Overview of Joining and Repairing Techniques of Ni-Based Superalloy for Industrial Gas Turbine Applications

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Abstract. In industrial gas turbine (IGT) engine manufacturing, nickel-based superalloys are used mainly to meet the needs of components of the hot gas pathway. Although these alloys have high-temperature capabilities, the parts are prone to damage during service. The high working temperatures of these engines lead to component degradation due to creep, fatigue, and oxidation reactions; therefore, due to the high cost of newly produced superalloy components, it is usually more cost-effective to repair the damaged parts rather than completely replacing them. Joining and repairing techniques are necessary when manufacturing and repairing these alloys. This article will present an overview of the Ni-based superalloy for industrial gas turbine application by studying the microstructure of Ni-based superalloy, weldability issues, and cracking phenomena. Joining/repairing techniques of Ni-base superalloy with advantages and limitations to each technique are discussed to know a suitable technique for use in the high-temperature application.

Keywords. Nickel (Ni) based superalloys, Gas turbine, joining, repairing, weldability.

1. Introduction

As shown in Figure 1, industrial gas turbines require using several types of superalloys to ensure stable operation under an aggressive environment, including high temperature and pressure. The materials used should offer superior mechanical properties; excellent mechanical strength, hardness, ductility, and resistance to creep, oxidation, and high-temperature corrosion. Ni-based superalloys IN738 and IN939 are commonly mentioned as materials meeting these requirements to find a pronounced application in blades and vanes for the gas turbines. Nickel-based superalloys are preferred for use in some applications due to their properties. Nickel-based alloys can be either a solid solution or precipitation strengthened. For applications involving only moderate strength, solid solution strengthened alloys such as Hastelloy X are used. A precipitation strengthened alloy is needed in the most demanding applications, such as hot sections of gas turbine engines [1,2].
These superalloys are obtained by precipitation hardening from $\gamma'$ phase Ni$_3$(Al, Ti) and MC type carbides [4]. Superalloys’ exposure to severe operating environments, mostly high temperatures in the hot sections of aero-engines and power-generating turbines. This usually damages the mechanical components, as shown in Figure 2, due to increased creep level, higher thermal fatigue, higher hot corrosion rates, and more oxidation experienced over a prolonged period.

Consequently, repair work is required to prolong these components’ service life and avoid replacement, which is expensive, as shown in Figure 3 [6]. Fusion welding, diffusion bonding, and brazing are three main repairing/joining techniques generally used in industry. The weldability of nickel-based superalloys depends widely on their Al and Ti contents. Three main repair/joining processes are commonly used in manufacturing, such as fusion welding, diffusion bonding, and brazing. The weldability of superalloys based on nickel is mainly dependent on their Al and Ti
contents. Due to its high susceptibility to solidification cracking within the fusion zone and the heat-affected zone (HAZ) during welding or post-weld heat treatment, precipitation hardened of nickel-based superalloys containing high Al and Ti concentrations is very difficult to weld [6-9]. The diffusion bonding and brazing process problems limit the use of these techniques to manufacture or repair superalloy components based on nickel. During the brazing process, brittle boride, silicide, and continuous phosphate phases can be developed in the joint region, and the joint properties can be degraded [10-14]. Transient Liquid Phase (TLP) bonding has evolved to solve the problem as a preferred technique capable of joining several alloys that are difficult to weld due to several technological and economic advantages. This joining technique combines the added benefit of brazing and diffusion bonding [15-17].

A transient liquid phase bonding technique, developed at P&W [15] in the 1970s, was the first diffusion brazing method. The MPD within the interlayer diffuses through the base material at the bonding temperature. This resulted in a compositional change in the braze occurring in the bond area that increased the interlayer and isothermal solidification's solidus. A long diffusion cycle is required to achieve complete isothermal solidification and joint homogenization using TLP in narrow gap brazing. The TLP mechanism is, therefore, limited to narrow gaps or cracks [18]. If a broader joint needs to be filled with a low melting braze alloy, the formation of brittle intermetallics such as borides and silicides will result from the inclusion of excessive MPD. In addition to the brittle nature, the existence of these phases also decreases the remelting temperature of the joint and provides a path for the propagation of cracks, decreasing the resistance to fracture [19].

![Figure 3](image-url)

**Figure 3.** Cracks repairing in the first stage nozzle using a vacuum furnace brazing process [20].

Therefore, wide gap brazing (WGB) techniques to repair cracks up to 1 mm in width have been developed to overcome these limitations. In applications requiring the repair of turbine components for aircraft engines, where wide cracks are rarely encountered, these methods are commonly used. Cracking due to thermal fatigue is one of the most common types of service damage experienced by gas turbine hot section components [19]. In high-temperature structural components such as aero engines and gas turbines, wide-gap brazing has been used extensively to repair the cracks formed on the surface during service[21-23]. Unlike traditional brazing where only the filler metal is used, wide gap brazing uses a mixture of filler metal and additive powder. Additive powders do not melt during wide-gap brazing, thus providing the necessary capillary force to attract and hold the molten filler metal [23-25].
2. Superalloy

