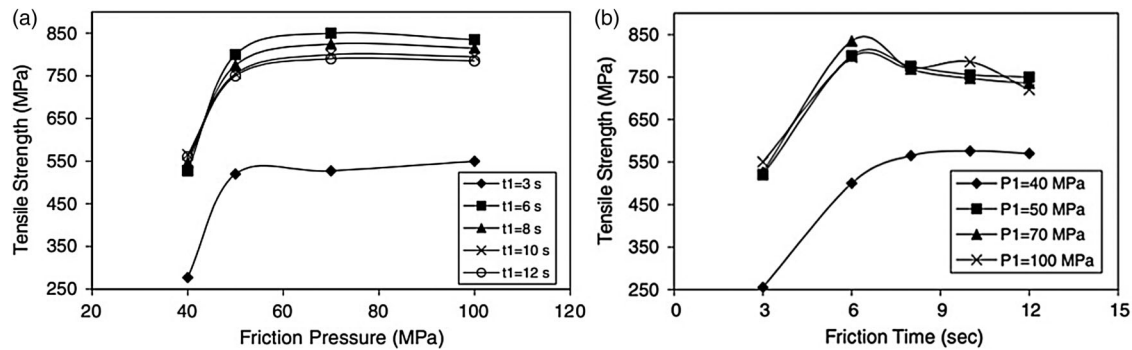
The focused and intense mode of heat generation in combination with dramatic cooling rate and large material flow leads to weld microstructures which are very different to conventional fusion weld microstructures. Recrystallisation, solid-state transformations, dissolution, reprecipitation and grain growth may take place during the intense thermo-mechanical coupling conditions of FW. All these phenomena will affect interfacial bonding and joint properties.

Macroscopic features of joints

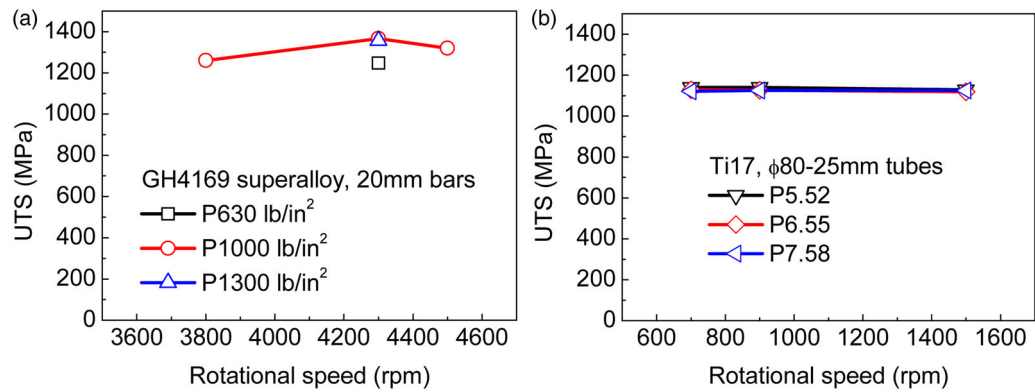
The typical macroscopic features of RFW'd joints (Fig. 15) are a separated curled flash on both sides of the joint due to the simultaneous rotation and extrusion action. The shapes of certain LFW joints show unique extrusion features, especially in the case of titanium alloys (Fig. 16a) and steels (Fig. 16b). This feature is highly



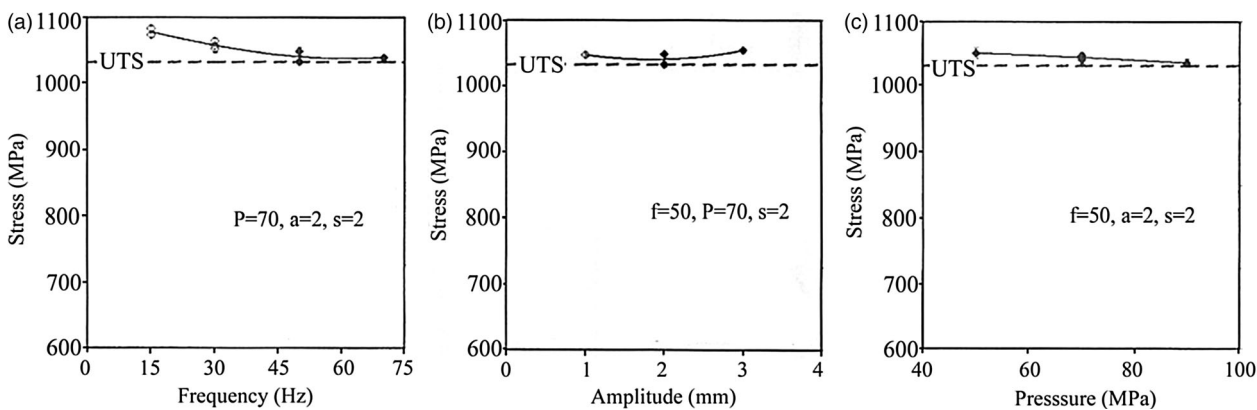
11 Effect of friction pressure *a* and friction time *b* on tensile strength of Cu to stainless steel dissimilar joints by CDFW in the as-welded state⁴⁹



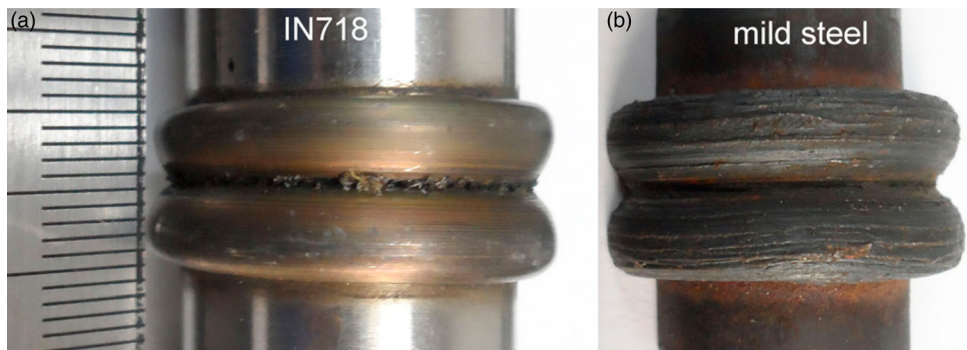
12 Effect of friction pressure *a* and friction time *b* on the tensile strength of CDFW'd MA956 Fe-based superalloy under constant forging pressure ($P_2 = 70$ MPa) in the as-welded state⁴⁰



13 Effect of rotational speed on the UTS of IFW'd *a* GH4169 [⁵⁶] and *b* Ti17 [⁵⁷] joints in the as-welded state



14 Tensile strength of Ti64 joints in the as-welded state in relation to LFW process parameters: *a* frequency of oscillation, *b* amplitude of oscillation and *c* frictional pressure. The open and filled markers denote the nonwelded and welded conditions, respectively²²



15 Typical appearance of RFW'd joints: *a* IFW of IN718 and *b* CDFW of mild steel

beneficial to the joint strength while avoiding root defects, like oxidation or partially bonded welds.

However, other materials like Al alloys (Fig. 16c) and superalloys (Fig. 16d) produce flash with shapes similar to RFW. Such flash formations may have root defects like cracks or oxide inclusions, as the oscillations of the specimen expose periodically parts of the interface to air. This issue is more pronounced in superalloys because of their high temperature mechanical properties and the ease by which they form compact oxide films. The underlying reasons for these differences in flash shape are still not understood. Although Chamanfar *et al.*⁵⁷ found that the integrity of an LFW'd WASPALOY superalloy joint could be improved by adjusting process parameters, the production of a 'well-formed' or 'continuous' flash in the case of LFW superalloys still remains a challenge. There might be certain microstructural features which contribute to the low strength of joints, as is discussed in the next section.

Texture formation in welds

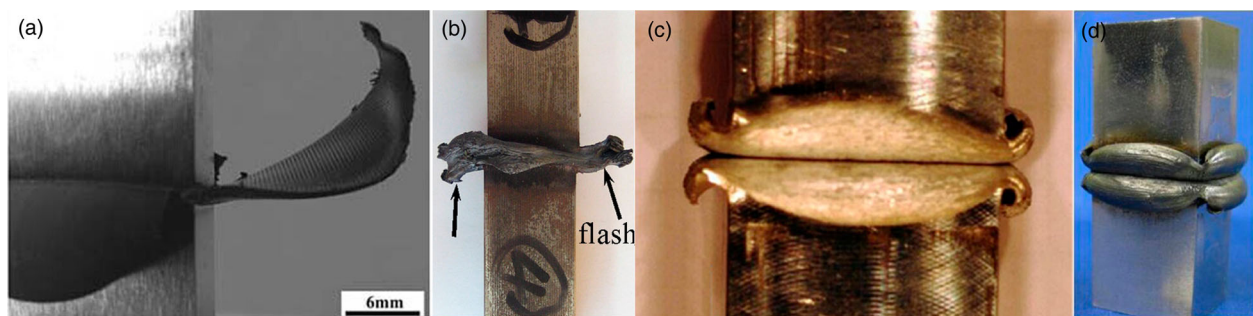
Microtexture is a common phenomenon in largely deformed metal parts, especially if they have a hexagonal close-packed (hcp) crystal structure. However reports on texture development in RFW or LFW are limited in numbers.

In the Ph.D. thesis of Dalgaard,⁵⁸ electron backscattered diffraction (EBSD) analysis of LFW'd Ti64 joints (Fig. 17)

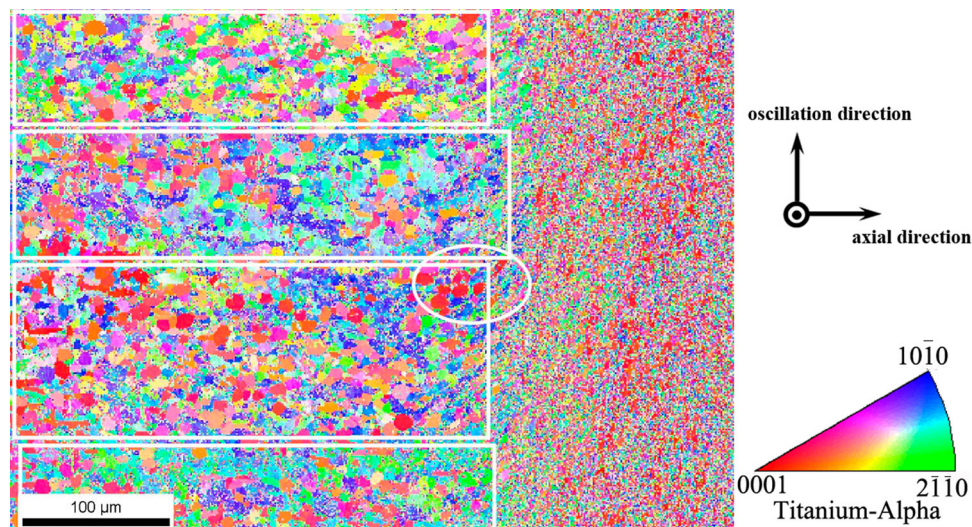
showed an α texture, a typical transverse one, in the weld centre zone (WCZ). Karadge *et al.*⁵⁹ investigated the LFW'd Ti64 and the post-weld heat treatment (PWHT) effect for lab (small) and full (large) scale specimens. An extremely strong transverse texture was found at the weld line irrespective of the specimen size (Fig. 18),⁵⁹ which texture size depends on the size of the specimen. In the large specimens alternating bands of transverse $\{11\bar{2}2\}<1123>$ type textures were observed at a small distance away from the weld line (Fig. 19). Both sized specimens showed weak basal and R1 type textures. PWHT did not affect the texture. Results from the same group⁶⁰ with the same material have suggested an association between welding parameters and texture development, where low friction pressures produce strong transverse α textures at the weld line, and high friction pressures an almost random α texture.

Attallah *et al.*⁶¹ studied IFW'd Ti-6246 alloy using EBSD mapping for both the as-welded and the PWHT conditions. The strongest textures were observed in the flash extrusion direction, i.e. the direction of largest deformation. A significant α -Ti variant selection occurred during the cooling and PWHT stages, producing a strong $\{0002\}_\alpha<11\bar{2}0>$ texture in the radial direction in all finished conditions.

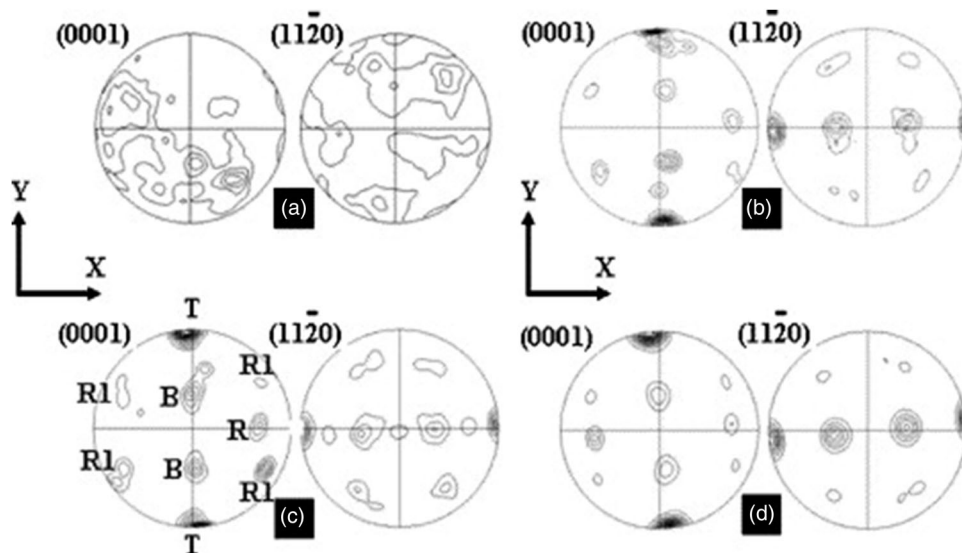
More recently, Dalgaard *et al.*⁶² studied the LFW'd near- β titanium alloy Ti-5553 (Fig. 20). The large equiaxed and deformed grains are predominantly oriented with their $<111>$ directions normal to the



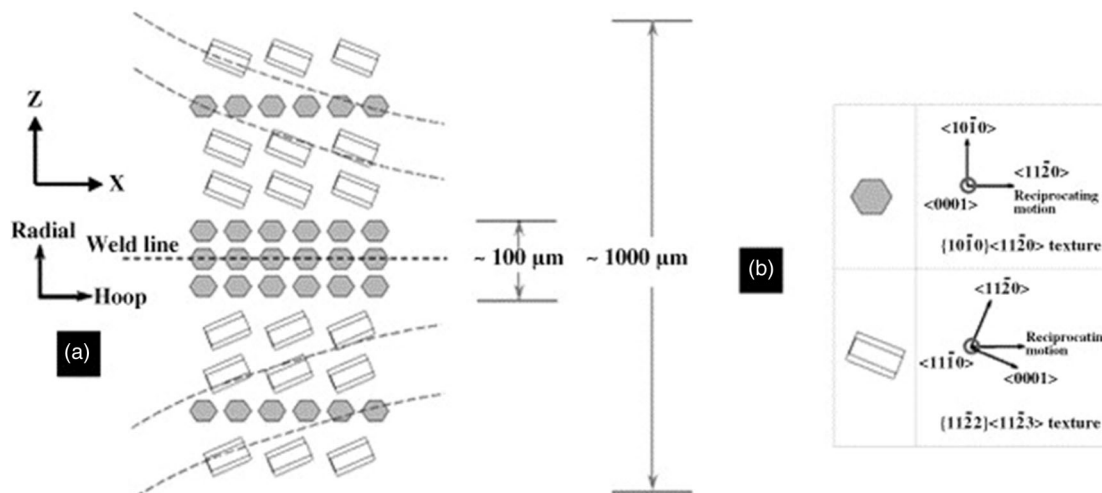
16 Photos of LFW joints: a Ti-6.5Al-1.5Zr-3.5Mo-0.3Si (Ti11),⁵⁵ b medium carbon steel,⁴⁵ c AA2024⁵⁸ and d WASPALOY⁵⁷



17 EBSD inverse pole figure map of α phase in a LFW'd Ti64 joint in the as-welded state; an image quality map has been overlaid to make grain boundaries more visible⁵⁸



18 $\{10\bar{1}0\}\langle 11\bar{2}0\rangle$ α textures in lab-scale specimens: **a** base alloy (4X random), **b** as-welded (transverse texture (T): 35X, rolling texture (R): 2X, basal texture (B): 5X, R1 type texture: 2X), **c** 1 h PWHT (T: 36X, R: 3X, B: 7X, R1: 5X) and **d** 8 h PWHT (T: 37X, R: 3X, B: 2X and R1: 1X) (note that XY is the weld plane)⁵⁹



19 Crystallite orientation at, and near, the weld line in full-scale specimen (XY: weld plane) **a**, and the texture **b**⁵⁹

sample surface, which is the weld oscillation direction as the primary slip system in this phase. When welding at a lower frequency of oscillation and higher friction pressure, the width of the recrystallised zone increases by 50%, with grain orientation remaining unchanged.

