ELSEVIER
Materials Science and Engineering A 447 (2007) 125–133
MATERIALS SCIENCE & ENGINEERING A
www.elsevier.com/locate/msea

Transient liquid phase bonding of Inconel 718 and Inconel 625 with BNi-2: Modeling and experimental investigations

M.A. Arafin [a,1], M. Medraj [a,*], D.P. Turner [b,2], P. Bocher [c,3]

[a] *Department of Mechanical and Industrial Engineering, Concordia University, Montreal H3G 1M8, Canada*
[b] *Metallurgical Planning, Pratt & Whitney Canada, Longueuil, Que. J4G 1A1, Canada*
[c] *Départment de Génie Mécanique, École de Technologie Supérieure, Montréal H3C 1K3, Canada*

Received 13 June 2006; received in revised form 25 August 2006; accepted 12 October 2006

Abstract

In this study, a combination of direct experimentation and computational modeling approach was used to predict the time required to complete isothermal solidification during the transient liquid phase bonding of Inconel 718 and 625 superalloys, two most commonly used superalloys in aero-engine hot section components, with nickel based filler alloy, BNi-2. However, unlike conventional modeling, the diffusion of solute atoms was modeled by the Random Walk Modeling technique which can take into account the physical and chemical uncertainties associated with the transient liquid phase bonding experiments. The model equations for migrating solid/liquid interface and solute distribution approaches have been modified and presented in this article. Cumulative probability distribution and probability density function of predicted isothermal solidification times were calculated for different process conditions. The predicted isothermal solidification time range with different confidence levels has been verified with experimental data. Good agreement was observed. The times required for complete isothermal solidification were found to be significantly less than those of other nickel superalloys with different nickel based brazing fillers. Further, significant reduction of holding time was observed with increasing bonding temperature and with decreasing joint gap and no significant grain growth has been observed in the temperature range being investigated (1325–1394 K).
© 2006 Elsevier B.V. All rights reserved.

Keywords: Transient liquid phase bonding; Random Walk Modeling; Inconel 718; Inconel 625; BNi-2

1. Introduction

Inconel 625 and 718 superalloys are extremely versatile austenitic nickel based superalloys with excellent strength and good ductility at very high temperature. Typical applications include aero-engine hot section components, miscellaneous hardware, tooling and liquid rocket components involving cryogenic temperatures. However, like other austenitic nickel based superalloys that contain a substantial amount of Ti and Al, they are highly susceptible to heat affected zone cracking during welding [1,2]. Typical high temperature brazing with nickel based filler alloys, containing boron and silicon as melting point depressants, evolved as an effective way to join these superalloys. However, these melting point depressants form eutectic structures that are extremely hard and contain very brittle intermetallic compounds with nickel and chromium which are detrimental to the mechanical properties of the brazed joint [3–5]. One method to prevent the formation of these deleterious phases is transient liquid phase bonding (TLP), also known as diffusion brazing [6,7]. The TLP bonding process uses a low melting filler alloy to wet the contacting base material and that subsequently solidifies isothermally via a fast diffusing element, e.g. boron. Unlike conventional brazing, the thermal exposure used for the TLP bonding cycle is sufficient to induce isothermal solidification at the bonding temperature [8]. Thus, at a relatively low melting temperature, diffusion brazing produces a joint that has a uniform composition profile, relatively more tolerance to surface oxides, geometrical defects and wide gaps [6,9]. These advantageous features have been exploited in a wide range of applications, from the production and repair of turbine engines in the aerospace industry to the connection of circuit lines in the microelectronic industry [7–9].

* Corresponding author. Tel.: +1 514 848 2424x3146; fax: +1 514 848 3175.
E-mail addresses: ma_arafi@encs.concordia.ca (M.A. Arafin), mmedraj@encs.concordia.ca (M. Medraj), Daniel.P.Turner@pwc.ca (D.P. Turner), Philippe.Bocher@etsmtl.ca (P. Bocher).
[1] Tel.: +1 514 848 2424x7095; fax: +1 514 848 3175.
[2] Tel.: +1 450 647 3342; fax: +1 450 647 2319.
[3] Tel.: +1 514 396 8645; fax: +1 514 396 8530.

0921-5093/$ – see front matter © 2006 Elsevier B.V. All rights reserved.
doi:10.1016/j.msea.2006.10.045

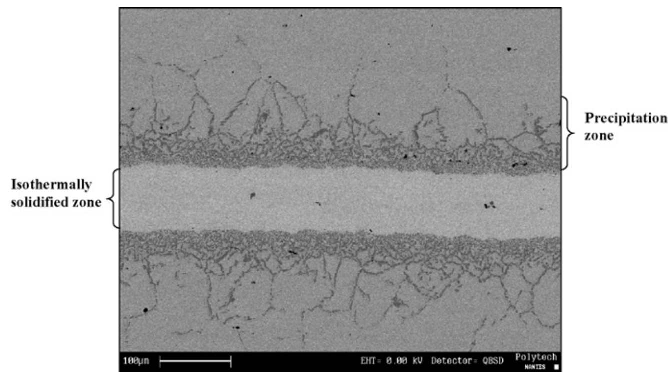

Fig. 1. SEM micrograph of an Inconel 625/BNi-2 joint showing complete isothermal solidification.

For a given operating temperature, TLP bonding process relies on the time required to complete the isothermal solidification to prevent the formation of the brittle eutectic phases in the resulting brazed joints, an example is shown in Fig. 1. Boron composition reached the solidus value during the holding period because of diffusion towards the base metal and thus, the formation of eutectic phases was avoided during cooling.