Superalloys are extremely important to gas turbine engines. They have exceptional qualities at very high temperatures and harsh environments, and therefore, have become most suitable for use in the aircraft and gas turbine industries [26]. There are three desired characteristics of a superalloy: its ability to withstand loading at operating temperatures close to the melting point, its resistance to mechanical degradation, and its resistance to environmental corrosion. They attain their high strength by solid solution strengthening to form gamma (γ) phase or precipitate strengthening to form gamma (γ′) phase or both [27]. Superalloys are classified as iron-based, nickel-based, and cobalt-based, but the most commonly used superalloys in aircraft and land-based turbines are nickel- and Co-based superalloys [28]. At temperatures of 760 to 890 °C, nickel superalloys are used, while Fe-Ni superalloys are designed for lower 650-815 °C temperatures (depending on the amount of Ni) and achieve lower strength characteristics at both lower and higher temperatures. Cobalt-based alloys have a lower value at medium and lower temperatures than Ni superalloys because they do not contain the γ′ phase. In high-temperature/low-pressure components with high durabilities, such as industrial turbine blades, these alloys are used [29].

2.1. Composition-microstructure relationships in nickel-based superalloys

For high-temperature structural applications, nickel-based superalloys are the material of choice, mainly when creep or fatigue resistance is necessary, and the risk of degradation due to oxidation or corrosion is severe. Depending on the microstructural changes or properties required, Ni-base superalloys could contain ten (sometimes even more) different alloying elements. Al, Ti, and Cr reduce the density and improve Ni-base superalloys' resistance to corrosion, whereas tungsten, rhenium, and tantalum increase the density and strength of the superalloy [30-32]. Through solid solutions and precipitation, nickel-based superalloys can be strengthened. For burner and combustor applications in gas turbine engines, the solid-solution strengthens superalloys, such as Hastelloy X are used. In particular, they have high-temperature resistance to corrosion, excellent fabricability, and weldability, but lower mechanical efficiency. In applications that require high-temperature strength and high corrosion and creep resistance, such as turbine blades and vanes in the gas turbine, for example, precipitation-strengthened alloys, such as IN 738LC and IN939 are constituted [7,28,33]. In Ni-based superalloys, high-temperature strength is ensured by strengthening the solid solution γ (austenitic phase) and strengthening the precipitated γ′-Ni3(Al, Ti) phase by precipitation, then further strengthening by dispersion strengthening by carbide particles [29]. The significant phases that may be present in nickel-base alloys are:

- **Gamma matrix (γ):** The gamma phase is a face-centered-cubic (FCC) nickel-based austenitic solid-solution phase that usually contains a high percentage of solid-solution elements such as cobalt, iron, chromium, molybdenum, and tungsten [34,35]. All nickel-base alloys contain this phase as the matrix.

- **Gamma Prime (γ′):** Aluminum and/or titanium, the essential solutes, are added in amounts and mutual proportions with a total concentration of typically less than 10 wt % in order to precipitate high volume fractions of primitive cubic γ′ [Ni3(Al, Ti)] coherent with the γ matrix. The γ′ phase “Figure 4” is the principal high temperature strengthening phase in most of the Ni-base superalloys [36]. Elements that promote the formation of γ′ include Al, Ti, and Ta. The ordered FCC L12-type Ni3(Al, Ta, Ti) γ′ phase particles have a lattice parameter that is very close to that of the γ matrix, thus enabling the coherent and homogeneous precipitation of the γ′ phase in the γ matrix and imparting stability at high temperatures [36]. Furthermore, the inherent ductility of γ′ prevents it from being a source of fracture. The morphology of the γ′ precipitates largely depends on their mismatch with the γ matrix. Hagel [37] reported that the γ′ exhibits a spherical shape at a lattice mismatch of 0 – 0.2%, then becomes cubic at a mismatch of around 0.5 – 1%, and above 1.25%, they have a plate-like morphology.
**Figure 4.** γ’ precipitates as a cubic morphology in the γ matrix [35].

- **Gamma double prime (γ’’):** It consists of nickel and niobium and forms a system of body-centered Tetragonal (BCT) Ni₃Nb [34]. Although containing large mismatch strains on the order of 2.9%, it can be coherent with the gamma matrix. This phase is responsible for very high strength at low to moderate temperatures but unstable at temperatures above about 650 °C. Most of this precipitate is present in nickel-iron alloys and Nb-rich alloys containing a low Al and Ti concentration [34,38].

- **Carbides:** Carbon (0.05 - 0.2 wt %) combines with refractory and reactive elements such as titanium, and chromium, to form MC carbides, as shown in Figure 5, which is after heat treatment and service generate lower carbides such as M₂₃C₆ and M₆C at grain boundaries [39].

**Figure 5.** SEM micrograph showing γ’ precipitates and MC carbides in as-cast IN738 [40].

- **Borides:** Carbon and boron are added solid solutions for grain boundary strengthening (and are only used with polycrystalline alloys). Elements such as Cr or Mo can combine with boron to form borides found in the γ grain boundaries, a relatively low density of boride particles forms when boron segregates to grain boundaries [34,41].

- **Grain Boundary γ’:** Heat treatments and service exposures generate a film of γ’ along the grain boundaries, improving the creep-rupture properties [34].
Other phrases that can be found in some superalloys, mostly in the service or aged condition, but these phases are carefully avoided by proper selection of superalloy composition, for example, the μ, σ, and Laves phases which cause brittleness and have detrimental effects on their mechanical properties [28,42-44].