When welding pure Ti,⁶³ the WCZ consisted of pure α phase with an evident texture (Fig. 21). The elongated grains present in the weld indicated incomplete recrystallization and the PWHT did not affect the texture of the weld.

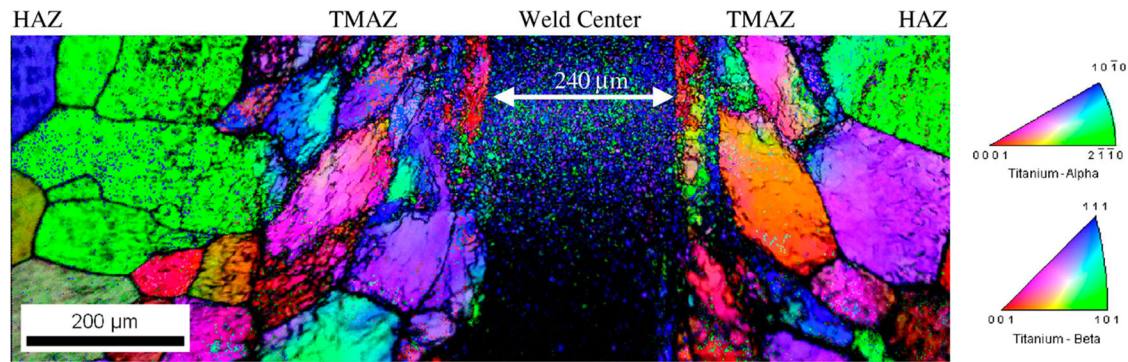
Microstructure of joints

General features

Both of these processes can be effectively viewed as processes where material is forged locally under very high strain rates at high temperatures if assuming that specimens do not rotate or oscillate. As a consequence, dynamic recrystallization (DRX) is a common feature

across all types of materials. A typical RFW'd or LFW'd joint consists of three distinct areas: (1) WCZ where metallurgical phase transitions occur and it is characterised by fine recrystallised grains, (2) TMAZ which is exposed to lower temperatures and strains, and characterised by deformed and/or partly recrystallised grains and (3) HAZ of limited thermal exposure. The different temperature and strain histories of these zones affect the final microstructures.

The microstructure characteristics of RFW'd joints have been reviewed appropriately by Uday *et al.*⁶⁴ This section therefore, besides LFW'd joints, reviews additional recent information related to RFW of dissimilar steel-aluminium and superalloy joints. The microstructural phenomena encountered are usually recrystallisation, dissolution and reprecipitation, solid-state phase transformation, formation of interface intermetallics. However, for certain materials, there may be more than a single phenomenon involved.

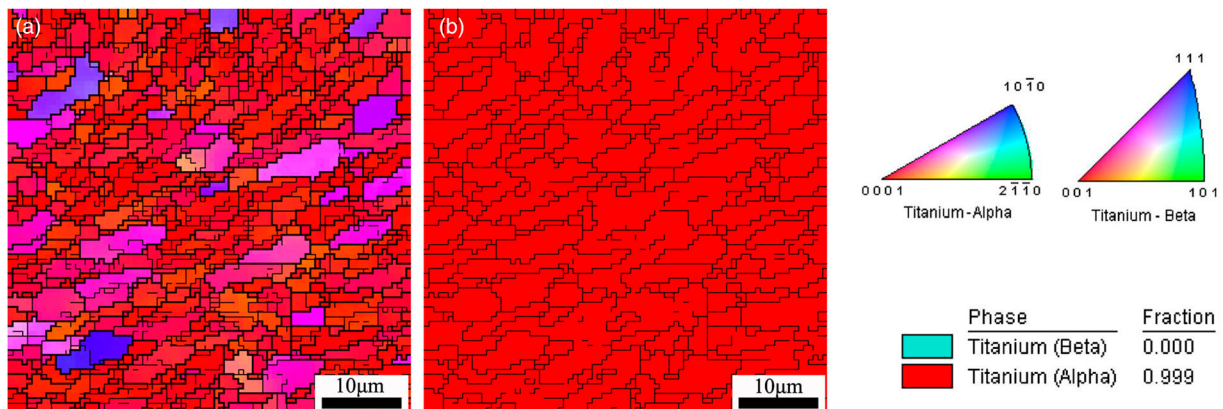


20 Image quality combined with inverse pole figure of LFW'd near- β titanium alloy Ti-5553⁶²

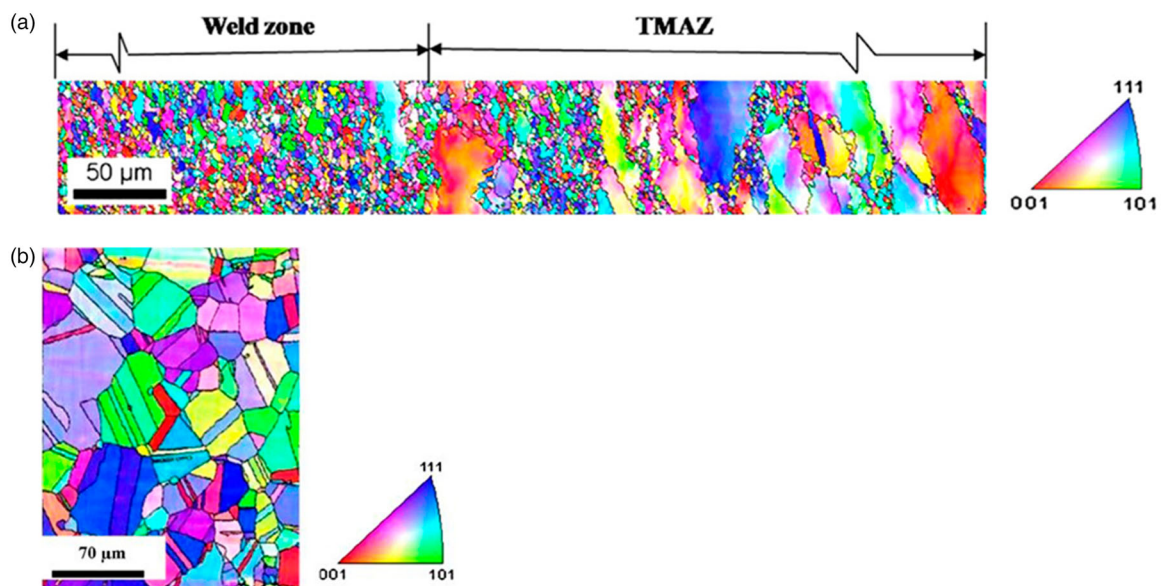
Superalloys

In superalloys, dissolution of the strengthening phase across the joint takes place. In Ni-based superalloys, DRX is usually observed in the region where intergranular and intragranular γ' have been dissolved during the process as they hinder DRX.

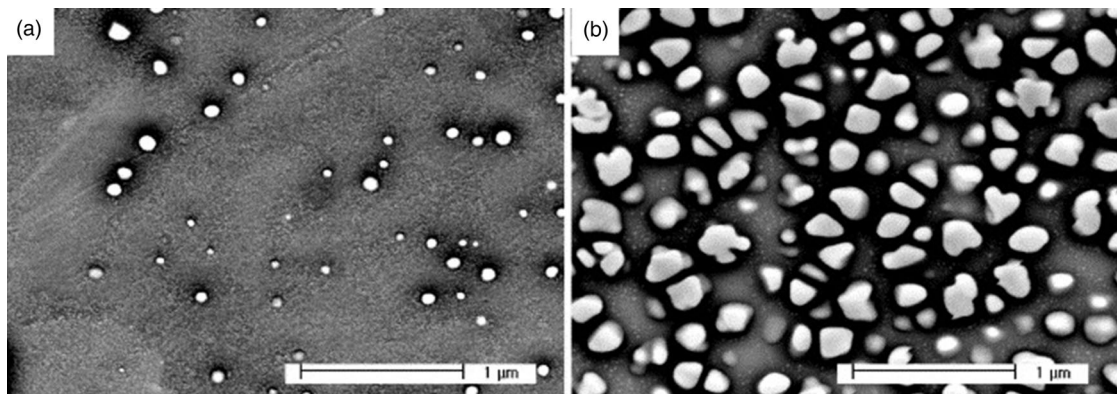
The WCZ of CDFW'd IN718 consists of extremely fine grains due to the full dynamic recrystallisation (Fig. 22a).⁶⁵ Severe dissolution of the strengthening precipitates in WCZ, TMAZ and HAZ causing a remarkable decrease in hardness in these zones, and especially in HAZ. The subsequent PWHT increased and evened out hardness (Fig. 22b).⁶⁵



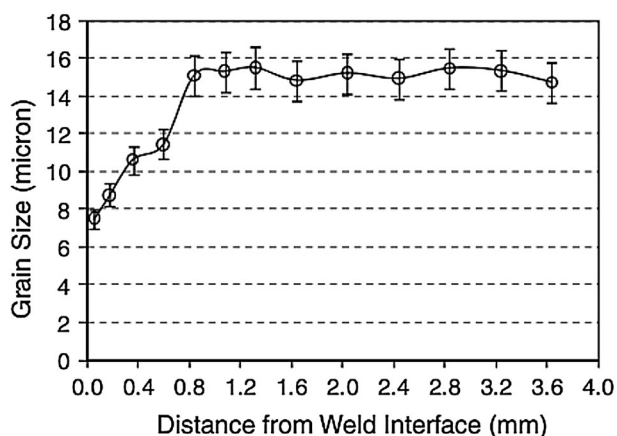
21 Inverse pole figure of the WCZ of LFW'd pure Ti a and its phase map b⁶³



22 EBSD (inverse pole figure (IPF) + grain boundary) map of a CDFW'd IN718 sample in the as-welded state (with STA (solution treatment followed by aging) as the pre-weld heat treatment) a, and EBSD (IPF + grain boundary) map of WCZ in the PWHTed state (with pre- and post-weld STA treatment) b⁶⁵



23 SEM micrographs of *a* the region within 40 μm of the weld line and *b* the parent region 1.5 mm away from the weld line of a RR1000 inertia weld in the as-welded state⁶⁷



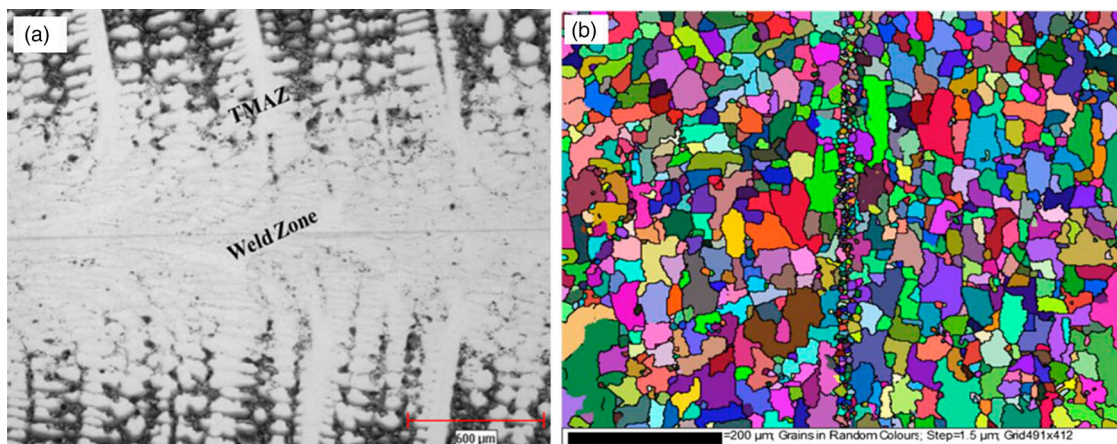
24 Grain size distribution of an LFW'd WASPALOY sample in the as-welded state⁶⁹

There have been interesting papers on the microstructures of IFW'd high volume γ' Ni-based superalloys. IFW of such materials produces extremely narrow WCZs. Huang *et al.*⁶⁶ reported the smallest WCZ width of 220 μm for the fine grain RR1000. The weld temperature was high enough to dissolve all parent γ' particles (Fig. 23).⁶⁷ During and after welding the plasticised WCZ would also be fully recrystallised and fine-sized

particles may have reprecipitated during the very fast cooling stage. Preuss *et al.*⁶⁸ also studied the same alloy and found that the driving force for reprecipitation of γ' was greater at the weld line (all γ' dissolved during welding) than about 1 mm away from it (primary and some secondary γ' still remaining) resulting in the greatest γ' depletion for the as-welded condition at about 1 mm from the weld line. Accordingly, hardness trough was observed at this position. In addition, the finest γ grain size was found at about 0.5 mm from the weld line, in a region where recrystallisation had just started but very little grain coarsening.

Chamanfar *et al.*^{57,69} investigated the microstructure of LFW'd WASPALOY joints for different process parameters. At 0.9 mm away from the weld interface, γ grain size remained almost constant, while closer it decreased (Fig. 24).⁶⁹ As γ' is the main strengthening phase for WASPALOY, Chamanfar *et al.*⁵⁷ found that it completely dissolved at the interface demonstrating that the supersolvus temperature has been exceeded. The subsequent PWHT reprecipitated γ' in the WCZ.

In the case of single crystal superalloy welds, recrystallisation is clearly evident. Ola *et al.*⁷⁰ showed that the LFW of a single crystal superalloy CMSX-486 had two distinct microstructural regions across the weld, the WCZ and TMAZ (Fig. 25). In the WCZ the γ' precipitates had completely dissolved and the MC carbides



25 Optical micrograph of the LFW'd CMSX-486 joint in the as-welded state *a* and EBSD-based orientation map showing the recrystallised grains in random colours across the weldline *b*⁷⁰

exhibited partial dissolution. In the TMAZ, however, only the Hf-rich Ni-based phase had completely dissolved, with the other second-phase particles exhibiting limited dissolution. Karadge *et al.*⁷¹ welded single crystal Ni-based superalloy (CMSX-4) to a polycrystalline Ni-based superalloy (RR1000) with LFW. There was significant crystallographic shear next to the weld line along with submicrometer-sized recrystallised grains and small precipitates. On the single crystal side, some of those grains recrystallised and a significant volume fraction of the γ' precipitates had dissolved.

Titanium alloys

Unlike superalloys, titanium alloys undergo an allotropic phase transformation during FW generating complex microstructures in particularly in the case of typical two-phase Ti alloys. As reported, for the $\alpha+\beta$ Ti64 alloy, the change in microstructure is a result of exceeding the β -transus temperature of 995°C during welding.⁷² Detailed microstructure analysis has in some cases revealed severe β grain size refinement at the weld line and as a result of the high cooling rates very fine, often blocky type of α lath structure within the refined β grains. In some cases a single α variant within each β grain was found resulting in very severe texture at the very weld line.⁶¹ However, the precise nature of the variant selection and texture evolution remains elusive.

Wanjara and Jahazi²² studied the effect of welding parameters on the recrystallised β grain size, to find them proportional to the frequency of oscillation up to 50 Hz, beyond which it remained unaltered. Increasing friction pressure produced similar behaviour with no upper boundary, but the opposite occurred with the amplitude of oscillation. Although amplitude and frequency of oscillation affect frictional heat generation in a similar fashion, this difference may be due to the effect of increasing exposure of interface areas to the environment during each cycle, which increases convective heat losses.