Tuah-Poku et al. [10] derived an expression for the holding time for silver/copper/silver sandwich joints based on stationary solid/liquid interface and their predicted values were found to be much higher than the experimentally determined values. Lee et al. [11] suggested that diffusion of the solute atoms could take place during liquid homogenization, which could result in the precipitations of second phases in the base metal and thus the holding time required for complete isothermal solidification would be considerably reduced. Other models based on migrating solid/liquid interface and solute distribution law have been used by several researchers [2,4,12–15] to predict the isothermal solidification completion times and the formation of second phase precipitates in the substrates for pure nickel, nickel based single crystal superalloys, Inconel 738 and duplex stainless steel base metals with binary Ni–P and Ni–B, or ternary Ni–Cr–B, or multi-component Ni–Cr–B–Fe–Si filler alloys, and good agreement with the experimental values have been reported. However, modeling studies and experimental investigations of isothermal solidification during TLP bonding of Inconel 625 and 718 superalloys with BNi-2 filler alloy, could not be found in the literature.

Although TLP bonding is an excellent bonding technique, the time required to complete isothermal solidification is usually long enough to discourage their potential applications in many industries. Therefore, a better understanding of the effect of other process variables, such as bonding temperature and joint gap, on the time required to complete isothermal solidification, is imperative to reduce the time requirement and thus to optimize the process. By a combination of direct experimentation with computational modeling, the optimum joining parameters, such as joint gap, bonding temperature and holding time can be set prior to actual field trials.

Mathematical modeling coupled with experimental data is widely used to determine the kinetic parameters such as diffusion coefficient of solute atoms into the base alloys during TLP bonding. However, when coupling experimental data with the mathematical model, the physical and chemical uncertainties associated with the diffusion brazing experiments should be addressed in a way that it best reflects the diffusion characteristics of the solute atoms into the base alloy. Taking only one or two sets of experimental data, often sufficient to solve the governing diffusion equations, will lead to erroneous results because another set of experimental data will result into a different value. Therefore, in such a situation, several sets of experimental data should be used to determine the range of diffusion coefficients and it should be then modeled as a random number based on the statistical distribution profile being observed, such as normal, weibull, uniform, or any other distribution. Such modeling approach is known as Random Walk Modeling and is widely used to simulate the diffusion characteristics of solute atoms in diffusion governing processes [16–19]. However, no such approach has been used so far to simulate the diffusion characteristics of solute atoms into the base alloys during TLP bonding and, single sets of kinetic parameters for diffusion of solute atoms, which is not representative for real life experiments, continue to appear in the literature.

Hence, the objectives of this work are to calculate the time required to complete isothermal solidification during TLP bonding of Inconel 718 and 625 superalloys with BNi-2 filler alloy using mathematical models based on migrating solid/liquid interface and solute distribution law taking the random diffusion of solute atoms into considerations, and to verify the predicted isothermal solidification times with experimental investigations.

2. Experimental investigations

2.1. Procedures

This research was conducted on both wrought Inconel 625 and 718 alloys. The microstructures of the as-received base metals are shown in Fig. 2. Wedge shape joint gap specimens with identical base alloys, shown in Fig. 3, were utilized to form an edge groove where the BNi-2 brazing filler paste was placed. The nominal compositions of the base and filler alloys are given in Table 1. The specimen was fixed by tack welds to form a variable brazing gap (0–250 μm).

The samples were nicro-blasted and then acid cleaned. To prevent the oxide build-up, the base alloy was pre-plated with very thin layer of nickel (nickel flash) and subsequently vacuum brazed at a vacuum pressure of 1.33 mPa (10^{-5} torr) according to the matrix shown in Table 2. The brazed samples were prepared metallographically and studied under the optical and scanning electron microscope (SEM) equipped with electron dispersive spectrometry (EDS).

2.2. Microstructures of the brazed joint

A typical micrograph of the Inconel 625/BNi-2 brazed joint and the corresponding EDS analyses are shown in Fig. 4. Intermetallic phases were formed along the centerline of the joint as

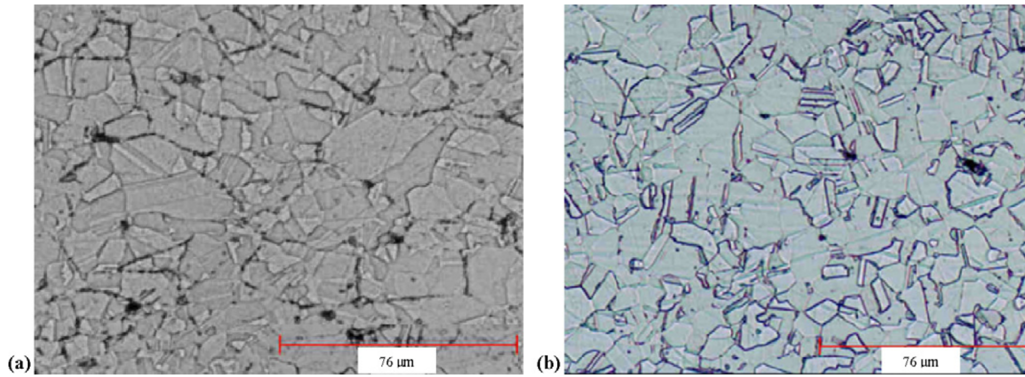

Fig. 2. Microstructures of the as-received wrought: (a) Inconel 625 and (b) Inconel 718.