2.2. Weldability issues and cracking phenomena of nickel-based superalloys

For industrial gas turbine engines, the continued drive for improved efficiency, performance, and reduced costs requires the expanded use of materials with high strength-high temperature capabilities, such as nickel-based superalloys. These materials must be capable of being appropriately welded to meet the requirements of component design and manufacturing engineers. The weldability is the ability of a metal or combination of metals to welded under manufacturing conditions into a particular suitably designed structure and to achieve satisfactory function in the intended service, as described by the American Welding Society (AWS) [45]. Many nickel-based superalloys are poorly weldable. Relatively high amounts of aluminum and titanium that are important to materials’ mechanisms for strengthening (γ’ formers) have big problems when traditional welding processes join the alloys. Age hardening in the heat-affected zone causes strain age cracking. Some describe them as unweldable due to the sensitivity of these alloys to microcrack during and after welding. Due to the wide use of welding for the manufacture, repair of hot section parts of aero and industrial gas turbine engines made from these materials, the weldability of heat resistant nickel-base superalloys has gained importance [46-49].

Based on the alloying content, the weldability of a Ni-base alloy differs significantly. Strengthened solid-solution alloys such as Hastelloy-X and IN625 are considered weldable and can be combined without problems with traditional Tungsten Inert Gas (TIG) or laser welding methods if suitable welding procedures and pre-and post-treatment are used. Precipitation-hardened low-volume fraction alloys such as IN718 and Waspaloy are more prone to cracking and are typically welded in solution-annealed condition. However, These alloys are, therefore, also known to be weldable. The alloys with high γ-volume fractions, such as IN939, IN738, and Mar-M247, are called "non-weldable" or "difficult-to-weld." These are higher-performance alloys and are used in essential components of the hot-gas path so that flexibility would be highly useful in their manufacture or repair by welding or additive manufacturing [50].

Depending on the temperature range within which they occur, the cracking processes occurring in nickel-based superalloys can be classified into various classes. Cold cracking is not usually directly connected to the welding process and includes, for example, hydrogen breaking at room temperature. In the heat-affected zone (and fusion zone during multi-pass welding operations), warm cracking phenomena, including strain age cracking and ductility dip cracking, occur at high temperatures but do not require a liquid phase to be present. Therefore, they are also referred to as solid-state cracking. Hot cracking happens at high temperatures, which involve the presence of liquid phases, with the possible crack formation in both the fusion zone (FZ) and heat-affected zone (HAZ) [44].

2.2.1. Solidification cracking: Solidification cracking is also referred to as weld-centreline cracking and falls within the super-set of “hot cracking” as it happens directly after welding. The solidification cracking mechanism is shown in Figure 6. Solidization cracking must be used carefully as a general term, as it concerns a particular cracking mechanism. The recently deposited weld metal solidifies from the weld pool’s sides inwards, and the centreline is the last part of freezing. The liquid film is broken up by the tensile stresses created by the shrinkage of solidification, creating a crack in the center of the weld fusion zone [50,51].
Grain boundary strengthening elements such as C, B, and Zr increase the susceptibility to cracking solidification as these elements separate after melting into the inter-dendritic regions and form the liquid film of the low-melting-point. P and S have the same influence and must be kept to the lowest possible levels since the alloy does not benefit from them. Alloying additions that expand the range of solidification would facilitate the solidification cracking [52]. Therefore, solute diffusion is an essential factor in solidification cracking since it influences both the solidification temperature range at the solid-liquid interface and the eutectic type phase forming. The existence of a solidification temperature range results in the formation of a solid+liquid zone called the mushy zone. It is within the mushy zone that the terminal liquid film forms, and a solidification crack propagates. The longer the mushy zone exists, the greater the window for solidification cracking is. The longer the mushy region exists, the wider the window for cracking solidification. Therefore, higher solidification rates can prevent the solidification cracking process, but this is strongly dependent on the alloy elements' diffusion rates. As a general indicator, the greater a nickel superalloy’s solidification temperature range, the more sensitive it is to cracking solidification [53].

2.2.2. Liquation cracking: Another form of hot cracking is Liquation Cracking, which occurs in the heat-affected welds zone in grain boundaries. HAZ liquation cracking is localized to the partially melted zone, usually referred to as PMZ, in contrast to weld solidification cracking, which occurs in the FZ of the weld. This region is located in the HAZ and is close to the FZ. Liquation cracking occurs when the heat from a weld run melts phases of low-melting-point formed on grain boundaries in the heat-affected region. A film of liquid is created on these grain boundaries, which is then pulled apart by the tensile thermal stresses of the weld solidification. In Ni-base superalloys, phases that are likely to be liquated include MC carbide, M6C carbides, the Laves process, and σ-phase [50]. Lower heat input can reduce the HAZ width, thus subjecting fewer materials to the temperature and stress conditions, leading to liquation cracking [52]. This is one factor that can be used to reduce this phenomenon of cracking. Another factor in the material's susceptibility to liquation cracking is its grain structure. In the HAZ, a coarse grain structure increases the chances of liquation cracking as grain boundaries are expected to have more segregated element content and are therefore more likely to form low-melting-temperature phases. Fine-grained microstructures are less susceptible to liquation cracking, such as those found in wrought or additively manufactured components. It is assumed that
the addition of Mg and Mn decreases liquation cracking. The elements are C, B, Zr, P, and S in IN939, most likely to form low melting-point phases [50,51].