As has been previously shown (Fig. 20), recrystallisation in near- β titanium alloys is more pronounced than in the two-phase Ti alloys. In that case most β phase was retained in the weld zone.⁶²

As for the LFW of pure α -Ti (Fig. 21),⁶³ no β phase was detected in the weld zone while there was incomplete recrystallisation of α -Ti evident from elongated α grains. Although the interface temperature exceeded β -transus and the α grains in WCZ were refined, the incomplete recrystallisation may be attributed to the relatively high stacking fault energy (SFE) of α -Ti.

Steels

Steels are of similar complexity as titanium alloys because they both undergo allotropic phase transformations and their microstructure changes. Compared to Ni-based superalloys and titanium alloys, steels have large thermal conductivity and hence produce a wider HAZ.

Ma *et al.*⁷³ studied in detail the LFW'd medium carbon steel (Fig. 26). Fine α +P structures form during the recrystallisation of deformed grains at the weld interface, which correspond to a normalising temperature higher than 800°C.⁷³ As for the LFW'd austenitic stainless steel (AISI316L),²¹ fine grains were also observed in a large thickness area close to the weld line (Fig. 27a). A

considerable amount of δ -ferrite appeared close to the weld line as shown in Fig. 27b, whose amount was strongly inversely proportional to the burn-off rate.

Interface reactions in dissimilar joints

A solid-state process like FW is suitable for joining dissimilar alloy combinations, especially those that form intermetallic compounds (IMCs), e.g. Fe-Al, Ti-Al, Fe-Ti, Fe-Cu and Al-Cu systems. However, when IMCs are formed, a significant reduction in fracture strength is observed. Mechanical intermixing is usually observed at the interface. In addition, the different recrystallisation and phase transformation behaviour of dissimilar material combinations will also affect interface bonding.

Joints between aluminium and steel usually result in multiple IMCs as a result of a solid-state reaction below the eutectic temperature (652°C) at the interface.⁷⁴ In traditional fusion welding, the IMCs of Fe-rich intermetallic phases (e.g. FeAl, Fe₃Al) appear at temperatures higher than 1200°C and have improved ductility compared to brittle Al-rich phases (e.g. Fe₂Al₅, FeAl₃, Fe₄Al₁₃). The mode of formation and distribution of IMCs in Fe-Al joints have been studied in a large number of papers.

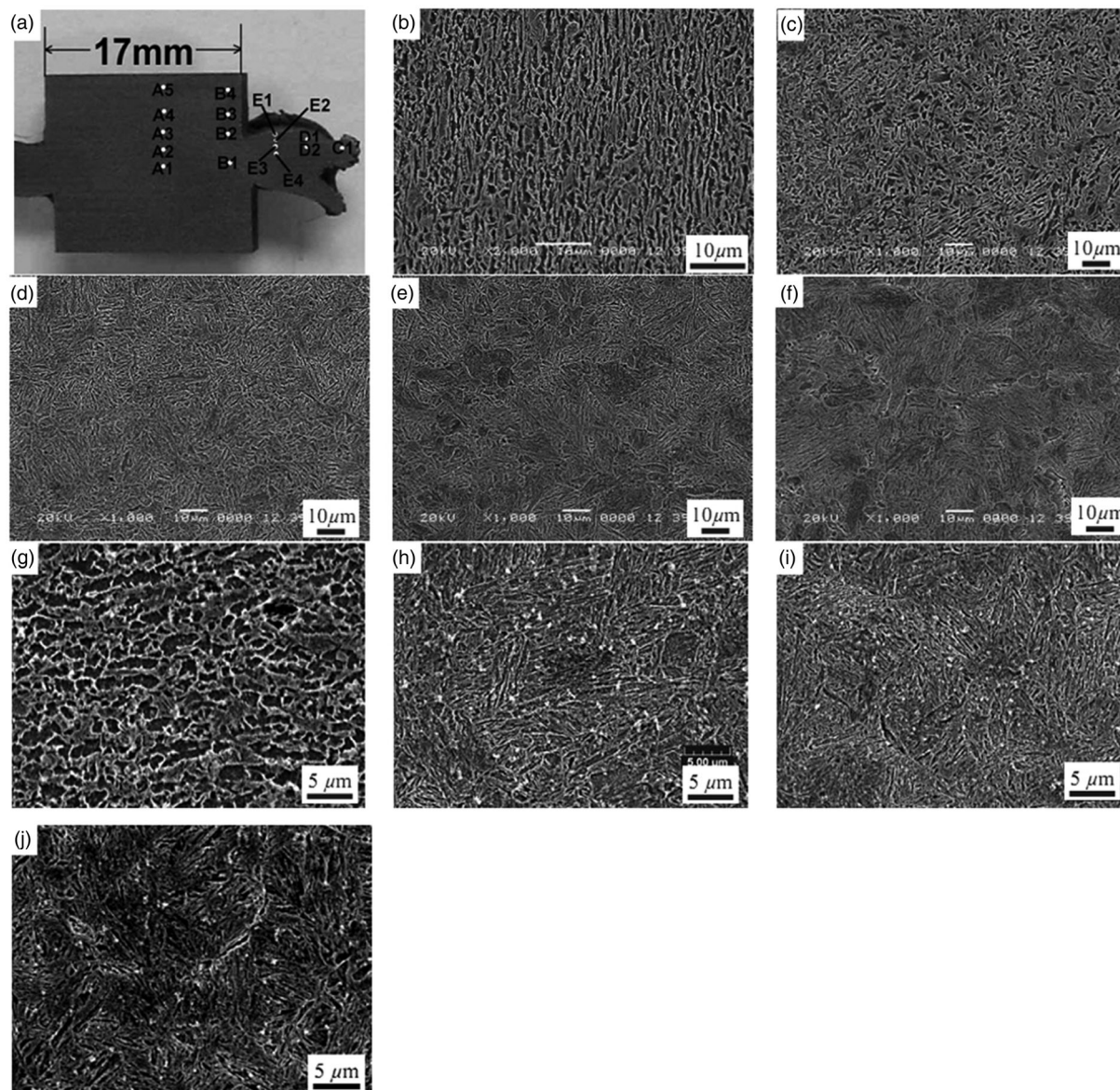
Fukumoto *et al.*⁷⁴ studied the mechanical mixture of dissimilar metals during the frictional stage and the reaction products in CDFW'd 304 stainless steel and 5052 aluminium alloy. A 300 nm thick reaction layer of FeAl and Fe₃Al phases was found between stainless steel and aluminium alloy (Fig. 28a). Taban *et al.*⁷⁵ found the FeAl and Fe₂Al₅ phases at the interface in CDFW'd 6061-T6 aluminium alloy to AISI1018 steel (Fig. 28b).

Yilmaz *et al.*⁷⁶ showed that the thickness of the intermetallic layer of Al/steel welds depends linearly on the square root of the friction time, indicating that its growth is related to diffusion. In order to reduce oxidation of welding, an Ar atmosphere was used, but did not completely eliminated oxides. Therefore, an interlayer phase can only form where chromium oxide does break away from the interface.

Ashfaq *et al.*⁷⁷ modified the geometry of the rubbing surface of a steel part by tapering its ends to reduce the IMCs remaining at the interface for a CDFW'd stainless steel AISI304 to AA6061 joint. It was found that the external tapered samples showed better strength than flat or internal tapered samples.

In order to improve the strength of Al/Fe joints, Reddy *et al.*⁷⁸ used electroplated Cu, Ni and Ag interlayers to CDFW'd AISI 304 stainless steel and AA6061 aluminium alloy. The Cu and Ni interlayers produced brittle welds as CuAl₂ and NiAl₃ formed, while Ag acted as an effective diffusion barrier for Fe. When the Ag interlayer was used welds became remarkably stronger and ductile.

The joints of other reactive combinations showed similar interface behaviour. More specifically Bhamji *et al.*⁷⁹ and Wanjara *et al.*^{80,81} studied LFW'd dissimilar Al/Cu joints, which had no voids or cracks (Fig. 29a),⁸⁰ but bond integrity was significantly dependent on welding parameters. At the interface zone on the Al side, trapped Cu debris were identified. In addition a CuAl₂ intermetallic was scattered across the interface. Wanjara *et al.*⁸¹ also investigated the effect of PWHT on IMCs of Cu₂Al and CuAl₂. Bhamji *et al.*⁸² also found that the interface of LFW'd dissimilar AA6082-T6 aluminium alloy/AZ31 magnesium alloy showed intermixing between the two materials (Fig. 29b).



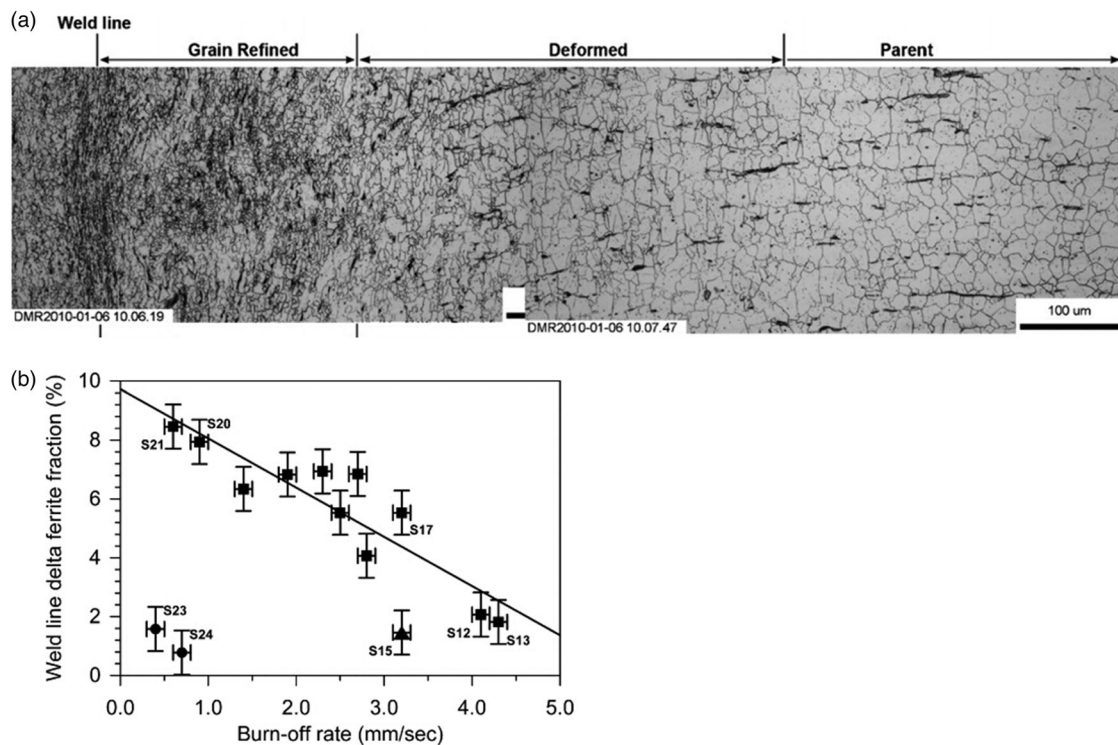
26 SEM microstructures of different zones in the medium carbon steel joint in the as-welded state as marked in a. b–j correspond respectively to zones A1–A5 and B1–B4 marked in a. b fine ferrite (α)+pearlite (P) in the WCZ, c and g deformed fine α +P in the TMAZ, d and h tempered sorbite (TS) in the HAZ, e and i tempered troosite (TT) in the HAZ, f and j tempered martensite (TM) in the HAZ⁷³

For certain dissimilar material combinations, although no evident interface diffusion and reaction could be found, material extrusion was different for their widely different thermal and mechanical properties which affect joint formation and bonding. Ma *et al.*⁸³ examined the weld microstructure of LFW'd dissimilar titanium alloys Ti64 and Ti11 (Fig. 30a). The Ti64 side showed the expected microstructure. On the Ti11 side some α precipitated along β boundary because of insufficient recrystallisation, and the bond consisted of recrystallised grains. In the LFW'd dissimilar Ti64/Ti6246 joint (Fig. 30b),⁸⁴ the WCZ microstructure was characterised by two sets of fine acicular-shaped features perpendicular to each other. The distribution of alloying elements was asymmetrical with no long-range interdiffusion across the interface.

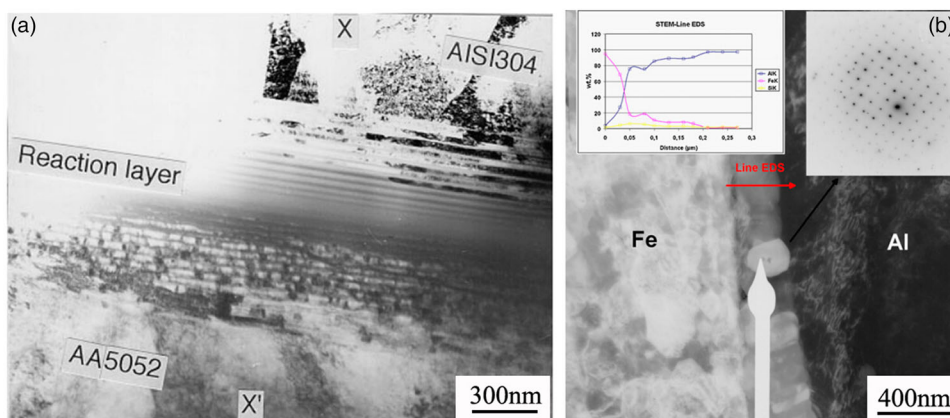
Abnormal phase formation

In most cases, when the appropriate process parameters are selected the weld zone has a microstructure with recrystallised grains and good bonding. However, when sufficient extrusion of the interface material cannot be achieved, defects may appear at the interface.

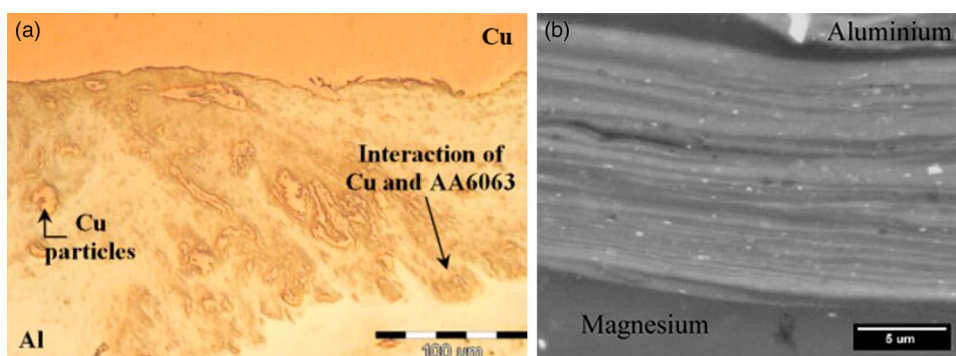
Li *et al.*⁸⁵ found an interesting microstructure phenomenon in the interface of certain LFW'd Ti64 welds, an abnormal spherical α phase (Fig. 31). While macroscopically, the weld line appears like a crack, of over 60% of the whole weld interface (Fig. 31a). At high magnification (Fig. 31b), these spherical α grains and the surrounding α' martensite and β matrix are present with almost no pores. These grains are responsible for the decrease of the joint tensile strength, and they remained even after PWHT (Fig. 31c). More recently, the same phenomenon



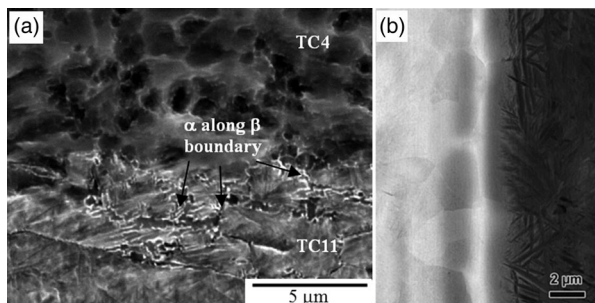
27 Typical microstructure of a LFW'd 316L joint in the as-welded state *a* and variation of weld line δ -ferrite fraction with burn-off rate *b*²¹



28 TEM images of the CDFW'd 304 stainless steel/AA5052 weld interface *a*⁷⁴ and 6061-T6 aluminium alloy to AISI1018 steel weld interface (embedded line energy dispersive spectrometer (EDS) scan across the interface and selected area diffraction pattern of IMC) *b*⁷⁵



29 Interface of the LFW'd Al/Cu *a*⁸⁰ and Al/Mg *b*⁸² joints



30 Interface of LFW'd Ti64 (TC4)/Ti11 (TC11) ^a and Ti64/Ti-6246 ^b joints

was observed for LFW'd Ti11.⁸⁶ The mechanism proposed for this formation is similar to hot working, where some lamellar α phase is broken up by boundary splitting and spheroidal objects similar to small balls are formed. Subsequently, the recrystallisation of the segments takes place and α lamellae transform into equiaxed structures. Under the applied friction forces, small α particles can coalesce to form bigger ones by squeezing out the softer intergranular β phase.