Table 1
Nominal compositions of Inconel 718, 625 and BNi-2

Alloy	Nominal composition (wt%)	Solidus (°C)	Liquidus (°C)
Inconel 718	Ni (+Co): 50–55%, Cr: 17–21%, Fe: bal, Co: 1%, Mo: 2.8–3.3%, Nb (+Ta): 4.75–5.5%, Ti: 0.65–1.15%, Al: 2–8%, C: 0.8%, Mn: 0.35%, Si: 0.35%, B: 0.006%, Cu: 0.3%	1260	1336
Inconel 625	Ni: 58% (min), Cr: 20–23%, Fe: 5%, Co: 1%, Mo: 8–10%, Nb (+Ta): 3.15–4.15%, Ti: 0.4%, Al: 0.4%, C: 0.1%, Mn: 0.5%, Si: 0.5%	1290	1350
BNi-2	Ni-7Cr-3.2B-4.5Si-3Fe-0.06Cmax	971	999

the part was cooled before the isothermal solidification finished. The residual liquid that was present at the end of the temperature holding eventually transformed on cooling into eutectic constituents. EDS analyses suggest that the phase marked X1 is the pro-eutectic γ-nickel solid solution and the phases marked with X2 and X3 are Cr and Ni rich borides, which are in agreement with the findings of other researchers [2–4,7,20] who worked with nickel based filler alloys containing boron and/or silicon as melting point depressants, with pure nickel or different nickel based superalloys.

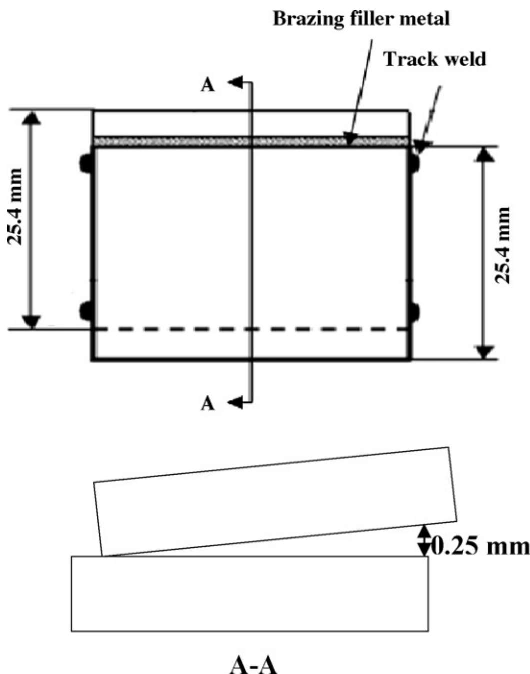

Fig. 3. The wedge shape joint gap specimen.

From the Ni–Si phase diagram [21], it is evident that Ni dissolves an average of 15 mol% Si over the brazing temperature range (1325–1394 K), and thus it is expected to have little or almost no silicides. However, EDS compositional analyses in Fig. 4 revealed a significant amount of silicon in the center of the joint that might form nickel silicides, which is in accordance with the findings of Jang and Shih [5]. This can be understood from the following solidification phenomenon [3]. During brazing, γ-nickel first solidified isothermally from the faying surfaces into the melt. Upon cooling the primary γ-nickel solidified as nodular dendrites which enriched the remaining melt with boron, silicon and chromium. As cooling proceeded, binary eutectic of γ-nickel and nickel boride were encountered, further enriching the melt with chromium. Subsequently, binary eutectic of γ-nickel and chromium boride occurred. The melt, which was further enriched in silicon, was then transformed into the ternary eutectic of γ-nickel, nickel boride and nickel silicides. Similar solidification phenomena are expected for the Inconel 625 and 718 superalloys with BNi-2 filler alloy when the holding time is not long enough to complete isothermal solidification.

Table 2
Braze tests matrix

Temperature (K)	Holding time (min)						
1325	10			50	60	70	
1358			30	50		70	90
1394	10	20	30	50			90

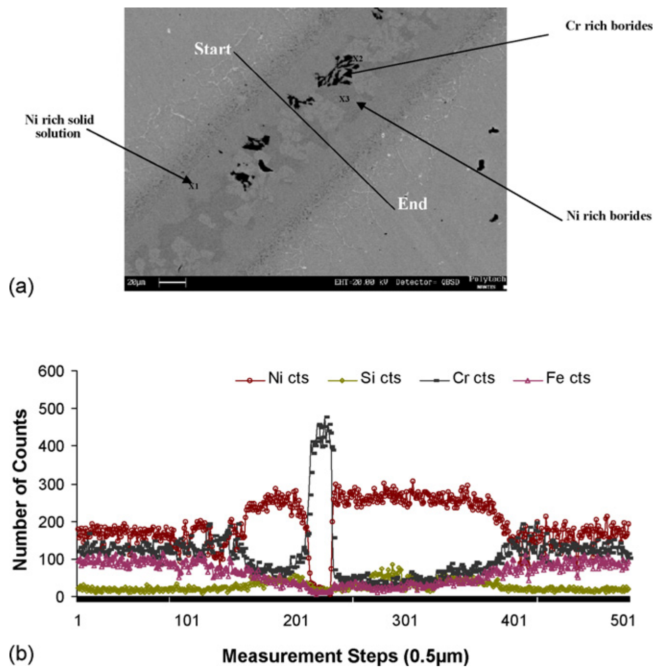

Fig. 4. (a) SEM micrograph of Inconel 625/BNi-2 joint brazed at 1325 K for 10 min showing centerline eutectics, (b) EDS number of counts vs. the measurement step (0.5 μm).

Conventional TLP models assume that solute diffusion in the base metal takes place under equilibrium conditions and thus the formation of boridic precipitates is avoided. However, because of rapid boron diffusion and its low solubility, parent metal close to the brazing gap is quickly oversaturated with boron which produces strong precipitation of borides having a very dense arrangement [22]. This was evident in all brazed joints microstructures, irrespective of brazing temperature, holding time and joint gap. It suggests that diffusion of solute atoms into the base alloy could actually take place during base metal dissolution and liquid homogenization.