2.2.3. Strain-age cracking: Strain-Age Cracking “SAC” is a specific phenomenon to precipitation-hardening Ni-base alloys. It is a form of cold cracking or solid-state cracking. This happens not when the material is welding but when re-heated for post-weld heat treatment, and it is often referred to as re-heat cracking or cracking by PWHT. SAC is generally seen in HAZ, and cracks are often intergranular. In the early 1960s and 70s, the general mechanism of SAC was studied by various researchers; however, researchers generally agree that the cracks form because of excessive stress from precipitation of γ' that cannot be accommodated by grain boundaries. There is stress created by the mismatch between the respective lattice parameters when γ' precipitates out of the γ matrix. Residual stresses from the welding process can exacerbate the stress, and the microstructure may have decreased ductility due to carbide precipitation at grain boundaries [50,52].

High SAC susceptibility, according to DuPont et al., is the reason why large γ' volume fraction alloys are considered unweldable. Alloys with the content of Al + Ti greater than 6 wt. % are particularly vulnerable. As a qualitative indicator of SAC susceptibility, the Al plus Ti content was used, and 'Figure 7' shows one such diagram. An important factor is the rate of precipitation, and alloys that precipitate γ' at a faster rate are more sensitive to cracking [50]. The phase γ' that strengthens alloys such as IN718 precipitates out from the matrix at a much slower rate, making SAC much less likely during welding of IN718 and other γ''-strengthened alloys. In fact, IN718 was developed to solve just that problem [52].

![Figure 7. Influence of Ti and Al content on SAC susceptibility [52].](image-url)

2.2.4. Ductility dip cracking. Ductility Dip Cracking (DDC) is a solid-state cracking phenomenon that often occurs in austenitic alloys. However, it has not been commonly recorded for Ni-base superalloys [52]. Also, in materials with low S, P, and B levels, it can occur to resist solidification. The liquation cracking DDC creates an intergranular cracking in the weld metal or HAZ, typically when the material is reheated, for example, when welding in multi-passes. Many materials have high ductility at temperatures near their melting point, which decreases gradually with decreasing temperature. However, some materials exhibit a “ductility-dip,” i.e., a range of temperature between around 0.6Tm and 0.9Tm where their ductility is uncharacteristically low. However, some materials show a "ductility-dip," i.e., a temperature range between about 0.6Tm and 0.9Tm where their ductility is exceptionally limited. DDC cracking happens as the temperature of the material moves into this area of low-ductility. This temperature range is between 800°C and 1150°C for Ni-base alloys. Since DDC
occurs by grain boundary sliding, resistance to DDC is promoted by a zigzag or serrated grain boundary morphology or precipitate-pinned grain boundary microstructure[54].

3. Joining and repair techniques for Ni-base superalloy
Joining techniques used in the Industrial gas turbines, microelectronic industry, and other applications where high performance is required consist of three methods: mechanical, chemical, and physical. The need for joining superalloys can be classified into three categories: Repair joining (pre- and post-service), Primary fabrication, and Joining of dissimilar (single crystal to polycrystalline) materials [55-57]. The compositions and microstructures of superalloys, especially the latest generation alloys, have been designed and developed to meet the demanding elevated temperature service requirements. Likewise, the joining techniques used for these materials must also be carefully tailored to be compatible with these operating temperatures, stresses, and environments [58-60].

Mechanical joining methods involve the use of mechanical fasteners like screws, bolts, and rivets to join materials that can be separated for a limited number of times, while the chemical methods of joining materials depend on chemical reactions to produce the bond to hold materials together, which results in a significant amount of residual stresses between the components that are bonded. Moreover, the physical methods of joining materials depend on the phase transition from liquid to solid. Welding, brazing, soldering, and applying solvent-based adhesives or glues are good examples of how materials can be physically joined. Physical joining methods are usually applied for nickel-based superalloys when manufacturing land-based turbine engines and turbine blades that are used at extremely high temperatures in severe environments because they are prone to rapid degradation due to creep, fatigue, and damage from foreign objects during service [61].

The replacement of these parts is expensive, and manufacturing new parts requires time due to their intricacy. Therefore, the most viable option is to repair them by using a physical joining method to join nickel-based superalloys to degraded areas, including the use of fusion welding, brazing, diffusion bonding, and TLP bonding [60]. Each of these processes will be reviewed, and their advantages, disadvantages, and limitations will be discussed as follows.

3.1. Fusion welding
Fusion welding is the process by which the materials are joined by melting and subsequent solidification of the adjacent areas of two parts. Filler metals can be added to the molten pool, but this is not necessarily applicable to all welding processes. Fusion welding can be classified into different types, for example, Gas Tungsten Arc Welding (GTAW), Submerged Arc Welding (SAW), Gas Metal Arc Welding (GMAW), Shielded Metal Arc Welding (SMAW), Resistance Welding (RW), Electron Beam Welding (EBW) and Laser Welding (LW). Heat sources that are often utilized include electron beams, plasma arcs, electrical resistance, and lasers. During welding operations, the base material remains at a much lower temperature than the weld pool, which results in a sharp temperature gradient between the base material and weld area [62]. In cases where the joint gap is wide or has a varying width, it is common practice to use filler metals with a slightly lower melting point than the component to ensure complete melting and flow of filler metal into the joint [63,64]. Figure 8 shows the general macrostructural components of a typical fusion welded joint.