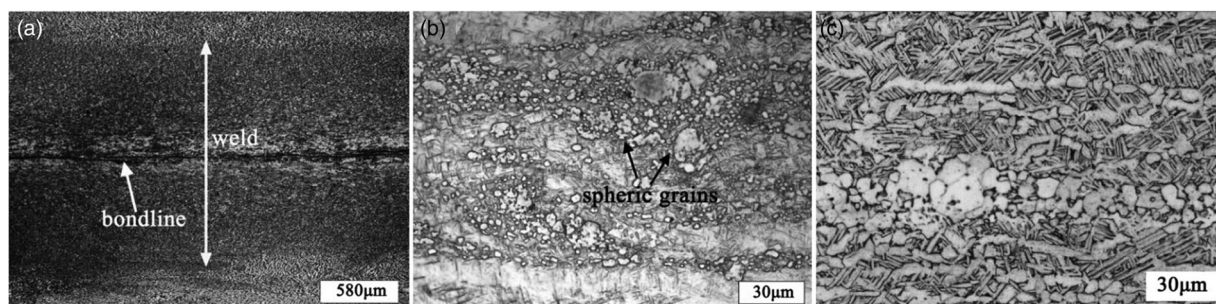
Local melting

The FW processes are characterised as 'solid-state processes', but under certain conditions there may be intergranular liquation for some materials, which means the definition of 'solid state' is not entirely true. In the work

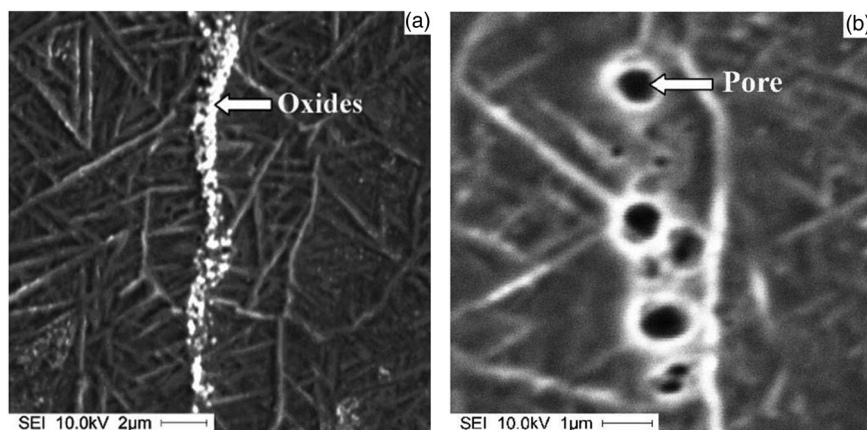
of Ola *et al.*,^{70,87} possible liquation of various phases during LFW of superalloys, i.e. CMSX-486 single crystal superalloy⁷⁰ and IN738 [89], was investigated using the Gleeble thermomechanical simulation of the process to identify local melting of the γ - γ' eutectic constituents. Significant grain boundary liquation was observed in both the Gleeble specimens and the HAZ of actual joints due to non-equilibrium reaction of second-phase particles, including the strengthening γ' phase. However, no HAZ cracking was observed, because of friction pressure during LFW.

Oxide inclusions

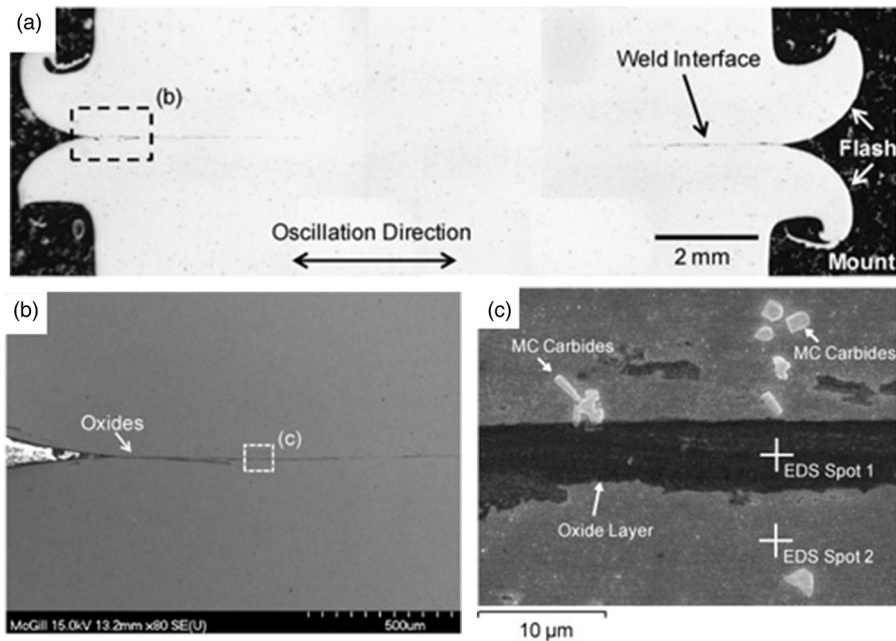
The appropriate welding conditions are necessary for the sufficient and continuous extrusion of interface material, whose absence allows defects to be present at the interface like residual oxides or micro-pores as shown in Fig. 32 in an LFW'd Ti64 joint.²² It should be pointed out that the interface defects are not typically found in RFW'd or LFW'd titanium alloys, while they may appear in superalloys because of their high strength and difficulty to deform at elevated temperatures. Figure 33 shows a typical cross-section of LFW'd WASPALOY superalloy. As mentioned in 'Macroscopic features of joints', evident crack-like defects form in the flash root zone while oxides were identified. Ola *et al.*⁷⁰ also found Hf-based oxide layers at the interface of a LFW'd single crystal CMSX-486 joint. More recently, Senkov *et al.*⁸⁸ studied IFW'd dissimilar Ni-based superalloys: LSHR which is a powder metallurgy alloy, and Mar-M247, a directionally



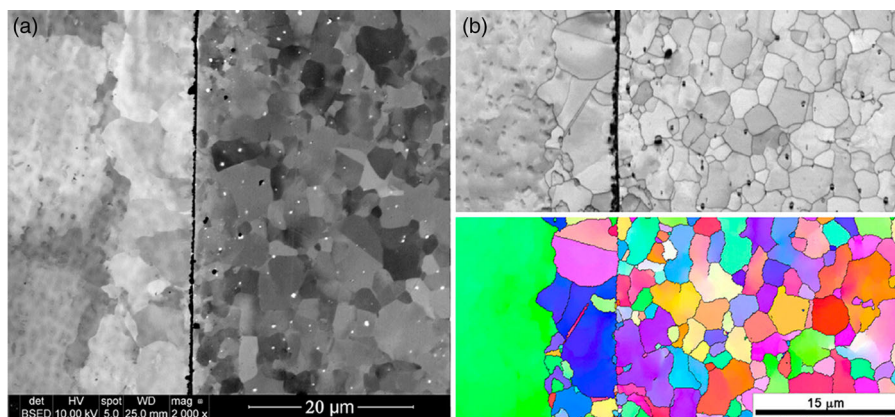
31 Cross-sectional microstructure of a LFW'd Ti64 joint along the oscillation direction: a OM image across the weld, b OM image of the bond line at high magnification showing spherical α phase and c SEM image of WCZ after PWHT⁸⁵



32 Characteristics of a weld centreline for a power input less than 2.4 kW and b axial shortening less than 2 mm during LFW Ti64²²



33 OM macrograph of a LFW'd Waspaloy *a*, and *b* low magnification SEM image of the area within the rectangle in *a*, and *c* high magnification SEM image of the area within the rectangle in *b*⁵⁷



34 Backscatter electron image of IFW'd dissimilar Mar-M247 to LSHR superalloys joint *a*, and EBSD images of the interface (image quality (top) and inverse pole figure (bottom) maps). Mar-M247 is on the left and LSHR on the right of the weld line⁸⁸

solidified alloy. The weld interface was covered with a continuous thin layer of aluminium oxide as shown in Fig. 34. When the flash root appears not closed in welds, oxygen is allowed inside despite the short welding time. It should also be pointed out that such oxides are often associated with insufficient self-cleaning. They commonly found in dissimilar joints between materials of significantly different resistances to deformation, which leads to the flash being produced only by the softer material side.

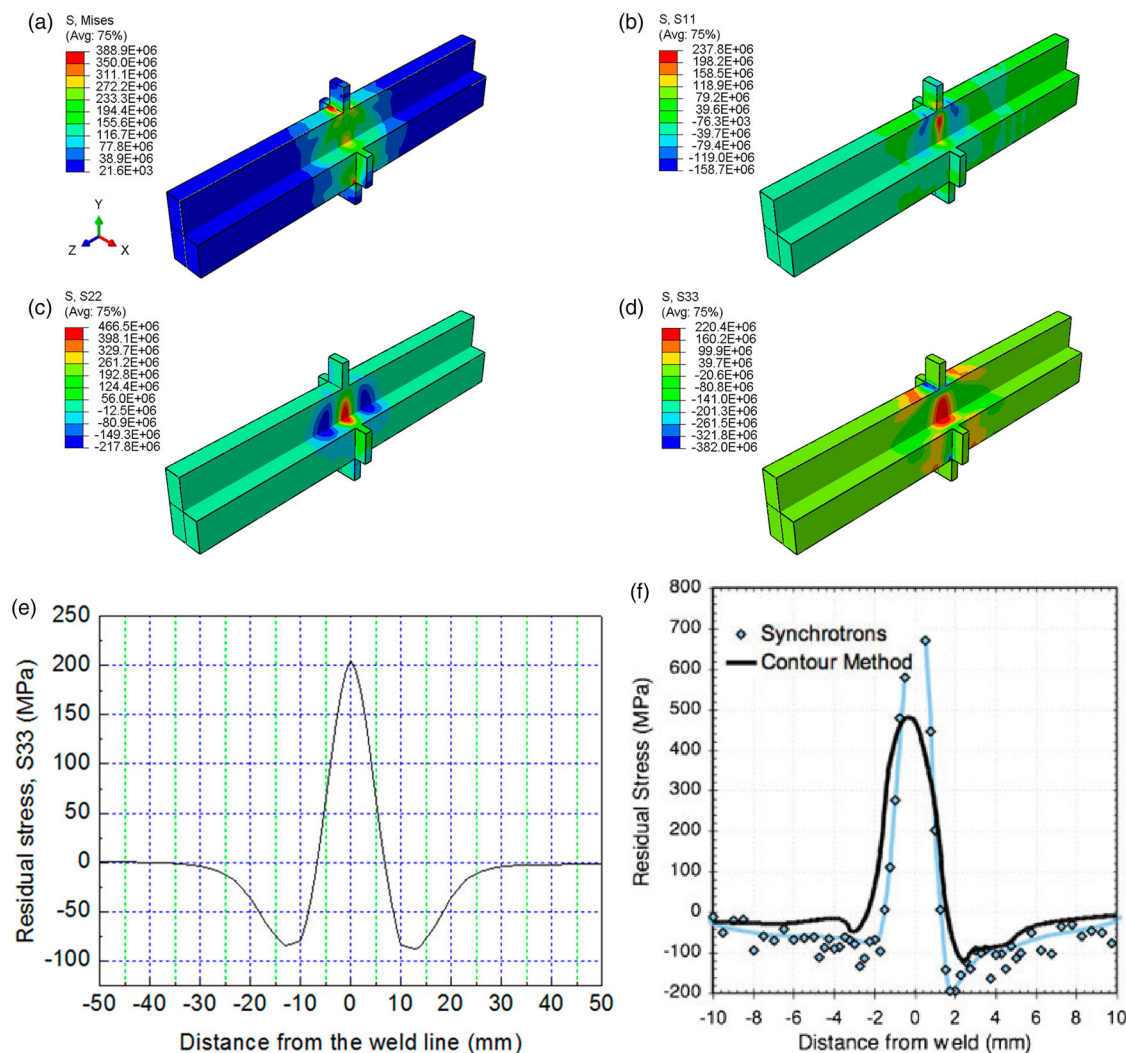
Residual stresses

Residual stresses are probably one of the most important factors affecting weld structures performance in service. Tensile residual stresses may reduce the performance or

cause failure of components as they affect fatigue strength, creep or environmental degradation. Compressive residual stresses, on the other hand, are mostly beneficial at the expense of reduced buckling loads. As a mixture of tensile and compressive residual stresses forms in joints at the end of the process, they need to be carefully controlled. There is a very small number of published works on residual stresses of RFW and LFW, with most of them focusing on superalloys and titanium alloys.

Estimation of residual stresses using numerical analysis

Residual stresses in FW are usually the result of thermal mismatch due to the different cooling rates at various positions across the weld and the final microstructural



35 Representative residual stresses in LFW'd Ti64 joint: a equivalent stress, von Mises, b X-direction stress, S11, c Y-direction stress, S22, d Z-direction stress, S33 and e distribution of S33 across the weld line along Z-axis. f Residual stress (y direction) profiles for LFW'd Ti64 sample with both synchrotron measurements and contour method¹⁶

changes. Changes in strain are caused by the combination of thermomechanical deformation, and constraints to thermal expansion.⁶⁰

According to the representative publications on residual stresses in LFW'd Ti64 by Frankel *et al.*¹⁶ and IFW'd superalloy by Grant *et al.*⁴¹, we built coupled thermomechanical models in ABAQUS, where a minimum simple flash formed around the weld, to illustrate the residual stress distribution. In each model, three steps were set up, the first one being the forging state, the second one being the cooling phase without pressure with the sample clamped, and in the third one the clamps being released. An initial temperature field was applied to the weld with a custom exponential distribution as a function of distance from the weld interface.