2.3. Dissolution of base alloy

The fusing dissolution of base metal is inevitable during transient liquid phase bonding. The main beneficial aspect is that it can enhance the alloying process and thus improves the mechanical properties of the brazed joints [23]. In this work, wedge-shape joint gap specimen model, as shown in Fig. 3, was utilized to account for the dissolution of the base metal and, to enable measuring the maximum brazing clearance since the initial V-configuration is known. The dissolution of the base metal was calculated by measuring the final width of the brazement. Fig. 5 shows the effect of holding time on the dissolution width of Inconel 718 base metal at different brazing temperatures for an initial joint gap of 75 μm. Although dissolution widths were found to be very scattered and difficult to measure due to the physical and chemical uncertainties associated with the TLP bonding experiments, it was obvious that dissolution was very rapid initially but quickly reached the saturation limit at any selected bonding temperature. It was also observed that

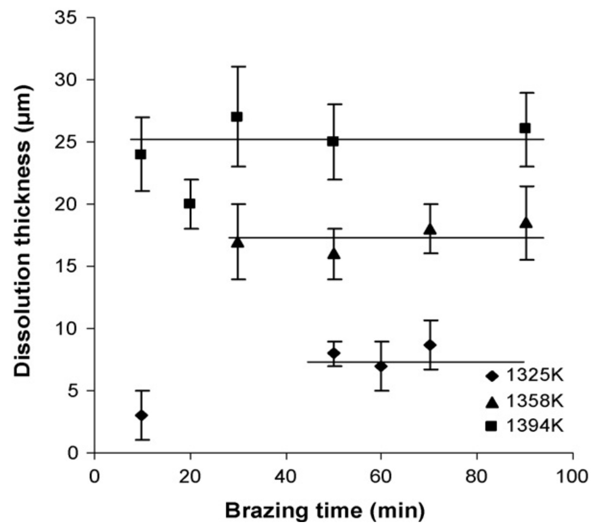

Fig. 5. Effect of bonding temperature and time on dissolution thickness of Inconel 718 base alloy for an initial joint gap of 75 μm.

the saturated dissolution thicknesses of the base metal increased significantly with increasing bonding temperatures.

2.4. Maximum brazing clearances

In the wedge gap brazed joint, a distinction is made between areas free of brittle phase and brittle phase containing seam sections. The beginning of brittle phase stabilization marks the maximum brazing clearance (MBC) for the combination of base metals and filler alloy brazed at a particular temperature and holding time. Figs. 6 and 7 show the maximum brazing clearances for the Inconel 718/BNi-2 and Inconel 625/BNi-2 combinations, respectively, brazed at 1325, 1358 and 1394 K with different holding times ranged from 10 to 90 min. Conversely, if a specified MBC is taken, the corresponding brazing time will represent the isothermal solidification time for that brazing clearance. Significant reduction of holding time has been observed with increasing bonding temperature and/or with decreasing joint gap.

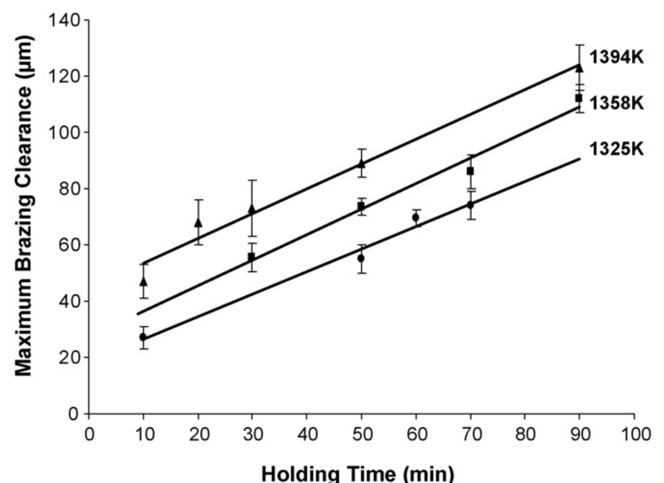

Fig. 6. Effect of holding time on the maximum brazing clearances for Inconel 718/BNi-2.

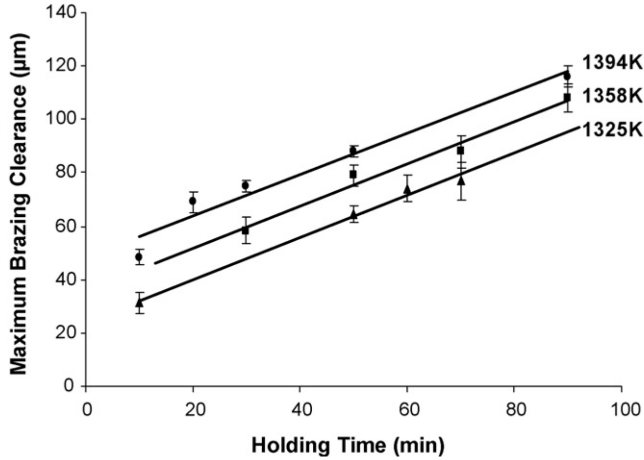

Fig. 7. Effect of holding time on the maximum brazing clearances for Inconel 625/BNi-2.

3. Modeling isothermal solidification time

Isothermal solidification is the most important stage of TLP bonding [24,25]. In this study, a combination of direct experimentation and computational modeling approach was used to solve a diffusion controlled two phase moving boundary problem.