![Figure 8. Macrostructure of a typical fusion weld.](image-url)
3.1.1. Limitation of fusion welding: Fusion welding is generally used to join various superalloy materials, but its application to Ni-base superalloys containing substantial amounts of Al and Ti has been restricted due to their high susceptibility to HAZ cracking during welding and post welding heat treatments. The rapid heating and cooling cycle, together with the joint's mechanical restraint, can create large amounts of residual stresses in the material. This, in turn, can lead to cracking and distortion. This is because Ni-base superalloys develop large shrinkage stresses due to the γ′ particles’ rapid precipitation during cooling down from the welding temperature and cracking probability is usually related to the volume fraction (Vf) of γ′ particles present in the alloy [65,66]. Additionally, localized melting at the grain boundaries causes thermally induced welding strains and very low ductility in the alloy, which can also induce cracking in the HAZ [12].

Some mechanical properties, like fatigue resistance of the welded joint, are usually lower than those of the base alloy [65]. This has been attributed to the stress concentration generated by the high thermal gradients produced during welding, which result in the distortion of the welded component in the HAZ area, and the difficulty in welding complex geometries since some surfaces may not be easily accessible to heat sources, Consequently, fusion welding is not a suitable process for joining nickel-based superalloys.

3.2. Diffusion bonding

Diffusion bonding is a solid-state joining process by which two nominally flat surfaces can be joined at an elevated temperature by utilizing an applied pressure and time. The diffusion process takes place in several steps, as shown schematically in Figure 9 [66]. The factors that affect diffusion bonding are melt temperature, pressure, time, and surface condition, defined in terms of roughness and determined by asperity or height to the different wavelengths of roughness. The melt temperatures require a range from 0.5 to 0.8 of the absolute melting point of the material. The pressure applied is typically various fractions of the room temperature yield stress to avoid macroscopic deformation. The time to realize diffusion varies from minutes to hours. In order to obtain a parent metal microstructure and properties that are free of gross macroscopic deformations after bonding, the actual values of these variables for a bond are carefully chosen [67].

Superalloys are designed to form a stable oxide layer to enhance their oxidation resistance. However, diffusion bonding suffers from the limitation of being unable to join systems with stable oxide layers readily [68]. Since diffusion bonding relies on interlayer species diffusion into the substrates, it thus requires longer bonding times to attain a perfect bond, eliminating non-bonded regions. This also requires high pressures, longer processing times, and if not conducted carefully, may lead to some deformation in the components during bonding, making the process too expensive for many applications [69]. When complex geometries are involved, it would be difficult to apply uniform bonding pressures normal to the mating surfaces, and it might require expensive and sophisticated tooling.

![Figure 9](image_url)

Figure 9. The sequence of metallurgical stages in the diffusion bonding process [66].
3.3. Brazing
Brazing is a joining process wherein metals are bonded together using a filler metal with a liquidus temperature greater than 450°C but lower than the solid temperature of the base metal [70]. The filler metal melts and wets the joint's faying surfaces by capillary action at the brazing temperature. The molten filler is retained within the joint by surface tension. Metallurgical reactions occur between the molten filler and the base material, causing erosion (melt-back) in the joint's original surfaces. The joined component is then cooled in order to solidify the molten filler metal. Conventional brazing processes typically involve short holding times; therefore, only a limited amount of solid-state diffusion of the solute elements is feasible [62]. The gap width can also determine the wettability of filler metal. With a narrow gap (< 100 μm), the liquid filler metal wets to the substrate and fills the joint through capillary action [70, 71]. With a wide gap (> 500μm), capillary action is not the driving force for filling a joint [71].

3.3.1. Advantages of Brazing: Brazing has many advantages; it is an economical means to join simple, complex, and multi-component assemblies and has excellent stress distribution and heat transfer properties. Also, components with complex geometries and varying thickness can usually be brazed together [72]. Brazing does not promote cracks like welding because there is no dissolution of the precipitate strengtheners [70]. Localized heating, which distorts some sections, will be avoided, unlike those produced during fusion welding. Superalloys have a high affinity to oxidize, and by brazing, oxidation can be prevented by using a vacuum chamber or inert atmospheres [73, 74]. Finally, brazing allows for the joining of dissimilar materials, such as metal to ceramic [70]. Brazing alloys have better corrosion resistance and high temperature creep resistance due to braze alloys' tailored compositions [73].

3.3.2. Limitations of brazing: For superalloys that contain significant amounts of Ti and Al, particularly Ni-base superalloys, brazing has proven to be a better joining technique compared to fusion welding. However, some challenges remain, as Melt-Back: During the metallurgical reaction of the liquid filler metal with the faying surfaces, the diffusion of melting point depressant (MPD) elements into the base alloy reaches a maximum concentration at the liquid-solid interface, which causes melting of this base alloy, known as melt back or base-metal erosion. This causes a reduction in the effective thickness of ductile base alloy and could have serious consequences when thin materials are brazed or the joint is to be used in an impact or vibration load conditions. Also, Formation of intermetallic phases: The short thermal cycle during brazing limits the diffusion of the MPDs into the base alloy, which triggers the formation of second phase precipitates or the so-called intermetallic precipitates (such as borides, phosphides, and silicide’s). The presence of intermetallic precipitates in Ni-base braze alloys used for high-temperature applications is detrimental as they reduce the ductility. These phases are brittle in nature, and when formed in a continuous manner, they could form a path for crack propagation, thereby reducing the joint's toughness [72].