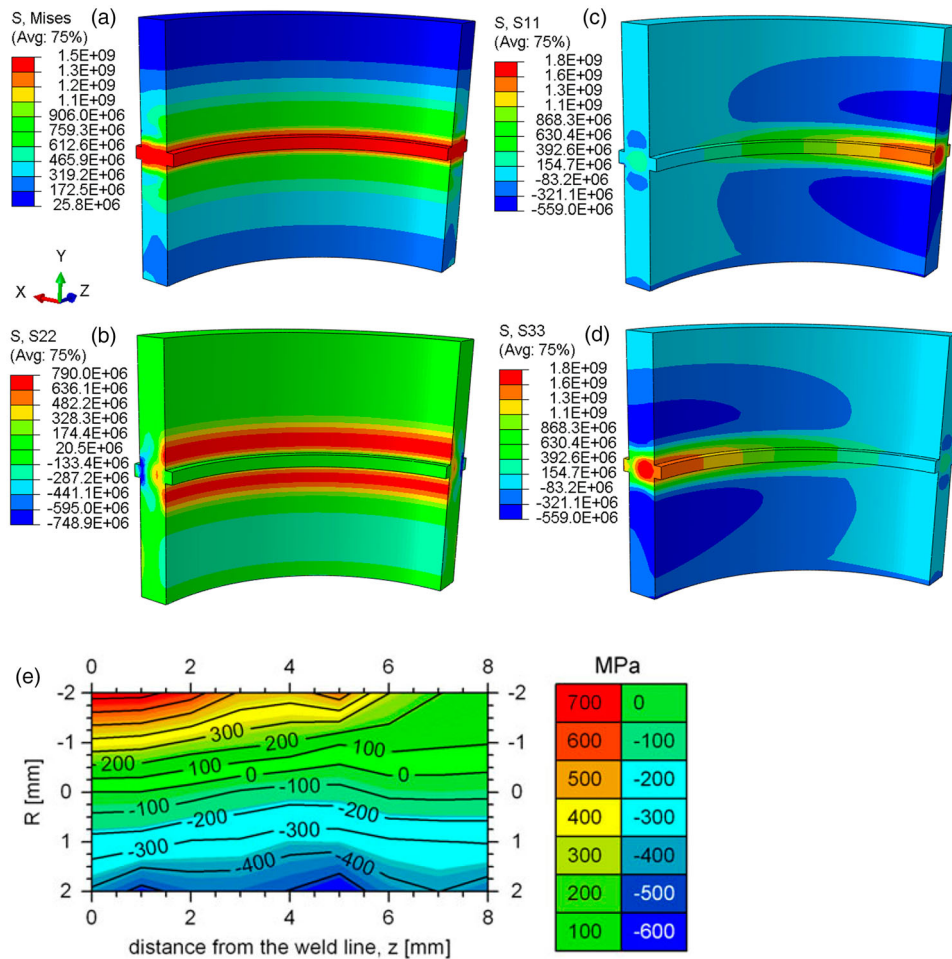
Results in Figs. 35 and 36 show the non-uniform development of large residual stresses across the weld line. For LFW'd Ti64, the residual tensile stresses at the interface are largest in the Y-direction (Fig. 35c), while the largest compressive stresses are present at the flash root in the Z-direction (Fig. 35d). These results are to some extent comparable to those measured by Frankel *et al.* (Fig. 35f).¹⁶

In the case of IFW'd superalloy tube (Fig. 36),^{89–91} the residual tensile stresses in the hoop direction are the largest near the inner wall, while the largest compressive stresses are in the axial direction near the outer wall of the tube. On the whole, the weld developed large von Mises residual stresses (Fig. 36a).

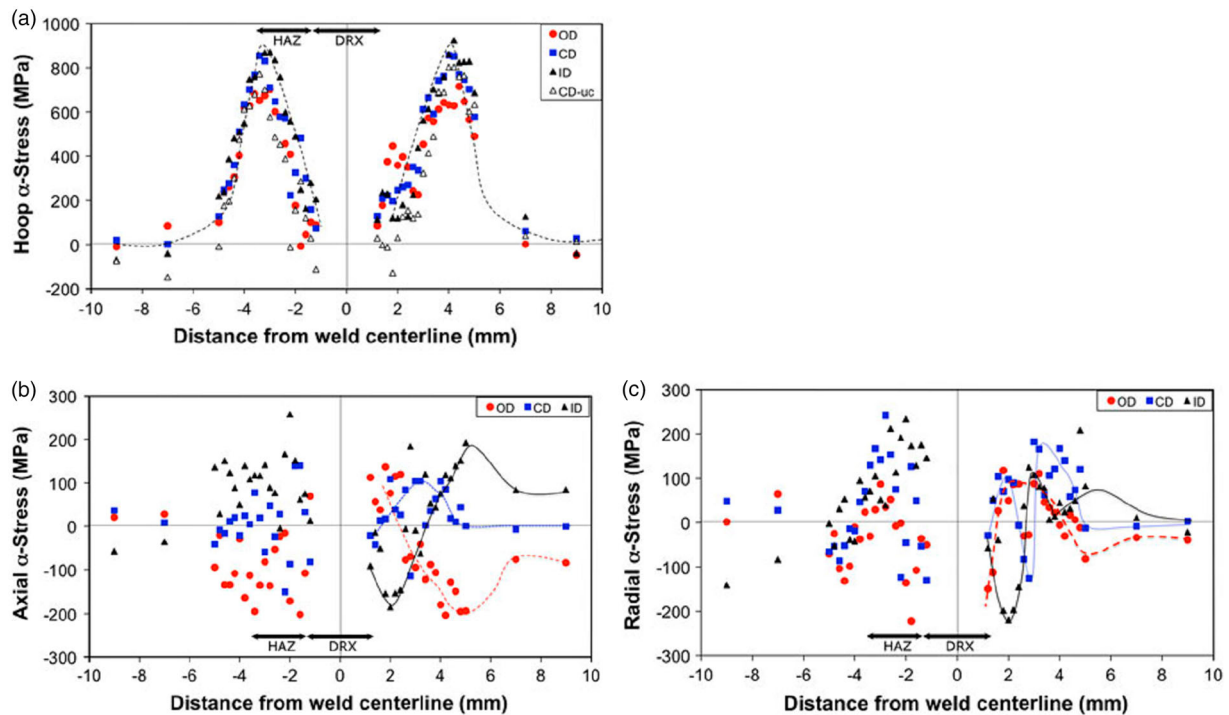
Effect of material type

The level of residual stresses generated in these welds is greatly affected by the high temperature properties of the material. Hence joints of Ni-based superalloys display larger residual stresses than for example joints of Ti alloys.

Attallah *et al.*⁶¹ were the first to measure residual stresses in IFW'd Ti-6246 (Fig. 37). The hoop direction showed the highest levels of stress which were mostly tensile (Fig. 37a). Within the HAZ and DRX region, stresses decrease from a maximum towards zero at the centre line. A very similar residual stress distribution was also found in the IFW'd high strength steel by Moat *et al.*⁹² Iqbal *et al.*⁹⁰ found that the original microstructure has little effect on the maximum residual stresses for a wide range



36 Representative residual stresses in IFW'd IN718 tube joint: a equivalent stress, von Mises, b X-direction stress, S11, c Y-direction stress, S22 and d Z-direction stress, S33. e Axial residual stresses for IFW 720Li Ni-based superalloy⁸⁹ compared to c.



37 The α -Ti residual stresses distribution in as-IFWed Ti-6246 (located at +2 mm [outer diameter: OD], 0 mm [centre diameter: CD] and -2 mm [inner diameter: ID] from the centre of the cylinder thickness): a hoop α -stress, b axial α -stress and c radial α -stress⁶¹

of energy input conditions for the IFWs of fine grain and coarse grain RR1000 nickel-base superalloys. Recently, Song *et al.*⁹³ measured the residual stresses in LFW'd AA2024 joints and found the maximum tensile stress to be located 10 mm away from the weld line, and the highest compressive stress at twice the distance.

Effects of process parameters

An important process parameter which affects residual stresses is pressure, whether friction or forging, during FW. Forging pressure was found to have a significantly effect in LFW'd Ti64 by Romero *et al.*⁶⁰ and Turner *et al.*⁹⁴ as well as friction pressure (Fig. 38).⁶⁰ As friction pressure increases, residual stresses reduce significantly as welds of lower temperature are produced. Usually the increase in pressure is proportional to power input, the decrease of welding time and the accelerated material flow, reduces the total weld energy and in effect the total heat load.

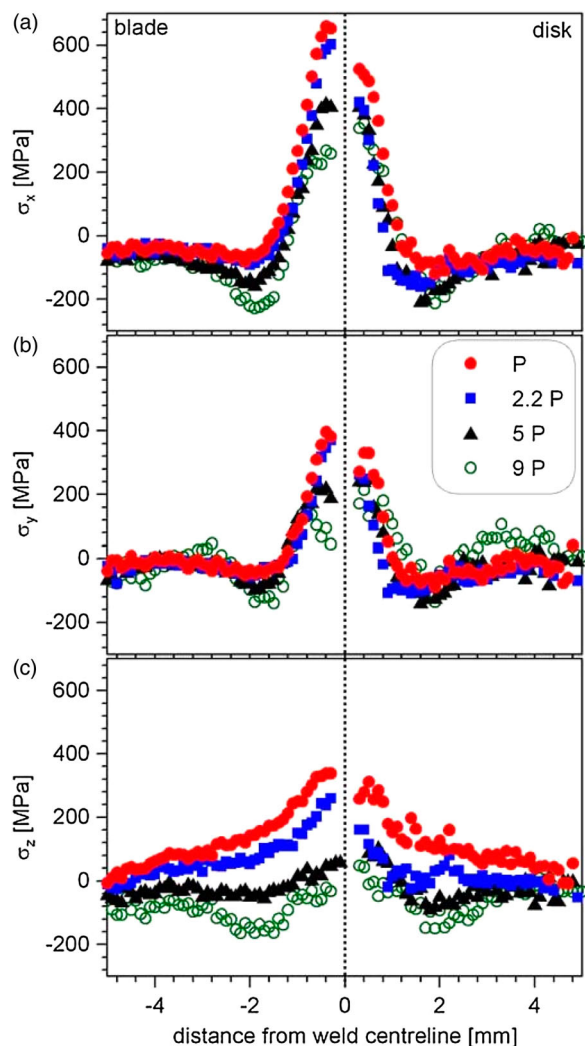
As expected, residual stress values are also sensitive to the workpiece size. This is associated with constraint stresses from the welding jig.⁸⁹

As for other process parameters, such as rotational speed, oscillation frequency and amplitude, and even friction time, they appear to be less important for the development of residual stresses as they affect heat input during welding and not while cooling.

Effect of PWHT

Experiments have shown that an appropriate PWHT reduces residual stresses to acceptable levels. Preuss *et al.*⁹¹ compared the residual stresses of the IFW'd IN718, Alloy 720Li and RR1000 Ni-based superalloy joints before and after heat treatment. For all three materials, axial stress fields display a significant bending moment and the largest tensile stresses in the hoop direction displayed were close to the inner diameter of the tubes. However, due to the significant difference in volume fraction of γ' and thermomechanical properties, the hoop stresses in Alloy 720Li and RR1000 were 1.5 times and 2.0 times higher than that of IN718, respectively. Moreover, residual stresses in the axial and radial directions decreased significantly, while the largest tensile stresses were unaffected for the three alloys after PWHT.

As for titanium alloys, Frankel *et al.*¹⁶ studied the effect of PWHT on residual stresses for Ti64 and Ti-6Al-2Sn-4Zr-2Mo (Ti6242) joints (Fig. 39). As expected, a heat treatment 100°C below the conventional aging temperature of Ti64 (PWHT1) was ineffective. The conventional heat treatment (PWHT2) was shown to reduce tensile peak stress and left intact those in the Ti6242 weld (see Fig. 39e), which was not surprising in view of its superior high temperature capabilities of the near α Ti alloy. A further 100°C increase in conventional heat treatment temperature (PWHT3) was necessary to relax stresses below 100 MPa (Fig. 39f). Preuss *et al.*⁹⁵ also measured the residual stresses of IMI550 (Ti-4Al-2Sn-4Mo-0.5Si) joints and found similar distribution of residual stresses. Moderate residual stresses were observed after increasing the PWHT temperature by 100 or 150°C. All conditions displayed a significant hydrostatic tensile stress field component, which increase the difficulty in relieving them.



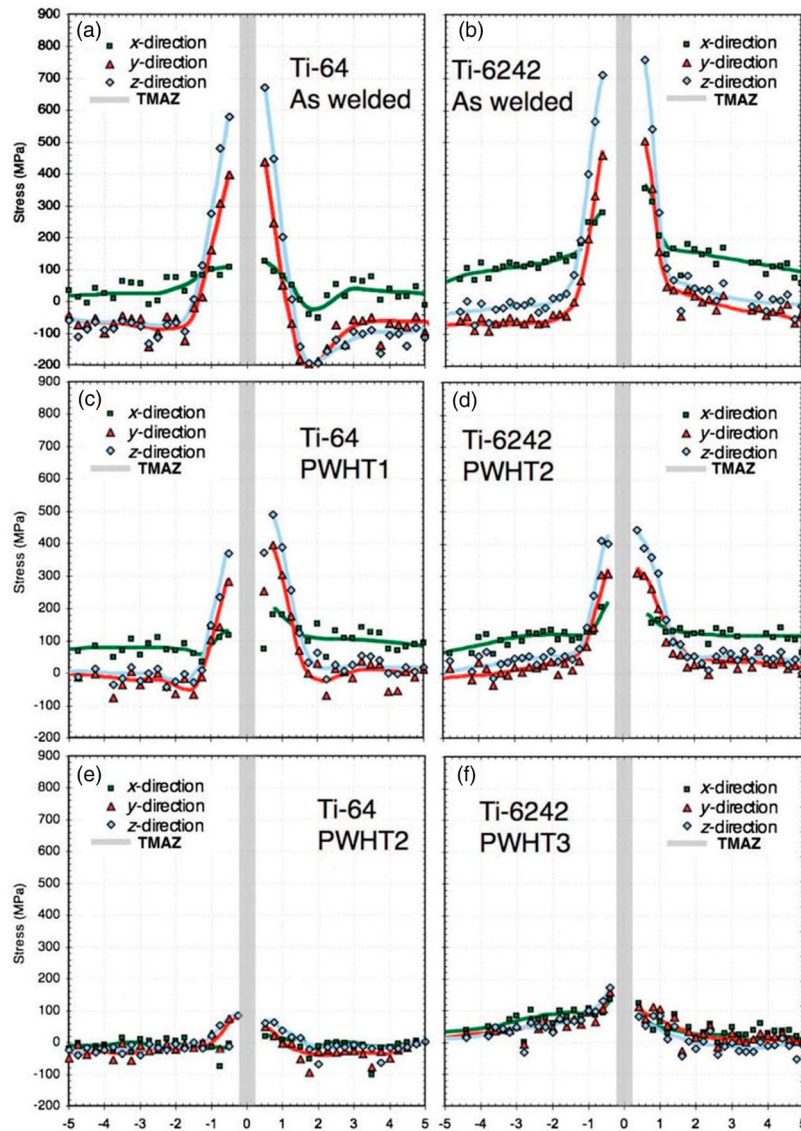
38 Influence of friction pressure on residual stress components (a x-direction: transverse direction; b y-direction: oscillation direction; c z-direction: axial direction) in LFW'd Ti64 joints⁶⁰

Properties of joints

The mechanical properties of FW'd joints largely depend on the distinct microstructural zones that have formed at the vicinity of the weld line during welding. In addition, the residual stresses also affect the service performance of joints. This section summarises the characteristics of joint properties, like the tensile strength, hardness, fatigue and fracture behaviours, and corrosion performance, with respect to microstructure. As the effects of process parameters on the overall tensile strength have already been described in section 'Process parameters', tensile strength is only briefly discussed here.

Tensile properties of joints

For dissimilar FWs, the aluminium to steel combinations as well as the dissimilar steels joints have attracted most of the attention. A perfect bond between Al and steel FWs would have no intermetallic phases present, where an interlayer were to be used. Şahin and Akata⁹⁶ found that the tensile strength decreases with increasing heat input.



39 Residual stresses in three principal directions (x-direction: oscillation direction, y-direction: transverse direction, z-direction: axial direction) for a–c LFW Ti64 and d–f LFW Ti6242 in as welded and PWHT conditions¹⁶

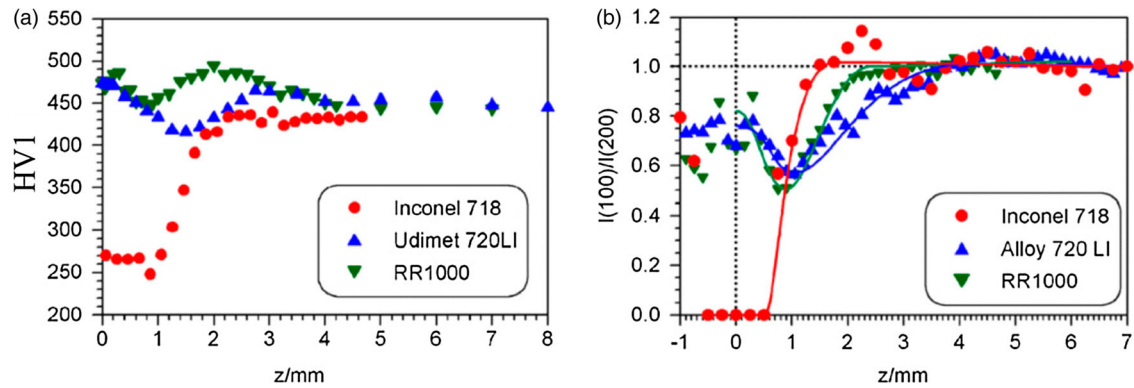
A technique to improve joint properties in many cases is with PWHT. The precipitation hardened Ni-based superalloys usually require PWHT to relieve high residual stresses and recover the precipitate microstructure. Damodaram *et al.*³² studied the room temperature tensile properties of CDFW'd IN718. It was found to be dependent on thermal history conditions, which was due to the dissolution of the strengthening γ'' precipitates in the weld zone. After a direct aging (DA) treatment, there was significant improvement of the mechanical properties to reach PM conditions.