A general error function solution, Eq. (1), can be assumed to calculate the concentration of the solute atom in the solid phase [4,26,27].

$$C(x,t) = A_1 + A_2 \operatorname{erf}\left(\frac{x}{\sqrt{4Dt}}\right) \tag{1}$$

where A_1 and A_2 are constants that can be determined by the boundary conditions. The base metal can be assumed semi-infinite. Therefore, when $x \to \infty$

$$C(\infty,t) = A_1 + A_2 = C_{\mathrm{m}} \tag{2}$$

where C_{m} is the initial solute concentration in the base metal. Let us assume that the solid/liquid interface moves to $x = x_{\mathrm{s}}$ after time t. Hence,

$$C(x_{\mathrm{s}},t) = A_1 + A_2 \operatorname{erf}\left(\frac{x_{\mathrm{s}}}{\sqrt{4Dt}}\right) = C_{\mathrm{s}} \tag{3}$$

where C_{s} is the solute concentration of the solid phase at the solid/liquid interface. The above equation is true for all values of t; therefore, x_{s} must be proportional to $t^{1/2}$:

$$x_{\mathrm{s}} = \gamma\sqrt{4Dt} \tag{4}$$

where γ is a constant that accounts for the moving boundary. The mass balance at the solid/liquid interface gives the following relationship [4,27]:

$$(C_{\mathrm{L}} - C_{\mathrm{s}})\frac{\mathrm{d}x_{\mathrm{s}}}{\mathrm{d}t} = D\left(\frac{\partial C(x,t)}{\partial x}\right)_{x=x_{\mathrm{s}}} \tag{5}$$

where C_{L} is the solute concentration at which the liquid interlayer homogenizes or, in other words, the solute concentration of the liquid phase at the migrating solid/liquid interface. Solving Eqs. (1)–(5) results in the following relation that can be used to determine the dimensionless constant γ [27]:

$$\frac{C_{\mathrm{s}} - C_{\mathrm{m}}}{C_{\mathrm{L}} - C_{\mathrm{s}}} = \gamma\sqrt{\pi}\exp\gamma^2(1 + \operatorname{erf}(\gamma)) \tag{6}$$

Now, from Eq. (4), the time required to complete isothermal solidification can be calculated using the following relationship:

$$t_f = \frac{(2h)^2}{16\gamma^2 D} \tag{7}$$

where $2h$ is the final maximum width of the molten zone.

Unlike stationary solid/liquid interface model, the migrating solid/liquid interface model takes into consideration the moving solid/liquid interface that exists during TLP bonding process. Moreover, it can be coupled with experimental isothermal solidification times to obtain the diffusion coefficients of solute atoms into the base alloys being used. However, it should be noted here that, although migrating solid/liquid interface modeling approach was used by several researchers [2,4,12,27], none of them took into consideration the uncertainties associated with the TLP bonding experiments. All of them reported/used one single set of kinetic parameters, such as activation energy and frequency factor, for each of the combination of base and filler metal being used which has very limited practical value. However, as described earlier, there are several physical and chemical uncertainties associated with the TLP bonding experiments which directly affect the kinetics of the diffusion process, and no single value of diffusion coefficient would be representative for real life TLP bonding experiments. Physical uncertainties include, but are not limited to, (i) variation of nickel layer thickness in the base alloy after nickel flushing, (ii) waviness of the faying surface, (iii) uncertainties in temperature, time and length measurements, (iv) variation of joint configurations, (v) heterogeneous wetting of base alloy by the filler alloy, etc. Chemical uncertainties include, but are not limited to, (i) compositional variations of solute atoms in the filler alloy throughout the joint gap, (ii) heterogeneity in the elemental composition of the base alloys, etc. These uncertainties directly affect the assumption of unidirectional diffusion of solute atoms. Therefore, it is quite obvious that diffusion of solute atoms cannot be modeled assuming an ideal case to predict the time required for complete isothermal solidification; rather, it should be modeled taking the random diffusion of solute atoms into considerations. However, the random numbers should be based on the type of statistical distribution (i.e. normal distribution, weibull distribution, uniform distribution, etc.), being observed for diffusion coefficients, obtained from several sets of experimentally determined isothermal solidification times.

Another approach that has been used to predict the time required to complete isothermal solidification during TLP bonding is the application of solute distribution law. According to Crank [28], for the unsteady state diffusion of a specie present in $2w$ thick region, into a semi-infinite substrate, solute distribution in the substrate is represented by:

$$C_{(x,t)} = C_{\mathrm{m}} + \frac{1}{2}(C_0 - C_{\mathrm{m}})\left\{\operatorname{erf}\frac{x+w}{\sqrt{4Dt}} - \operatorname{erf}\frac{x-w}{\sqrt{4Dt}}\right\} \tag{8}$$

where C_m is the initial solute concentration in the base metal; C_0 is the initial solute concentration in the interlayer; $C_{(x,t)}$ is the solute concentration as a function of distance from the centre of the interlayer (x) and time (t); D is the diffusion coefficient of the solute in the substrate.

Holding time can be estimated considering the fact that isothermal solidification is completed when the solute concentration at the centre of the interlayer is reduced to the solidus value C_s. Substituting $C_{(x,t)} = C_s$ at $x = 0$ yields the following equation:

$$C_s - C_m = (C_0 - C_m)\left\{\mathrm{erf}\frac{w}{\sqrt{4Dt_f}}\right\} \quad (9)$$

Although this is a simple modeling approach which does not take into consideration the dissolution of base metal, sometimes it can be very useful to have a reasonable approximation of holding time required to complete isothermal solidification during TLP bonding. Several researchers [2,13,29] used this approach to predict the isothermal solidification time during TLP bonding and to predict the precipitation of second phases in the substrate. However, again, the uncertainties associated with TLP bonding experiments were not taken into account while predicting the isothermal solidification times. For example, the use of the linear relationships between the eutectic width and square root of holding time to get the extrapolated isothermal solidification times suffers from the drawback that the times used are not the actual ones, and the complexity in measuring exact eutectic width poses a big challenge on the assumption of linear relationship between the eutectic width and square root of holding time. Moreover, the use of only two sets of data to solve the diffusion equations is not representative for real life transient liquid phase bonding experiments that involves numerous physical and chemical uncertainties as mentioned earlier.

An attempt has been made in this study to take into consideration the uncertainties associated with TLP bonding experiments and both the migrating solid/liquid interface model and solute distribution model equations have been modified using Random Walk Modeling technique.