Overcoming these limitations of brazing in the joining of Ni-base superalloys, Scot et al. [75] optimized the brazing process and patented a process that they called “transient liquid phase (TLP) bonding.” The process has received wide acceptance since it was first used by Scot et al. to join heat-resistant alloys, as it holds high promise, especially for alloys susceptible to hot cracking during fusion welding. A review of this process is discussed in the following.

3.4. Transient liquid phase bonding
TLP bonding is a hybrid method of joining similar and dissimilar materials that have been in use for many years and still in use because of their advantages over other joining techniques, especially in the aerospace, power, and micro-electric- industries. TLP bonding produces a very strong joint with an interface free of remnants of the bonding agent. It differs from diffusion bonding because a thin liquid interlayer is formed, which removes the need for a high bonding or clamping force to hold the substrate together during the set up for brazing [15,16,76]. TLP bonding was patented by Daniel
Paulonis, David Scott Duvall, and William Owczarskiin in 1971 to improve the existing bonding methods and overcome the deficiencies of the current bonding techniques in joining superalloys [77]. TLP bonding makes use of the mating surface interlayer that melts temporarily and later resolidified at the bonding temperature to produce a joint with characteristics that resemble those of the base metals. This helps to fabricate joints in parts that are complex in shape with simple tools and a simplified preparation process of mating surfaces. The process is done in a conventional vacuum or argon atmosphere furnace so that many parts can be joined in a cycle of heat treatment. TLP bonding has been successfully used to join similar nickel-based superalloys and dissimilar metals, including nickel to Co alloys [15]. Two different types of interlayers can be used to join base metals in the TLP bonding process: low melting point alloy (Type I) and pure metallic (Type II).

As an interlayer (filler) material, Type I contains melting point depressants (MPD) elements, such as Boron (B), Silicon (Si), and Phosphorus (P) as the alloying elements with a base element that is usually the same as the base metal. The composition of the MPD elements is usually chosen to allow a eutectic or near eutectic alloy composition, in which the base metal has a lower melting point than the joined base metals. Upon heating to the bonding temperature, the interdiffusion of the alloying elements between the solid substrate and liquid interlayer occurs in order to establish a local thermodynamic equilibrium at the base metal/interlayer interface, and thus changes the compositions to an equilibrium value. Upon establishing local equilibrium, the constant diffusion of the alloying species from the interlayer into the base metal begins. This will deplete the near-interface liquid from the MPD and increases its melting temperature, which finally leads to its solidification.

Further diffusion of MPD into the base metal will make the solidified front advance into the liquid interlayer. Since this type of solidification is purely driven by diffusion, it is called isothermal solidification. Completion of the joining process mainly depends on the time that is required to complete isothermal solidification. Long holding times are usually required to complete isothermal solidification because of solid-state diffusion's inherent slow rates. Upon the completion of isothermal solidification, the material may be held at the bonding temperature or other elevated temperatures for more extended periods to homogenize the joined base metals [78-80].

Type II interlayer is made of pure metallic material. For example, the joining of aluminum (Al) base metals can be done using a thin interlayer of pure copper (Cu). In the process of heating up to the bonding temperature and subsequent holding, two small regions of eutectic liquid are produced at the two interlayers/base metal interfaces as a result of the interdiffusion of the Cu and Al elements. Continuation of the interdiffusion will expand the liquid regions to entirely dissolve the interlayer and form a single liquid region. After this point, the process is similar to that with a Type I interlayer [80].

The main stages of the TLP bonding process shown in Figure 10.

Duvall et al. [15] proposed the following three stages as the main stages of the TLP bonding process: 1) base metal dissolution, 2) isothermal solidification, and 3) homogenization of the bonded material. Niemann et al. [81] observed that during the joint's heating up to the bonding temperature, some solid-state diffusion of the MPD solute from the interlayer into the base material occurs. They concluded that the heating stage could be considered one of the main stages of the TLP bonding process. Finally, Zhou et al. [82] reclassified the TLP bonding process into four main stages.

Stage I: Heating: During this stage, the assembly to be bonded is heated to the bonding temperature, TB, which is generally above the filler alloy's melting point TM. The extent of the diffusion mainly depends on the heating rate and the eutectic temperature of the interlayer. The heating rate should be as high as possible for Type I interlayers. Thus, the filler alloy melts and fills the joint. Before the interlayer reaches the melting point during the heating stage, some solid-state diffusion may occur between the filler alloy and the base metal. The amount of diffusion will depend on several factors: the heating rate and the diffusivity of the MPD element[83].

Stage II: Base Metal Dissolution and Widening: Base metal dissolution and widening start by heating the interlayer from the melting point to the bonding temperature. It then continues as isothermal
dissolution occurs at the bonding temperature. For a Type I interlayer, local thermodynamic equilibrium has to be maintained at the solid/liquid interfaces upon melting. To reach this equilibrium, the base metal is dissolved into the liquid to decrease the MPD concentration to the corresponding liquidus value. The interface at the base metal side has a solidus value of that of the MPD concentration. For a Type II interlayer, the melting and dissolution do not take place as quickly as when a Type I interlayer is used. Instead, Stage II can be divided into two distinct parts. In the first part (Stage II-a), the interdiffusion of the interlayer and base metal elements will lead to the formation of a small eutectic liquid regions at both solid/interlayer interfaces. In the second part (Stage II-b), small liquid regions form in the interlayer, then completely dissolve the interlayer, and then form a single liquid layer. Then, the rest of Stage II with a Type II interlayer is the same as that with a Type I interlayer [80, 70].