Hardness of joints

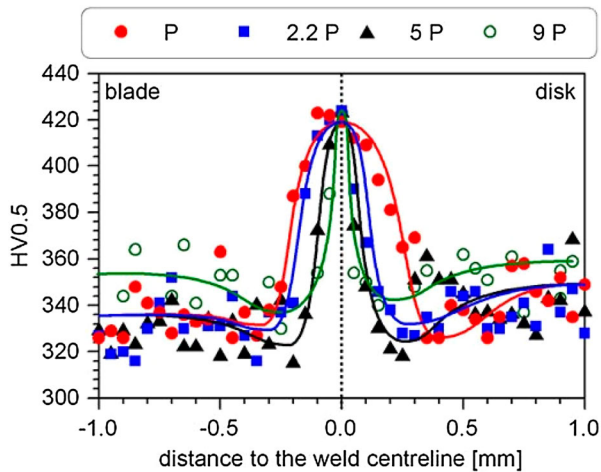
Microhardness of FW'd joints may be affected by a combination of some or all of the following factors: modification of microstructure, dissolution and/or reprecipitation of strengthening phases, cold work and grain size. Microhardness studies have proven to be the most reliable method to identify microstructural changes across the weld region. Some interesting aspects have been

reported on superalloys, titanium alloys, steels and dissimilar joints, as well as the effect of PWHT.

Preuss *et al.*⁹¹ compared the hardness of Ni-based superalloys with different γ' contents for IFWs. Two different types of hardness profiles for Ni-based superalloys were identified. The IN718 exhibited a significant hardness drop in WCZ due to the absence of reprecipitation in this region after welding. However, Udimet 720Li and RR1000, of a higher γ' , showed a hardness peak in the HAZ (Fig. 40) which is because of the large driving force for precipitation even at very high cooling rates. For both 720Li and RR1000 the volume fraction of γ' recovers close to the weld line, as the high levels of γ' stabilising alloying elements (Ti and Al) in solution force reprecipitation even at extremely high cooling rates. In a further, more detailed study of IFW'd RR1000 samples, Preuss *et al.*⁶⁸ found a bimodal distribution of hardness with the peaks located at 0.3 and 2 mm away from the weld line, while the PWHT sample had a single hardness peak.⁶⁹



40 Vickers hardness profiles *a* and *b* the normalised measured integrated intensity of the γ' superlattice reflection in IFW'd IN718, U720Li and RR1000 joints in the as-welded conditions⁹¹



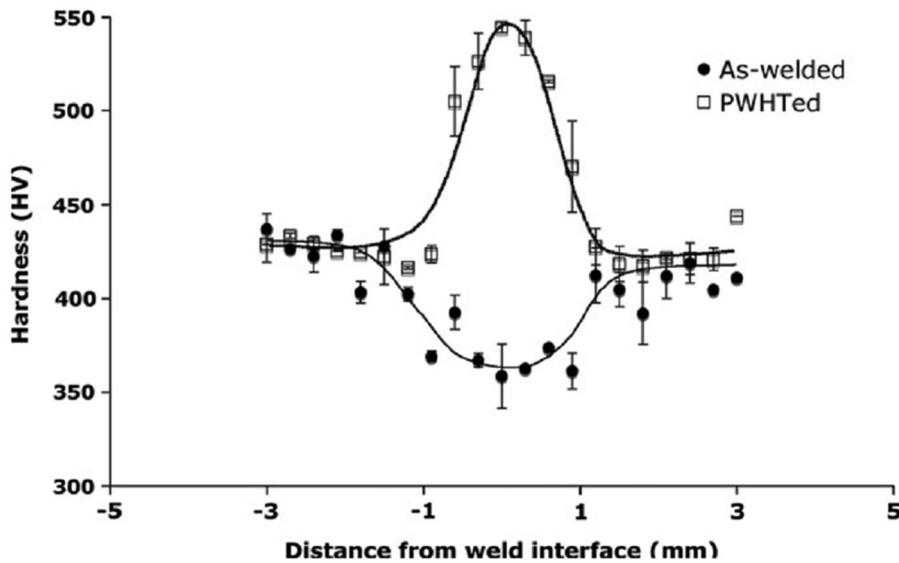
41 Effect of axial pressure on microhardness of LFW'd Ti64 (error in measurement ± 10 HV in the WCZ and TMAZ and ± 20 HV in the PM)⁶⁰

Hardness measurements have also been carried out for various titanium alloys including Ti64, Ti17, Ti6242, Ti6246 and Ti5553. The exact shape of the hardness profile depends on the alloy type and the

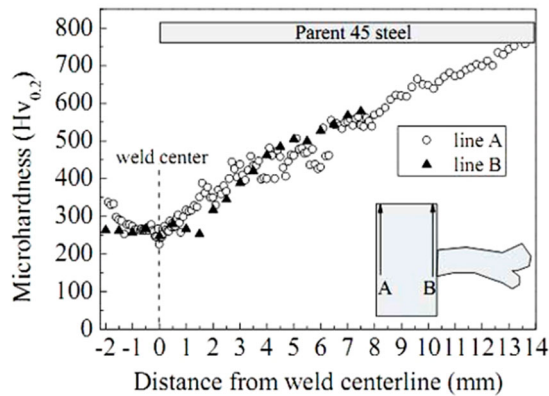
process parameters employed. They usually exhibit a 'W' shape with the highest values occurring in the WCZ, due to phase transformations and microstructural refinement. Romero *et al.*⁶⁰ investigated the effect of forging pressure on the microhardness of Ti64 and found the hardness in the weld region to be remarkably higher than TMAZ and PM (Fig. 41). The high hardness region was found to be inversely proportional to friction pressure while maxima at the very weld line remained unaffected.

More recently, Guo *et al.*⁹⁷ found the opposite hardness profile for LFW Ti6246 (Fig. 42), due to the β phase co-existing with the soft α'' martensite in the as-welded conditions. After PWHT, the hardness in WCZ increased to an aged microstructure condition. Dalgaard *et al.*⁶² also reported a similar hardness distribution for Ti-5553 joints. In this case, softening was attributed to the depletion of the strengthening globular α phase at grain boundaries in the weld.

Mateo *et al.*¹⁸ performed single, double and triple Ti6246 welds to simulate repeated repairs with LFW in the same position of a component. The effect of each successive repeat weld was to increase in a similar fashion the number of maxima of hardness. At the same time, the maximum value was becoming lower than that of the



42 Microhardness profile across the weld interface at as-weld and PWHTed conditions obtained from the LFW'd Ti6246 weld⁹⁷



43 Microhardness of LFW'd medium carbon steel joint⁷³

single weld for each repeat weld, but remained unshifted with respect to the weld centre.

In the case of carbon steels, microhardness is highly dependent on the resultant microstructure. Ma *et al.*⁷³ studied the microhardness variation of LFW'd medium carbon steel joints (Fig. 43), which decreases from the PM to the WCZ. The microstructure in the WCZ was normalised to superfine ferrite (α)+pearlite (P) structures at a temperature higher than 800°C.

In the case of LFW'd AA2024-T4, Rotundo *et al.*⁵⁶ found that a slight decrease in hardness in the TMAZ, while it was slightly higher in the WCZ. When they studied the hardness of LFW'd 2124Al/25 vol.-% SiC_p,⁹⁸ a decrease by 10% in the WCZ was found.

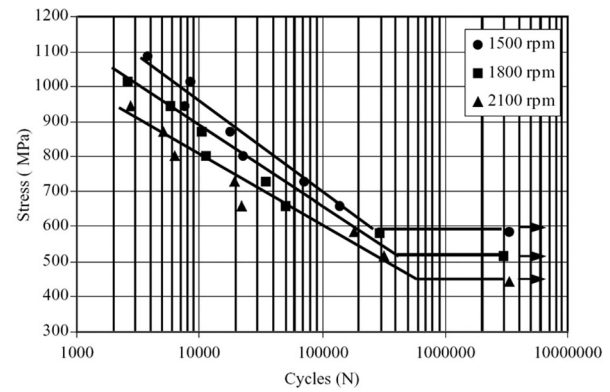
Fatigue properties of joints

Fatigue strength is of particular importance in mechanical design, as components are usually subject to fluctuating stresses. Despite the small number of materials tested it can be seen that the fatigue properties of FW'd joints are almost the same or close to those of PM.

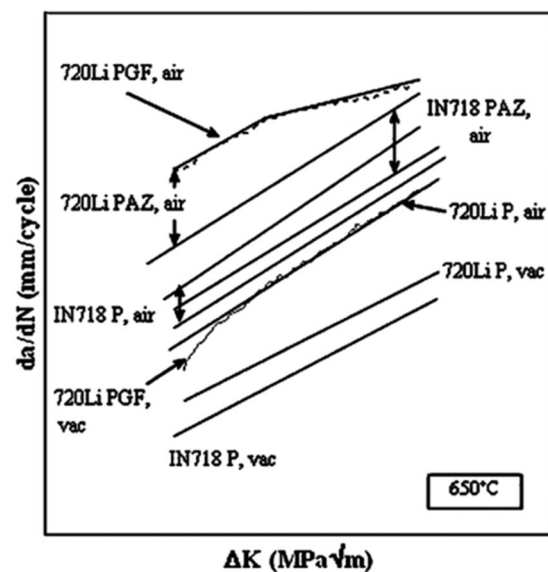
Hasalik *et al.*⁹⁹ investigated the fatigue behaviour of CDFW'd AISI 304 stainless steel to AISI 4340 steel joints (Fig. 44). The rotation bending fatigue strength decreased because of the chromium carbide precipitation at the welding interface with increasing rotational speed, as the carbides at the GBs provide preferential sites for cavity nucleation. Sahin *et al.*⁵¹ reported that the fatigue strength of CDFW'd AISI 1040 steel was very close to the PM.

For the case of IFW'd Ni-based superalloys, Li *et al.*¹⁰⁰ investigated the high temperature behaviour of PWHTed welds between 720Li/RR1000 and IN718 alloys. All welds displayed a fatigue strength which is comparable to the weaker PM, IN718. However, the fatigue crack growth (FCG) resistance was lowered from those of the PMs to various extents when a crack was introduced into different zones of the welded joints (Fig. 45). Such reduction in FCG resistance results from the weld microstructure being more sensitive to environmental embrittlement at high temperatures. The formation of M₂₃C₆ carbide films along grain boundaries was observed to have an additional detrimental effect on high temperature crack growth rates.

There are very few publications on the fatigue performance of LFW'd joints. Mateo *et al.*¹⁸ measured a repaired



44 S-N curves of CDFW'd AISI304 to AISI4340 for different rotational speeds⁹⁹

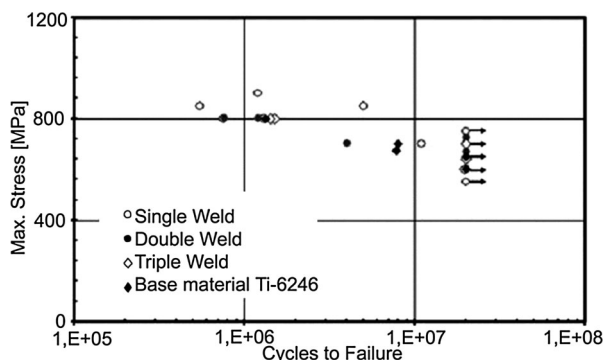


45 Fatigue crack growth resistance curves obtained for different locations of IFW'd joints (720Li to IN718) at 650°C in air compared with PM in vac. PGF: primary γ free zone, PAZ: plastically affected zone, P: parent¹⁰⁰

Ti6246 part at room temperature with a limited number of specimens in a resonance machine (Fig. 46). In no case did the specimens fracture at or near the weld line, with cracks originating either at the surface or in the bulk. Recently, Wen *et al.*¹⁰¹ studied LFW'd Ti64 joints. Both the PM and the weld exhibited equivalent fatigue lifetime, while the cyclic strain hardening exponent of the welded joint was lower than that of PM. Cyclic stabilisation appeared at low strain amplitudes up to 0.6% in both the PM and weld. However, cyclic softening occurred at high strain amplitudes. Rotundo *et al.*⁵⁶ performed fatigue experiments with LFW'd AA2024 joints and found a fatigue strength of about 90% of the PM, as no welding defects existed and the joint had a fine microstructure.

Impact and fracture toughness of joints

For toughness of RFW'd joints researchers have focused on joints of similar or dissimilar steels. Sathiyar *et al.*¹⁰²



46 S-N curves of LFW Ti6246 welds¹⁸

investigated the Charpy impact strength of CDFW'd ferritic stainless steel AISI430. Although the joint exhibited a 90–95% tensile strength of the PM, the impact strength was reduced significantly by chromium carbides at the grain boundaries in the HAZ. However, Sahin⁵¹ found a slightly higher notch-impact toughness of CDFW'd AISI1040 steel than the PM. In the case of CDFW'd AISI304, the impact strength was double that of PM.⁵⁰

Arivazhagan *et al.*¹⁰³ found that the impact toughness of the CDFW'd dissimilar AISI4140 to AISI304 in the as-welded condition was lower than the low alloy steel. Satyanarayana *et al.*¹⁰⁴ studied the CDFW'd dissimilar austenitic to ferritic stainless steel joints, and found that the impact toughness of the weld was better than the base ferritic stainless steel. Regression analysis showed that individual welding parameters do not affect notch tensile strength with the exception of friction and forge pressures. Burn-off has an influence on impact toughness while forging pressure and burn-off have an interactive effect.

Reddy *et al.*¹⁰⁵ improved the notch tensile strength and impact toughness of CDFW'd maraging steel and low alloy steel joints with nickel as an interlayer. In the case of En-24 steel with an interlayer the PWHT produced welds with the same toughness as PM, while when there was no interlayer present it was further reduced.

Although impact and fracture toughness are essential properties to know when designing high performance structures, there is scarcity of publications on LFW'd joints. Ma *et al.*¹⁷ analysed the impact toughness of LFW'd Ti64 joints using Mesnager U-notch specimens. In all cases failure occurred in the PM, with the crack initiating in the notch at the WCZ and propagating through the TMAZ into the PM. Ma *et al.*¹⁰⁶ also studied the positive effect of PWHT on the impact toughness of LFW'd Ti17 joints. Corzo *et al.*¹⁰⁷ tested the fracture toughness of LFW'd Ti6246, which was lower than the PM. In one case, both the pre-crack and the final fracture crack grew within the weld line due to the very fine microstructure in the weld zone, which has a very high hardness value but low fracture toughness. In other experiments, the crack travelled towards the HAZ, with fracture toughness values corresponding to the HAZ toughness.

Corrosion properties of joints

Corrosive environments pose an additional challenge to structures particularly when they include a weld. In the case of dissimilar material joints, particular attention

should be paid to the galvanic potential arising from a dissimilar material couple.

Tumuluru¹⁰⁸ investigated the sulphide stress corrosion cracking (SCC) behaviour of IFW'd AISI 4134-H and API Grade E pipe steel (both are low alloy steels). The crack originated very close to the weld line and fracture occurred at the AISI 4137-H steel side, to produce a cracking resistance lower than that of the PM for all stress levels.

Satyanarayana *et al.*¹⁰⁴ measured pitting corrosion of CDFW'd austenitic and ferritic stainless steels joints in an electrolyte of H₂SO₄ and HCl. The similar ferritic stainless steel joints exhibited the highest pitting resistance while the dissimilar ones showed the least. Recently, Hazra *et al.*¹⁰⁹ investigated the pitting corrosion resistance of high nitrogen steel friction welds in an electrolyte of H₂SO₄ and NaCl. As in previous work, the pitting resistance of all joints was inferior to that of PM and was inversely dependent on forging force, and the localised corrosion behaviour was affected significantly by microstructure changes within the weld interface.