3.1. Modified model equations

The *migrating solid/liquid interface model* equations are thus modified as following:

$$D_{i,j} = \frac{(2h_{i,j})^2}{16\gamma^2 t_{f_{i,j}}} \quad (10)$$

where $i = 1, 2, 3, \ldots, n$, which denotes the number of experimentally determined post brazed maximum joint widths free of eutectic phases and the corresponding holding times. $j = 1325$, 1358, 1394 K, which denotes the bonding temperature.

Diffusion coefficient at a specified bonding temperature can then be written as:

$$D_j = [D_{1,j}, D_{2,j}, D_{3,j}, \ldots, D_{n,j}] \quad (11)$$

Diffusion coefficient for a specified bonding temperature can be written as following:

$$D_{R(j)} = R_{D_j} \quad (12)$$

where R_{D_j} is a random number based on the statistical distribution profile of D_j, as described earlier.

Isothermal solidification time for a specified bonding temperature can then be calculated using the following relation:

$$t_{f_{i,j}}^{1/2} = \frac{1}{4\gamma}\left(\frac{2h_{i,j}}{D_{R(j)}^{1/2}}\right) \quad (13)$$

Since the initial composition of boron in both Inconel 718 and 625 superalloys, $C_m = 0$, *solute distribution model* equations can be modified as following:

$$D_{i,j} = \frac{w_{i,j}^2}{\left(\mathrm{erf}^{-1}(C_s/C_0)\right)^2 \times 4 \times t_{i,j}} \quad (14)$$

where $w_{i,j}$ are the halves of the maximum brazing clearances obtained experimentally and $t_{i,j}$ are the corresponding holding times.

The isothermal solidification time can then be predicted using the following equation:

$$t_{i,j} = \frac{w^2}{\left(2 \times \mathrm{erf}^{-1}(C_s/C_0) \times \sqrt{R_{D_j}}\right)^2} \quad (15)$$

where w is half of the initial joint gap thickness for which the isothermal solidification time is to be calculated.

4. Comparison between models and experiments

4.1. Migrating solid/liquid interface model

Boron has very low solubility in nickel. Ojo et al. [2] reported that the presence of substitutional alloying elements, like Cr and Co, in nickel based superalloys does not affect this solubility significantly and the C_s value can be taken as 0.3 at%, which is the maximum solubility of boron in nickel and remains almost constant in the bonding temperature range. Similarly, the presence of small amount of alloying elements in the nickel based filler alloys is not expected to change the C_L value significantly from that of the Ni–B system which is 16.6 at%. Similar approach was also used by Sakamoto et al. [4] and Rhee et al. [29]. Therefore, γ was calculated by taking C_s and C_L as the average solidus and liquidus boron compositions of the Ni–B system in the bonding temperature range, 0.3 and 16.6 at%, respectively.

Ideally, diffusion coefficient of solute atoms at a specified bonding temperature is constant. However, as described earlier, it is impossible to obtain a constant diffusion coefficient of the solute atoms at the bonding temperature. Range of diffusion coefficients, for both Inconel 718/BNi-2 and Inconel 625/BNi-2 combinations, for each of the three bonding temperatures have been obtained using 28 sets of experimentally determined maximum brazing clearances and the corresponding holding times, and are presented in Tables 3 and 4.

Table 3
Range of diffusion coefficients for Inconel 718/BNi-2 combination

Bonding temperature (K)	Diffusion coefficients ($m^2 s^{-1}$) × 10^{-10}			
	D_{min}	D_{max}	Mean	S.D.
1325	5.82	7.7	6.99	0.84
1358	14.4	17.6	15.5	1.4
1394	22.3	33.6	28.2	5.27

Table 4
Range of diffusion coefficients for Inconel 625/BNi-2 combination

Bonding temperature (K)	Diffusion coefficients ($m^2 s^{-1}$) × 10^{-10}			
	D_{min}	D_{max}	Mean	S.D.
1325	8.43	13.05	10.54	1.05
1358	14.4	18.07	16.5	1.62
1394	21.5	39.5	31.2	8.3

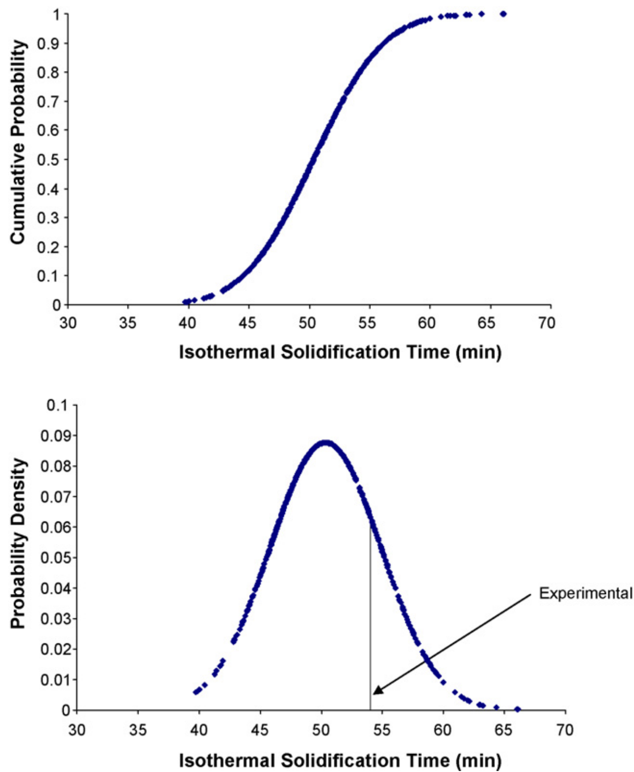

Fig. 8. Cumulative probability plot and probability density plot of isothermal solidification time for Inconel 718/BNi-2 for an initial joint gap of 75 μm and 1358 K bonding temperature.