**Stage III: Isothermal Solidification at Bonding Temperature:** After maintaining local equilibrium at the interfaces, the base metal will no longer dissolve. Then, due to the difference in MPD solutes' chemical potentials in liquids and solids, they will continuously diffuse into the base metal. As a result, the liquid regions near the interface are depleted from the solutes, and consequently, solidification takes place by the interface growing into the liquid. Since the solidification is solely driven by diffusion and takes place at a constant temperature, it is called isothermal solidification. It should be noted that unlike the dissolution stage, the driving force for solute diffusion and isothermal solidification in this stage is the final thermodynamic state of the system, which in the case of a sufficiently thin interlayer, would be a solid solution of the base metal and MPD solute. To complete the isothermal solidification, often long holding times at the bonding temperature is required [70].

**Stage IV: Homogenization:** Following the isothermal solidification stage, a homogenization process is carried out at a temperature different from the bonding temperature. During homogenization, the remainder of the MPD solute diffuses out of the joint, and its concentration is decreased while other alloying elements diffuse from the base alloy into the joint. Ideally, at the end of the homogenization process, the joint will be identical in chemistry and microstructure to the base alloy. After homogenization is completed, there can be a tolerable amount of the MPD solute that remains in the joint, which generally depends on the material and the intended application of the repaired part, and its practicality homogenization treatment [84].

*Figure 10. Stages of TLP bonding - Interlayer melting, and substrate dissolution [76, 17].*
Terminology:
A = substrate
CA = Composition of the substrate
CE = Initial (eutectic)composition of the liquid
T = Temperature
TB = Bonding temperature
Cs = Solidus composition
CL = Liquidus composition
MPD = Melting point depressant
CR = Room temperature solubility of MPD in A, Figure 11.

Figure 11. Stages of TLP bonding Isothermal Solidification, and Solid-State Homogenization [76, 17]. The terminology is the same as in Figure 10.

3.4.1. Advantages of TLP bonding. Generally, difficult to weld alloys, including Ni-based superalloys which contain high amounts of Al and Ti, have been successfully joined by employing TLP bonding due to advantages for this method as follows:

1. Compared to diffusion bonding, TLP bonding has the advantage of not requiring relatively high pressure, such as the typically involved in a solid-state diffusion bonding process [76]. Moreover, it can be suitably employed to join intermetallic base materials, which have stable oxide films and are difficult to join by diffusion bonding techniques [85].
2. It can successfully join heat resistant alloys that are inherently susceptible to HAZ cracking during welding or post-weld heat treatment [70]. It is also capable of producing good joints between dissimilar alloy combinations and metal-matrix composites.
3. Complex-shaped parts can be joined by using simple tooling and joint surface preparation.
4. It allows for the mass production of parts, and hence processing costs can be significantly reduced.
5. Joints with microstructural and mechanical properties similar to those of the parent metal can be produced. It is also possible to enhance the joints' quality by employing suitable post-joining heat treatment [82].
3.4.2. **Limitations of TLP bonding.** TLP bonding requires long processing times that can typically be several hours. This is due to the dependence of the isothermal solidification stage on the solid-state diffusion of the solute element from the liquid interlayer into the base metal. Insufficient holding time at the bonding temperature can result in the formation of brittle phases. These brittle phases generally tend to degrade the mechanical and chemical properties of the joint. Also, Diffusion of the MPD solute from the liquid interlayer into the base metal causes the precipitation of second-phase particles at the brazed joint interface. These particles can have adverse effects on the mechanical properties and corrosion resistance of the joint. However, it is essential to note that the advantages outweigh the disadvantages, and most importantly, many of these shortcomings can be overcome by optimizing the bonding parameters during TLP bonding [16].

3.4.3. **Variants of transient liquid phase bonding.** There are some variants of TLP bonding, which will be discussed below:

3.4.3.1. **Temperature gradient transient liquid phase bonding.** Temperature gradient (TG) TLP bonding involves the use of a TG to provide a non-planar bond interface. This will result in a stronger bond because of greater metal to metal contact and the non-polar interface compared to the planar interfaces in conventional TLP bonding [77,86,87].

3.4.3.2. **Wide-gap transient liquid phase bonding.** Wide-gap TLP bonding can be used to bond or repair gaps of 100 μm to 500 μm by using a melting and a non-melting layer or a mixture of powders [77,88]. It is very useful in repairing cracks that have occurred as a result of exposure to extreme environments. The main challenge with wide gap TLP bonding is the long holding time for eliminating the eutectics to achieve complete isothermal solidification. However, the long holding time can be rectified using a composite powder mixture containing a filler alloy powder with MPD and another filler alloy powder without MPD [7].

3.4.3.3. **Active transient liquid phase bonding.** Active TLP bonding is when a nonmetallic material, ceramic, and a metallic material are joined together by using a multi-component interlayer in which one of the constituents of the interlayer reacts with the ceramic while the other will diffuse into the metal to trigger isothermal solidification that allows joining of the two different materials [77,89,90]. However, challenges have been reported in the joining of metals and ceramics due to the differences in their mechanical and thermal expansion coefficients. These differences often cause fractures to occur in the ceramic substrate, but the use of ductile material as the interlayer will relieve the residual stresses and make the joining possible [7, 91].