Arivazhagan *et al.*^{103,110} investigated the hot corrosion behaviour of dissimilar CDFWs with AISI 4140/AISI 304 for different burn-off lengths in a molten salt environment. In each case, the weld interface was found to be more susceptible to degradation than the PM due to inter-diffusion of element across the interface and the formation of IMCs. As AISI 4140 has a low Cr content corrosion on its side was more severe. It was also shown that increasing the burn-off length could reduce the corrosion resistance of the joints, as more carbide and IMC precipitates formed at the interface with longer welding times, as more heat was input.

It is expected that the corrosion behaviour of LFW'd joints would be the same as RFW'd ones for the same materials. There is only a recently published paper by Astarita *et al.*,¹¹¹ which investigated LFW'd dissimilar stainless steel-titanium alloy joints in an aqueous environment. A crack-like unbonded zone, which often appears at the root of flash due to flash curling, caused galvanic corrosion at the weld interface.

Summary and outlook

FW, which includes RFW and LFW among its other variants, has been widely used to join various components as it produces high quality joints. RFW is a mature technique with an exceptionally wide range of applications. LFW is being used for the production of aero-engine blisks, which is still the only commercial application. A number of possible industrial applications of LFW have been demonstrated, like the production of low volume near-net-shape components and airplane structural components.

The choice of welding parameters affects in a profound way heat generation and material flow during FW and hence microstructure and residual stresses in the weld zone as well as the structural integrity of the weld. The key process parameters of LFW are the oscillation frequency, oscillation amplitude and friction pressure, while in RFW, in addition to friction pressure, there is the relative rotational speed instead of the oscillating frequency and amplitude that is a defining welding parameter. Each individual parameter can determine if a structurally sound joint is formed.

Similar to RFW, the LFW microstructure is characterised by a WCZ where metallurgical phase transitions occur in the presence of high strain rates and rapid heating and cooling often resulting in grain refinement due to recrystallisation. In the case of a material undergoing an allotropic transformation, the high cooling rates also result in the formation of a very fine low temperature phase (i.e. α phase in Ti alloys). When dissimilar materials are joined, the high temperature at the weld line combined with the severe deformation is also likely to promote the formation of IMCs. Next to this region, there is the TMAZ which is exposed to lower temperatures and strains than the weld line, and is characterised by deformed and/or only partly recrystallised grains.

With regards to the residual stresses of the LFW'd joints, they appear in the welding plane, with large tensile strains in the HAZ, dropping to small compressive strains around the weld, and decaying slowly to zero in the far field. Due to the geometry of RFW, where axisymmetric components are joined, residual stresses are a combination of axial, hoop and radial components with the hoop stresses typically being the largest. In many cases, PWHTs are necessary to improve the performance of the joint.

Finally, it should be pointed out that under optimum welding parameters, both LFW and RFW of titanium alloys, Ni-based superalloys, steels, and dissimilar materials joints possess good mechanical properties, with fracture during tensile testing occurring mostly in the PM and the strength of joints being comparable to that of the PM. As for impact toughness and fatigue properties, these joints are also comparable to the PM although severe grain refinement can result in low fracture toughness at the weld line. The joint corrosion properties, especially of the dissimilar joints, suffer from galvanic corrosion effects, and need careful consideration when used in aggressive environments.

With increasingly more demanding specifications for standard components, LFW and RFW technologies attract more attention to solve manufacturing challenges presented by new materials. The physical understanding of the LFW and RFW processes are not complete as some important aspects have not been fully explored, like the formation of the joining mechanism or the reasons that certain groups of alloys cannot be welded.

New variants of these processes are being developed to solve specific problems, for example hybrid RFW where IFW is combined with CDFW. Such an approach deals with the problem of the very high initial torques required to reach extrusion stage. In such a machine an inertial machine arrangement would suffice for the beginning followed by a direct-drive machine which would cover the lower torque requirements of the subsequent phases, and weld large sized parts for example.

Full process modelling has not been achieved yet. Most of the research has been on modelling temperature history, deformation and stress fields. The models developed for RFW and LFW cannot represent the whole process from start to finish and rely on experimental data to represent individual phases of the process. Such efforts should ideally result in predictive tools that provide guidance in the choice of process parameters. However, the complexity of RFW and LFW simply exceeds current process modelling and hardware capabilities. The fluid-dynamics-based Eulerian method or particle-based

meshless method may provide an effective way to model the whole FW process in the future.

As structural material requirements become ever more demanding, high temperature, high corrosion resistant alloys such as Ni-based superalloys continue to remain an important class of materials. As their temperature capability improves, they become more difficult to weld because of their complex alloy chemistry, microstructure and ability to lock in residual stresses during cooling from a very early stage. Areas of further research could be on identifying the process parameter envelope for successful welds for different materials and part geometries, the recrystallisation behaviour of the WCZ (e.g. in a single crystal superalloy), the control of oxidation at the interface and the high temperature corrosion performance.

Finally, an area that has received less attention from the research community, at least when considering the literature available in the public domain, is joint properties such as fatigue, creep and impact among others, which form an essential part of the knowledge that design engineers require to meet specifications.

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References

1. M. Maalekian: 'Friction welding - critical assessment of literature', *Sci. Technol. Weld. Joi.*, **2007**, **12**, (8), 738–759.
2. R. S. Mishra and Z. Y. Ma: 'Friction stir welding and processing', *Mater. Sci. Eng. R.*, **2005**, **50**, 1–78.
3. P. L. Threadgill, A. J. Leonard, H. R. Shercliff and P. J. Withers: 'Friction stir welding of aluminium alloys', *Int. Mater. Rev.*, **2009**, **54**, 49–93.
4. G. Çam: 'Friction stir welded structural materials: beyond Al-alloys', *Int. Mater. Rev.*, **2011**, **56**, 1–48.
5. M. E. Nunn: 'Aero engine improvements through linear friction welding', Proc. 1st Int. Conf. on 'Innovation and integration in aerospace sciences', Belfast, UK, August 2005, CEIAT, Paper 0040.
6. I. Bhamji, M. Preuss, P. L. Threadgill and A. C. Addison: 'Solid state joining of metals by linear friction welding: a literature review', *Mater. Sci. Technol.*, **2011**, **27**, (1), 2–12.
7. Gatwick Technologies Ltd, <http://www.fpe.co.uk/processes/friction-welding>
8. A. Vairis and M. Frost: 'On the extrusion stage of linear friction welding of Ti-6Al-4V', *Mater. Sci. Eng. A*, **1999**, **271**, 477–484.
9. A. Vairis and M. Frost: 'Modelling the linear friction welding of titanium blocks', *Mater. Sci. Eng. A*, **2000**, **292**, 8–17.
10. A. Vairis and M. Frost: 'High frequency linear friction welding of a titanium alloy', *Wear*, **1998**, **217**, 117–131.
11. J. H. Bevington, The New process Welding And Spinning Co.: 'Mode of welding the end of wire, rods, etc.', US Patent 463134A, published 17 November 1891.

12. Handbook Committee: 'Welding, brazing and soldering', in *ASM handbook*, (ed D. L. Olson et al.), 10th edn, Vol. 6, 503–515, 879–886; 1993, Materials Park, OH, ASM International.
13. R. Maurya and J. Kauzlarich, Caterpillar Tractor Co.: 'Bonding apparatus-friction welding by reciprocal motion', US Patent 3420428-A, DE1552871-A, CA844858-A, published 7 January 1969.
14. J. H. L. Garton: 'Investigation into linear friction welding of blisks', B.E. thesis, University of Bristol, Bristol, England, 1987.
15. H. H. Koo and W. A. Baeslack III: 'Structure, properties, and fracture of linear friction welded Al-Fe-V-Si alloy 8009', *Mater. Charact.*, 1992, **28**, 157–164.
16. P. Frankel, M. Preuss, A. Steuwer, P. J. Withers and S. Bray: 'Comparison of residual stresses in Ti-6Al-4V and Ti-6Al-2Sn-4Zr-2Mo linear friction welds', *Mater. Sci. Technol.*, 2009, **25**, (5), 640–650.
17. T. J. Ma, W. Y. Li and S. Y. Yang: 'Impact toughness and fracture analysis of linear friction welded Ti-6Al-4V alloy joints', *Mater. Des.*, 2009, **30**, 2128–2132.
18. A. Mateo, M. Corzo, M. Anglada, J. Mendez, P. Villechaise, J. P. Ferte and O. Roder: 'Welding repair by linear friction in titanium alloys', *Mater. Sci. Technol.*, 2009, **25**, (7), 905–913.
19. K. T. Slattery, The Boeing Company: 'Structure assemblies and preforms thereof formed by linear friction welding', US Patent 7225967B2, published 5 June 2007.
20. A. C. Addison: 'Linear friction welding information for production engineering', TWI Industrial Member Report Summary 961/2010, TWI, Granta Park, UK, 2010.
21. I. Bhamji, M. Preuss, P. L. Threadgill, R. J. Moat, A. C. Addison and M. J. Peel: 'Linear friction welding of AISI 316L stainless steel', *Mater. Sci. Eng. A*, 2010, **528**, 680–690.
22. P. Wanjara and M. Jahazi: 'Linear friction welding of Ti-6Al-4V: processing, microstructure, and mechanical-property inter-relationships', *Metall. Mater. Trans. A*, 2005, **36**, (8), 2149–2164.
23. V. I. Vill: 'Friction welding of metals'; 1962, New York, USA, American Welding Society.
24. Y. X. Cai, S. T. Sun, G. Z. Zhu and X. Wang: 'The widening process of the deformation layer and the high temperature area in friction welding', *Trans. of the China Welding Institution*, 1984, **5**, (2), 61–68. (in Chinese)
25. X. X. Zhang, B. L. Xiao and Z. Y. Ma: 'A transient thermal model for friction stir weld. Part I: the model', *Metall. Mater. Trans. A*, 2011, **42**, (10), 3218–3228.
26. A. Vairis: 'Investigation of frictional behaviour of various materials under sliding conditions', *Eur. J. Mech. A – Solid.*, 1997, **16**, (6), 929–945.
27. M. Maalekian, E. Kozeschnik, H. P. Brantner and H. Cerjak: 'Comparative analysis of heat generation in friction welding of steel bars', *Acta Mater.*, 2008, **56**, (12), 2843–2855.
28. Y. C. Yang, W. L. Chen and H. L. Lee: 'A nonlinear inverse problem in estimating the heat generation in rotary friction welding', *Numer. Heat Transfer, Part A*, 2011, **59**, (2), 130–149.
29. C. J. Bennett, M. M. Attallah, M. Preuss, P. H. Shipway, T. H. Hyde and S. Bray: 'Finite Element Modeling of the Inertia Friction Welding of Dissimilar High-Strength Steels', *Metall. Mater. Trans. A*, 2013, **44**, (11), 5054–5064.
30. A. Vairis: 'Superplasticity effects and strain rate dependency in a material joining process', *J. Eng. Sci. Technol. Rev.*, 2008, **1**, (1), 28–32.
31. L. B. Yang: 'Modelling of the inertia welding of INCONEL 718', PhD thesis, The University of Birmingham, UK, 2010.
32. R. Damodaram, S. G. S. Raman and K. P. Rao: 'Microstructure and mechanical properties of friction welded alloy 718', *Mater. Sci. Eng. A*, 2013, **560**, 781–786.
33. M. Soucail, A. Moal, E. Massoni, C. Levaillant, L. Naze and Y. Bienvenu: 'Microstructural study and numerical simulation of inertia friction welding of astroloy', *Superalloys* 1992, 1992, 847–856.
34. T. C. Nguyen and D. C. Weckman: 'A thermal and microstructure evolution model of direct-drive friction welding of plain carbon steel', *Metall. Mater. Trans. B*, 2006, **37**, (2), 275–292.
35. H. Kucsu, I. Becenen and M. Sahin: 'Evaluation of temperature and properties at interface of AISI 1040 steels joined by friction welding', *Assembly Autom.*, 2008, **28**, (4), 308–316.
36. S. A. Mousavi and A. R. Kelishami: 'Experimental and numerical analysis of the friction welding process for the 4340 steel and mild steel combinations', *Weld. J.*, 2008, **87**, (7), 178s–186s.
37. S. K. Singh, K. Chattopadhyay, G. Phanikumar and P. Dutta: 'Experimental and numerical studies on friction welding of thixo-cast A356 aluminum alloy', *Acta Mater.*, 2014, **73**, 177–185.
38. G. D. Wen, T. J. Ma, W. Y. Li, X. Li, J. L. Li, T. Chen, R. Wen, J. Niu and H. Z. Guo: 'Mathematical modelling of joint temperature during linear friction welding of dissimilar Ti-6.5Al-3.5Mo-1.5Zr-0.3Si and Ti-5Al-2Sn-2Zr-4Mo-4Cr alloys', *J. Eng. Sci. Technol. Rev.*, 2012, **5**, (3), 35–38.
39. W. Y. Li and F. F. Wang: 'Modeling of continuous drive friction welding of mild steel', *Mater. Sci. Eng. A*, 2011, **528**, 5921–5926.
40. H. Ates, M. Turker and A. Kurt: 'Effect of friction pressure on the properties of friction welded MA956 iron-based superalloy', *Mater. Des.*, 2007, **28**, 948–953.
41. B. Grant, M. Preuss, P. J. Withers, G. Baxter and M. Rowson: 'Finite element process modelling of inertia friction welding advanced nickel-based superalloy', *Mater. Sci. Eng. A*, 2009, **513–514**, 366–375.
42. F. F. Wang, W. Y. Li, J. L. Li and A. Vairis: 'Process parameter analysis of inertia friction welding nickel-based superalloy', *Int. J. Adv. Manuf. Technol.*, 2014, **71**, (9–12), 1909–1918.
43. W. Y. Li, T. J. Ma and J. L. Li: 'Numerical simulation of linear friction welding of titanium alloy: Effects of processing parameters', *Mater. Des.*, 2010, **31**, 1497–1507.
44. W. Y. Li, F. F. Wang, S. X. Shi, T. J. Ma, J. L. Li and A. Vairis: '3D finite element analysis of the effect of process parameters on linear friction welding of mild steel', *J. Mater. Eng. Perform.*, 2014, **23**, (11), 4010–4018.
45. W. Y. Li, T. J. Ma, S. Q. Yang, Q. Z. Xu, Y. Zhang, J. L. Li and H. L. Liao: 'Effect of friction time on flash shape and axial shortening of linear friction welded 45 steel', *Mater. Lett.*, 2008, **62**, 293–296.
46. S. Y. Wang, G. F. Zhang, F. X. Qiu and Y. Guo: 'Study on the electric current on conducting-friction welding process', *Journal of Xi'an Jiaotong University*, 1944, **28**, (2), 37–41. (in Chinese)
47. W. Y. Li, M. Yu, J. L. Li, G. F. Zhang, S. Y. Wang: 'Characterizations of 21–4N to 4Cr9Si2 stainless steel dissimilar joint bonded by electric-resistance-heat-aided friction welding', *Mater. Des.*, 2009, **30**, (10), 4230–4235.
48. J. Luo, L. F. Li, Y. L. Dong, X. L. Xu: 'A new current hybrid inertia friction welding for nickel-based superalloy K418-alloy steel 42CrMo dissimilar metals', *Int. J. Adv. Manuf. Technol.*, 2014, **70**, 1673–1681.
49. M. Sahin, E. Çil and C. Misirli: 'Characterization of properties in friction welded stainless steel and copper materials', *J. Mater. Eng. Perform.*, 2013, **22**(3): 840–847.
50. M. Sahin: 'Evaluation of the joint-interface properties of austenitic-stainless steels (AISI 304) joined by friction welding', *Mater. Des.*, 2007, **28**, 2244–2250.
51. M. Sahin, H. E. Akata and T. Gulmez: 'Characterization of mechanical properties in AISI 1040 parts welded by friction welding', *Mater. Charact.*, 2007, **58**, 1033–1038.
52. M. Sahin, H. E. Akata and K. Ozel: 'An experimental study on joining of severe plastic deformed aluminium materials with friction welding method', *Mater. Des.*, 2008, **29**, 265–274.
53. H. C. Dey, M. Ashfaq, A. K. Bhaduri and K. P. Rao: 'Joining of titanium to 304L stainless steel by friction welding', *J. Mater. Process. Technol.*, 2009, **209**, (18), 5862–5870.
54. P. Sammaiah, A. Suresh and G. R. N. Tagore: 'Mechanical properties of friction welded 6063 aluminum alloy and austenitic stainless steel', *J. Mater. Sci.*, 2010, **45**, (20), 5512–5521.
55. B. Lang, T. C. Zhang, X. H. Li and D. L. Guo: 'Microstructural evolution of a TC11 titanium alloy during linear friction welding', *J. Mater. Sci.*, 2010, **45**, 6218–6224.
56. W. H. Zhu and J. H. Wang: 'Experimental study of inertia friction welding parameters for GH4169 alloy', *Aviation Engineering & Maintenance*, 1994, **5**, 23–24. (in Chinese)
57. H. J. Xu, L. X. Yin, Z. Y. Wei, M. Xie and T. C. Zhang: 'Microstructures and properties of Ti17 alloy inertia friction welded joints', *Trans. of The China Welding Institution*, 2009, **12**, 89–92. (in Chinese)
58. F. Rotundo, A. Morri and L. Ceschini: 'Linear friction welding of a 2024 Al Alloy: microstructural, tensile and fatigue properties', in *'Light Metals 2012'*, (ed C. E. Suarez), 493–496; 2012, Hoboken, NJ, USA, John Wiley & Sons, Inc.
59. A. Chamanfar, M. Jahazi, J. Gholipour, P. Wanjara and S. Yue: 'Maximizing the integrity of linear friction welded Waspaloy', *Mater. Sci. Eng. A*, 2012, **555**, 117–130.