Numerical simulations were carried out using MATLAB 7.0.1 with modified model equations as discussed earlier. Fig. 8 shows the cumulative probability and probability density of isothermal solidification time for Inconel 718/BNi-2 combination for an initial joint gap of 75 μm at 1358 K bonding temperature. Isothermal solidification time for the process condition has been predicted as a range where different values have different individual probabilities. Cumulative probability distribution is a very useful tool because it is the measure of the probability that isothermal solidification will take place for less than or equal to a given holding time, e.g. a holding time of 60 min would include the probabilities of isothermal solidification times that are less than or equal to 60 min. Therefore, it is a measure of the confidence level that isothermal solidification would take place if the corresponding length of time is elapsed in the furnace at the bonding temperature.

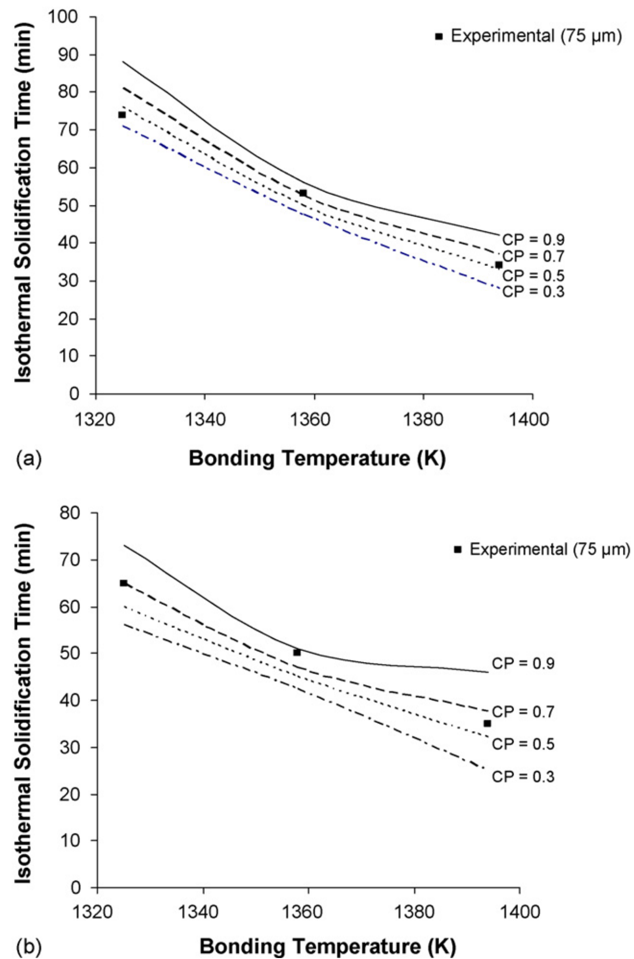

Fig. 9. Comparison of predicted isothermal solidification times with different confidence levels (migrating solid/liquid interface model) with experimental data for an initial joint gap of 75 μm for: (a) Inconel 718/BNi-2 and (b) Inconel 625/BNi-2 [CP = cumulative probability = confidence level].

The predicted isothermal solidification times for three different bonding temperatures with different confidence levels have been compared with experimentally determined values, for an initial joint gap of 75 μm, for both Inconel 718/BNi-2 and Inconel 625/BNi-2, as shown in Fig. 9. It should be noted here that a lower confidence level, such as 50% confidence, is not an indication that the probability of occurrence of that event is lower than that of a higher confidence level. In fact, for a perfectly normal distribution, 50% confidence level values have the highest individual probability or in other words, the maximum likelihoodness.

4.2. Solute distribution law approach

Similar studies were carried out for the solute distribution modeling approach. The value of C_s was taken as 0.3 at% due to the reasons described earlier.

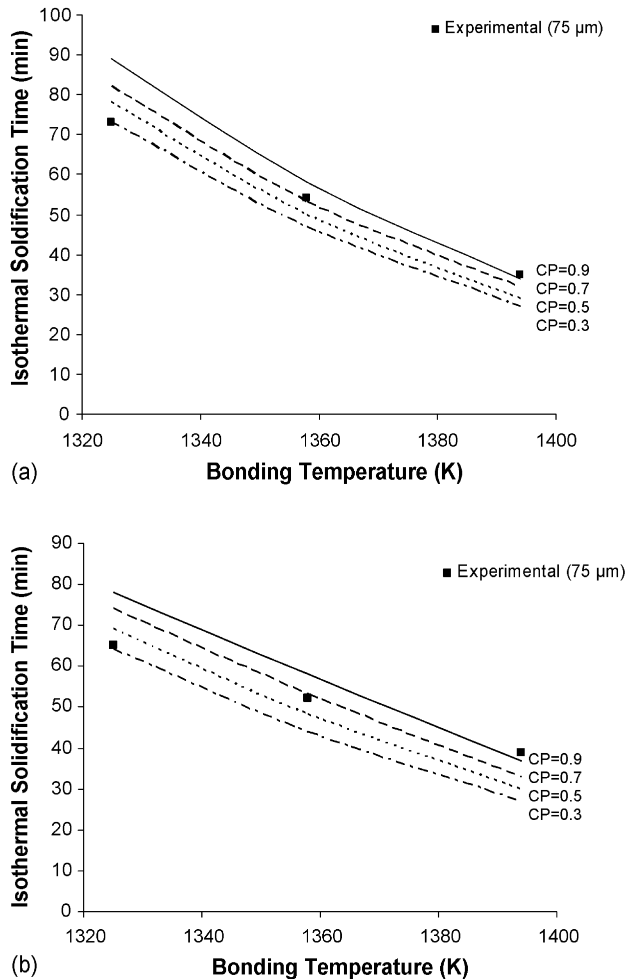

Fig. 10. Comparison of predicted isothermal solidification times with different confidence levels (solute distribution model) with experimental data for an initial joint gap of 75 μm for (a) Inconel 718/BNi-2 and (b) Inconel 625/BNi-2 [CP = cumulative probability = confidence level].