3.4.3.4. **Partial transient liquid phase bonding.** Partial TLP bonding is mainly used in joining ceramics, and the process and principles are the same as those in conventional TLP bonding. A thick refractory metal/alloy is sandwiched between thin layers of alloys with a low melting point as the interlayer materials and heated to the bonding temperature. Melting or eutectic reactions with the refractory core by the two thin layers produce the liquid that wets each of the ceramic substrates and, at the same time, diffuses into the solid refractory core. The refractory bond is produced by the isothermal solidification of the liquid regions and the joint's homogenization [77].

3.5. **Wide gap brazing**

Wide-gap brazing is often used in the repair of large defects and wide gap cracks. A wide gap is considered to be larger than 500 μm [71] and rebuild large worn surface areas of airfoils. This technique involves using a filler, which is a mixture of filler metal and a high-temperature-melting powder (usually referred to as gap-filler). Wide-gap brazing employs a combination of filler metal and additive material, unlike traditional brazing, where only the filler metal is used. Additive powders do not melt during wide-gap brazing, thus supplying the necessary capillary force to attract and retain the molten filler metal [70]. Activated diffusion healing created by General Electric is used in this type of process.
3.5.1 Activated diffusion healing. In wide gap brazing, it is challenging to keep the filler metal in the braze joint, and, therefore, other methods are used in parallel with TLP bonding. General Electric uses a method known as activated diffusion healing (ADH) or activated diffusion bonding (ADB). Activated diffusion healing uses similar concepts from transient liquid phase bonding, but with the addition of an additive powder. Activated diffusion healing uses a mixture of a low melting temperature powder containing melting point depressants, with a high melting temperature powder, with a chemical composition similar to the substrate. The additive powder serves three purposes: (1) adds surface area for the filler metal to adhere to, keeping the braze in the gap through capillary action; (2) acts as a boron and silicon sink for transient liquid phase bonding; and (3) add additional alloying elements to strengthen the joint [91-93]. Also, with the addition of the additive powder, less filler metal is applied, and therefore fewer brittle phases can form [94]. Filler metals used for ADH are BNi-type filler metals containing Cr, B, and Si, which lower the alloy's melting point [95, 96]. The braze alloy's flowability can be tailored by varying the ratio of filler metal and additive powder [95].

3.5.2. Activated Diffusion Healing Process. There are three stages in the activated diffusion healing process: (1) brazing, (2) diffusion, and (3) aging. The braze alloy is mixed with a binder and applied to the crack. Upon heating up, the binder is baked off, sintering the powder. When the brazing temperature is reached, the filler metal becomes liquid while the additive powder remains solid and suspended in the liquid [94]. The temperature then drops to the diffusion hold, where elements such as boron and silicon diffuse out of the filler metal and into the additive powder and base substrate. This is where transient liquid phase bonding begins till the braze alloy is isothermally solidified [92]. Finally, the temperature drops to the aging hold for solid diffusion homogenization and precipitation growth.

4. Summary
The cost-effective manufacturing of high-performance IGT engines depends on the ability to join a range of nickel-based superalloys. Joining or repairing techniques is very useful in manufacturing and repairing different components with complex shapes in various industries. Different joining methods have been discussed in this review, including the strengths and weaknesses of each of these methods. Welding is one of the most commonly applied techniques for joining structural materials. Unfortunately, the high susceptibility to cracking, predominantly in the heat-affected zone (HAZ), has been a significant weldability problem when using this technique to join superalloy parts containing high contents of Aluminum and Titanium. As such, these superalloys are generally considered difficult-to-weld materials. In addition, diffusion and brazing are two other techniques that have been widely used for various types of advanced alloys in recent decades. However, the essential needs for applying high pressure and temperature and high-quality surface preparation of the samples during diffusion bonding and difficulties in the brazing process (e.g., low melting temperature and shear strength of the joints) limit their industrial applicability. Transient Liquid Phase (TLP) bonding has evolved as a preferred technique capable of joining several difficult to weld alloys due to many technological and economic advantages. This joining technique combines the added advantage of brazing and diffusion bonding. It involves sandwiching an interlayer material containing a melting point depressant element such as Boron, Silicon, or Phosphorus between the substrate materials; such that, after subjecting the assembly to a high temperature, the interlayer melts and forms a liquid phase. After that, a bond region is formed as the liquid disappears isothermally and solidifies as a result of interdiffusion between the interlayer and base materials. This solidified bond zone can exhibit matching chemistry with the substrate. Thus, the problem of weld cracking can be avoided by using the TLP bonding for difficult-to-weld superalloys. TLP process is limited to narrow gap or cracks smaller than 500 μm, whereas a more massive joint has to be entirely filled by a low melting braze alloy, the addition of excessive MPD will result in the creation of brittle intermetallics such as borides and silicides. The existence of these phases also decreases the remelting temperature of the joint and the brittle nature, which provides a path for crack propagation, reducing fracture resistance. Therefore, wide gap brazing (WGB) techniques to repair cracks up to 1 mm in width have been developed to
overcome these limitations. In applications involving the repair of turbine components for aircraft engines, where wide cracks are rarely encountered, these techniques are widely used.

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