60. E. C. Dalgaard: 'Evolution of microstructure, microtexture and mechanical properties in linear friction welded titanium alloys', PhD thesis, McGill University, Montreal, Canada, 2011.
61. M. Karadge, M. Preuss, C. Lovell, P. J. Withers and S. Bray: 'Texture development in Ti-6Al-4V linear friction welds', *Mater. Sci. Eng. A*, **2007**, **459**, (1–2), 182–191.
62. J. Romero, M. M. Attallah, M. Preuss, M. Karadge and S. E. Bray: 'Effect of the forging pressure on the microstructure and residual stress development in Ti-6Al-4V linear friction welds', *Acta Mater.*, **2009**, **57**, 5582–5592.
63. M. M. Attallah, M. Preuss, C. Boonchareon, A. Steuwer, J. E. Daniels, D. J. Hughes, C. Dungey and G. J. Baxter: 'Microstructural and residual stress development due to inertia friction welding in Ti-6246', *Metall. Mater. Trans. A*, **2012**, **43**, (9), 3149–3161.
64. E. Dalgaard, P. Wanjara, J. Gholipour, X. Cao and J. J. Jonas: 'Linear friction welding of a near- β titanium alloy', *Acta Mater.*, **2012**, **60**, 770–780.
65. W. Y. Li, X. Y. Wang, S. S. Zhang, G. D. Wen, T. J. Ma and A. Vairis: 'Development of microstructure and texture in linear friction welded pure titanium', Proc. 67rd Int. Conf. on 'Advance technology in welding and joining for heavy, automotive and electronics industries', Seoul, Korea, July 2014, Sheraton Grande Walkerhill & W Seoul Walkerhill.
66. M. B. Uday, M. N. Ahmad Fauzi, H. Zuhailawati and A. B. Ismail: 'Advances in friction welding process: a review', *Sci. Technol. Weld. Joi.*, **2010**, **15**, (7), 534–558.
67. R. Damodaram, S. G. S. Raman and K. P. Rao: 'Effect of post-weld heat treatments on microstructure and mechanical properties of friction welded alloy 718 joints', *Mater. Des.*, **2014**, **53**, 954–961.
68. Z. W. Huang, H. Y. Li, G. Baxter, S. Bray and P. Bowen: 'Electron microscopy characterization of the weld line zones of an inertia friction welded superalloy', *J. Mater. Process. Technol.*, **2011**, **211**, 1927–1936.
69. J. W. L. Pang, M. Preuss, P. J. Withers, G. J. Baxter and C. Small: 'Effects of tooling on the residual stress distribution in an inertia weld', *Mater. Sci. Eng. A*, **2003**, **356**, 405–413.
70. M. Preuss, P. J. Withers, J. W. L. Pang and G. J. Baxter: 'Inertia welding nickel-based superalloy: Part I. Metallurgical characterization', *Metall. Mater. Trans. A*, **2002**, **33**, (10), 3215–3225.
71. A. Chamanfar, M. Jahazi, J. Gholipour, P. Wanjara and S. Yue: 'Mechanical property and microstructure of linear friction welded WSPALLOY', *Metall. Mater. Trans. A*, **2011**, **42**, (3), 729–744.
72. O. T. Ola, O. A. Ojo, P. Wanjara and M. C. Chaturvedi: 'A study of linear friction weld microstructure in single crystal CMSX-486 superalloy', *Metall. Mater. Trans. A*, **2012**, **43**, 921–933.
73. M. Karadge, M. Preuss, P. J. Withers and S. Bray: 'Importance of crystal orientation in linear friction joining of single crystal to polycrystalline nickel-based superalloys', *Mater. Sci. Eng. A*, **2008**, **491**, 446–453.
74. W. Y. Li, T. J. Ma, Y. Zhang, Q. Z. Xu, J. L. Li, S. Q. Yang and H. L. Liao: 'Microstructure characterization and mechanical properties of linear friction welding Ti-6Al-4V alloy', *Adv. Eng. Mater.*, **2008**, **10**, (1–2), 89–92.
75. T. J. Ma, W. Y. Li, Q. Z. Xu, Y. Zhang, J. L. Li, S. Q. Yang and H. L. Liao: 'Microstructure evolution and mechanical properties of linear friction welded 45 steel joint', *Adv. Eng. Mater.*, **2007**, **9**, (8), 703–707.
76. S. Fukumoto, H. Tsubakino, K. Okita, M. Aritoshi and T. Tomita: 'Amorphization by friction welding between 5052 aluminum alloy and 304 stainless steel', *Scripta Mater.*, **2000**, **42**, 807–812.
77. E. Taban, J. E. Gould and J. C. Lippold: 'Characterization of 6061-T6 aluminum alloy to AISI 1018 steel interfaces during joining and thermo-mechanical conditioning', *Mater. Sci. Eng. A*, **2010**, **527**, 1704–1708.
78. M. Yilmaz, M. Col and M. Acet: 'Interface properties of aluminum/steel friction-welded components', *Mater. Charact.*, **2003**, **49**, 421–429.
79. M. Ashfaq, N. Sajja, H. K. Rafi and K. P. Rao: 'Improving strength of stainless steel/aluminum alloy friction welds by modifying faying surface design', *J. Mater. Eng. Perform.*, **2013**, **22**, (2), 376–383.
80. M. G. Reddy, S. A. Rao and T. Mohandas: 'Role of electroplated interlayer in continuous drive friction welding of AA6061 to AISI 304 dissimilar metals', *Sci. Technol. Weld. Joi.*, **2008**, **13**, (7), 619–628.
81. I. Bhamji, R. J. Moat, M. Preuss, P. L. Threadgill, A. C. Addison and M. J. Peel: 'Linear friction welding of aluminium to copper', *Sci. Technol. Weld. Joi.*, **2012**, **17**, (4), 314–320.
82. P. Wanjara, E. Dalgaard, G. Trigo, C. Mandache, G. Comeau and J. J. Jonas: 'Linear friction welding of Al-Cu: Part 1-process evaluation', *Can. Metall. Q.*, **2011**, **50**, (4), 350–359.
83. E. Dalgaard, P. Wanjara, G. Trigo, M. Jahazi, G. Comeau and J. J. Jonas: 'Linear friction welding of Al-Cu: Part 2-interfacial characteristics', *Can. Metall. Q.*, **2011**, **50**, (4), 360–370.
84. I. Bhamji, M. Preuss, R. J. Moat, P. L. Threadgill and A. C. Addison: 'Linear friction welding of aluminium to magnesium', *Sci. Technol. Weld. Joi.*, **2012**, **17**, (5), 368–374.
85. T. J. Ma, B. Zhong, W. Y. Li, Y. Zhang, S. Q. Yang and C. L. Yang: 'On microstructure and mechanical properties of linear friction welded dissimilar Ti-6Al-4V and Ti-6.5Al-3.5Mo-1.5Zr-0.3Si joint', *Sci. Technol. Weld. Joining*, **2012**, **17**, (1), 9–12.
86. Y. N. Guo, Y. L. Chiu, M. M. Attallah, H. Y. Li, S. Bray and P. Bowen: 'Characterization of dissimilar linear friction welds of α - β titanium alloys', *J. Mater. Eng. Perform.*, **2012**, **21**, 770–776.
87. W. Y. Li, H. Wu, T. J. Ma, C. L. Yang and Z. W. Chen: 'Influence of parent metal microstructure and post-weld heat treatment on microstructure and mechanical properties of linear friction welded Ti-6Al-4V joint', *Adv. Eng. Mater.*, **2012**, **14**, (5), 312–318.
88. W. Y. Li, J. D. Suo, T. J. Ma, Y. Feng and K. Kim: 'Abnormal microstructure in the weld zone of linear friction welded Ti-6.5Al-3.5Mo-1.5Zr-0.3Si titanium alloy joint and its influence on joint properties', *Mater. Sci. Eng. A*, **2014**, **599**, 38–45.
89. O. T. Ola, O. A. Ojo, P. Wanjara and M. C. Chaturvedi: 'Crack-Free welding of IN 738 by linear friction welding', *Adv. Mater. Res.*, **2011**, **278**, 446–453.
90. O. N. Senkov, D. W. Mahaffey, S. L. Semiatin and C. Woodward: 'Inertia friction welding of dissimilar superalloys Mar-M247 and LSHR', *Metall. Mater. Trans. A*, **2014**, **45**, 5545–5561.
91. M. Karadge, B. Grant, P. J. Withers, G. Baxter and M. Preuss: 'Thermal relaxation of residual stresses in nickel-based superalloy inertia friction welds', *Metall. Mater. Trans. A*, **2011**, **42**, (8), 2301–2311.
92. N. Iqbal, J. Rolph, R. Moat, D. Hughes, M. Hofmann, J. Kelleher, G. Baxter, P. J. Withers and M. Preuss: 'A comparison of residual stress development in inertia friction welded fine grain and coarse grain nickel-base superalloy', *Metall. Mater. Trans. A*, **2011**, **42**, (13), 4056–4063.
93. M. Preuss, P. J. Withers and G. J. Baxter: 'A comparison of inertia friction welds in three nickel base superalloys', *Mater. Sci. Eng. A*, **2006**, **437**, (1), 38–45.
94. R. J. Moat, D. J. Hughes, A. Steuwer, N. Iqbal, M. Preuss, S. E. Bray and M. Rawson: 'Residual stresses in inertia-friction-welded dissimilar high-strength steels', *Metall. Mater. Trans. A*, **2009**, **40**, (9), 2098–2108.
95. X. Song, M. Xie, F. Hofmann, T. S. Jun, T. Connolly, C. Reinhard, R. C. Atwood, L. Connor, M. Drakopoulos, S. Harding and A. M. Korsunsky: 'Residual stresses in linear friction welding of aluminium alloys', *Mater. Des.*, **2013**, **50**, 360–369.
96. R. Turner, R. M. Ward, R. March and R. C. Reed: 'The magnitude and origin of residual stress in Ti-6Al-4V linear friction welds: an investigation by validated numerical modeling', *Metall. Mater. Trans. B*, **2012**, **43**, 186–197.
97. M. Preuss, J. Q. Fonseca, A. Steuwer, L. Wang, P. J. Withers and S. Bray: 'Residual stresses in linear friction welded IMI550', *J. Neutron Res.*, **2004**, **12**, (1–3), 165–173.
98. M. Şahin and H. E. Akata: 'Joining with friction welding of plastically deformed steel', *J. Mater. Process. Technol.*, **2003**, **142**, (1), 239–246.
99. Y. N. Guo, T. Jung, Y. L. Chiu, H. Y. Li, S. Bray and P. Bowen: 'Microstructure and microhardness of Ti6246 linear friction weld', *Mater. Sci. Eng. A*, **2013**, **562**, 17–24.
100. F. Rotundo, L. Ceschini, A. Morri, T. S. Jun and A. M. Korsunsky: 'Mechanical and microstructural characterization of 2124Al/25 vol.%SiCp joints obtained by linear friction welding (LFW)', *Compos. Part A*, **2010**, **41**, 1028–1037.
101. A. Haşcalik, E. Ünal and N. Özdemir: 'Fatigue behaviour of AISI 304 steel to AISI 4340 steel welded by friction welding', *J. Mater. Sci.*, **2006**, **41**, (11), 3233–3239.
102. H. Y. Li, Z. W. Huang, S. Bray, G. Baxter and P. Bowen: 'High temperature fatigue of friction welded joints in dissimilar nickel based superalloys', *Mater. Sci. Technol.*, **2007**, **23**, (12), 1408–1418.
103. G. D. Wen, T. J. Ma, W. Y. Li, J. L. Li, H. Z. Guo and D. L. Chen: 'Cyclic deformation behavior of linear friction welded Ti6Al4V joints', *Mater. Sci. Eng. A*, **2014**, **597**, 408–414.

104. P. Sathiy, S. Aravindan and A. N. Haq: 'Effect of friction welding parameters on mechanical and metallurgical properties of ferritic stainless steel', *Int. J. Adv. Manuf. Technol.*, **2007**, **31**, (11–12), 1076–1082.
105. N. Arivazhagan, S. Singh, S. Prakash and G. M. Reddy: 'An assessment of hardness, impact strength, and hot corrosion behaviour of friction-welded dissimilar weldments between AISI 4140 and AISI 304', *Int. J. Adv. Manuf. Technol.*, **2008**, **39**, (7–8), 679–689.
106. V. V. Satyanarayana, G. M. Reddy and T. Mohandas: 'Dissimilar metal friction welding of austenitic-ferritic stainless steels', *J. Mater. Process. Technol.*, **2005**, **160**, 128–137.
107. G. M. Reddy and P. V. Ramana: 'Role of nickel as an interlayer in dissimilar metal friction welding of maraging steel to low alloy steel', *J. Mater. Process. Technol.*, **2012**, **212**, (1), 66–77.
108. T. J. Ma, W. Y. Li, B. Zhong, Y. Zhang and J. L. Li: 'Effect of post-weld heat treatment on microstructure and property of linear friction welded Ti17 titanium alloy joint', *Sci. Technol. Weld. Joi.*, **2012**, **17**, (3), 180–185.
109. M. Corzo, Y. Torres, M. Anglada and A. Mateo: 'Fracture behaviour of linear friction welds in titanium alloys', *Anales de Mecánica de la Fractura*, **2007**, **1**, 75–80.
110. M. D. Tumuluru: 'Sulfide stress corrosion cracking in low-alloy steel inertia friction welds', *Weld. J.*, **1987**, **66**, (3), 61s–72s.
111. M. Hazra, K. S. Rao and G. M. Reddy: 'Friction welding of a nickel free high nitrogen steel: influence of forge force on microstructure, mechanical properties and pitting corrosion resistance', *J. Mater. Res. Technol.*, **2014**, **3**, (1), 90–100.
112. N. Arivazhagan, K. Senthilkumaran, S. Narayanan, K. Devendranath Ramkumar, S. Surendra and S. Prakash: 'Hot corrosion behavior of friction welded AISI 4140 and AISI 304 in K₂SO₄60 wt-% NaCl mixture', *J. Mater. Sci. Technol.*, **2012**, **28**, (10), 895–904.
113. A. Astarita, M. Curioni, A. Squillace, X. Zhou, F. Bellucci, G. E. Thompson and K. A. Beamish: 'Corrosion behaviour of stainless steel-titanium alloy linear friction welded joints: Galvanic coupling', *Mater. Corros.*, **2014**, doi:10.1002/maco.201307476.