The predicted isothermal solidification times with different confidence levels, for an initial joint gap of 75 μm and for three different operating temperatures, are compared with experimental data in Fig. 10. It was observed that unlike migrating solid/liquid interface model, solute distribution model underestimated the isothermal solidification times at higher temperature (1394 K) bonding operation for both of the combinations. This can be attributed to the fact that solute distribution model does not take into consideration the dissolution of base metal which is significant at higher bonding temperature. It can, therefore, be inferred that although this simple model can be used for a reasonable approximation of isothermal solidification time, modified migrating solid/liquid interface model should be used for better accuracy and reliability.

The constant search for a suitable combination of base and filler alloy with significantly low isothermal solidification time has led to several studies of isothermal solidification times during transient liquid phase bonding of different nickel superalloys with compatible nickel based filler alloys. However, the shortest possible isothermal solidification time reported so far is still in the order of several hours, which is quite impractical for industrial applications. In this study, it was observed that the times required for complete isothermal solidification for both Inconel 718/BNi-2 and Inconel 625/BNi-2 combinations were significantly less, which is of great importance from industrial point of view.

5. Summary and conclusions

The kinetics of isothermal solidification during TLP bonding of Inconel 718 and Inconel 625 superalloys with nickel based filler alloy BNi-2 have been studied through migrating solid/liquid interface modeling and solute distribution law. However, unlike conventional modeling approaches, the diffusion of solute atoms have been modeled using the Random Walk Modeling technique which can take into account the physical and chemical uncertainties associated with TLP bonding experiments. The modified model equations for both of the two modeling approaches have been developed and presented.

Cumulative probability distribution along with probability density functions of isothermal solidification times were calculated for different process conditions and predicted isothermal solidification times with different confidence levels were compared with the experimental results. It was observed that although the modified solute distribution model is useful to have reasonable estimations of isothermal solidification time, modified migrating solid/liquid interface model is more reliable and accurate.

Unlike other currently used combinations, the isothermal solidification completion times for Inconel 718 and 625 with BNi-2 filler alloy were found to be much less. Further significant reduction of holding time was observed with increasing bonding temperature and with decreasing joint gap thickness.

Acknowledgements

The authors wish to thank Prof. René LeGall at Université Polytechnique de Nantes (France) for helpful discussions and accessibility to EDS equipment, Ms. Sharmin Sultana for her valuable suggestions, and both CRIAQ (Center for Research and Innovation in Aerospace in Quebec) and Pratt & Whitney Canada for financial support to conduct this research.

References

[1] W. Chen, M.C. Chaturvedi, N.L. Richards, Metall. Trans. A 32 (2001) 931–939.
[2] O.A. Ojo, N.L. Richards, M.C. Chaturvedi, Sci. Technol. Weld. Joining 9 (2004) 532–540.
[3] S.K. Tung, L.C. Lim, M.O. Lai, Scripta Mater. 34 (1996) 763–769.
[4] A. Sakamoto, C. Fujiwara, T. Hattori, S. Sakai, Weld. J. 68 (1989) 63–71.
[5] J.S.C. Jang, H.P. Shih, J. Mater. Sci. Lett. 22 (2003) 79–82.
[6] C.E. Campbell, W.J. Boettinger, Metall. Trans. A 31 (2000) 2835–2847.
[7] O.A. Ojo, N.L. Richards, M.C. Charturvedi, Sci. Technol. Weld. Joining 9 (2004) 209–220.
[8] W.F. Gale, Mater. Sci. Forum 426–432 (2003) 1891–1896.
[9] C.E. Campbell, U.R. Kattner, Calphad 26 (2002) 477–490.
[10] I. Tuah-Poku, M. Dollar, T.B. Massalski, Metall. Trans. A 19 (1988) 675–686.

[11] C.H. Lee, T.H. North, H. Nakagawa, 71st American Welding Society Convention, Anaheim, CA, 1990, pp. 243–246.
[12] J.E. Ramirez, S. Liu, Weld. J. 71 (1992) 365s–375s.
[13] W.F. Gale, E.R. Wallach, Metall. Trans. A 22 (1991) 2451–2457.
[14] K. Ohsasa, T. Shinmura, T. Narita, J. Phase Equillib. 20 (1999) 199–206.
[15] T. Shinmura, K. Ohsasa, T. Narita, Mater. Trans. 42 (2001) 292–297.
[16] K.N. Dimou, E.E. Adams, Estuar. Coast Shelf S. 37 (1993) 99–110.
[17] V. Nassehi, S. Passone, Environ. Fluid Mech. 5 (2005) 199–214.
[18] A.W. Visser, Mar. Ecol-Prog. Ser. 158 (1997) 275–281.
[19] C.F. Scott, J. Environ. Eng. 123 (1997) 919–927.
[20] X. Wu, R.S. Chandel, H. Li, J. Mater. Sci. 36 (2001) 1539–1546.
[21] T. Tokunaga, K. Nishio, H. Ohtani, M. Hasebe, Mater. Trans. 44 (2003) 1651–1654.
[22] B. Zorc, L. Kosec, Rev. Metal Madrid 36 (2000) 100–107.
[23] X.P. Zhang, Y.W. Shi, Scripta Mater. 50 (2004) 1003–1006.
[24] Y. Zhou, W.F. Gale, T.H. North, Int. Mater. Rev. 40 (1995) 181–196.
[25] W.F. Gale, D.A. Butts, Sci. Technol. Weld. Joining 9 (2004) 283–300.
[26] P.V. Danckwerts, Trans. Faraday Soc. 46 (1950) 701.
[27] Y. Zhou, J. Mater. Sci. Lett. 20 (2001) 841–844.
[28] J. Crank, The Mathematics of Diffusion, second ed., Oxford University Press, Oxford, UK, 1975.
[29] B. Rhee, S. Roh, D. Kim, Mater. Trans. 44 (2003) 1014–1023.