RESIDUAL STRESS PREDICTION AND DETERMINATION IN 7010 ALUMINIUM ALLOY FORGINGS

ABSTRACT

Precipitation hardened aluminium alloys gain their high strength through heat treatment involving a severe quenching operation, which can have the adverse effect of introducing residual stresses. The finite element code ABAQUS is used to simulate the quenching of aluminium alloy 7010 in an attempt to predict the residual stress distribution that develops in simple shapes. The rate of heat transfer from the material is determined by using the finite element method to predict the heat transfer coefficient from surface cooling curves achieved experimentally. The flow stress of the material is assumed to be strain rate dependent and to behave in a perfectly plastic manner. The predicted residual stress magnitudes and directions are compared to values determined using the hole drilling strain gage method and the X-ray diffraction technique.

INTRODUCTION

7010 is an Al-Zn-Mg-Cu alloy used by the aerospace industry in Europe. The alloy has a composition (Table 1), similar to that of 7050, where the wrought structure is grain refined by zirconium dispersoids (Al_3Zr). This has the additional benefit of allowing the retention of high strength levels in thick sections due to a low quench sensitivity. High strength combined with excellent fracture toughness behaviour makes 7010 popular as a structural material for strength critical forgings in aircraft manufacture.

Table 1 Specification composition of the aluminium alloys 7010, 7050 and 7150.

Element (Wt.%)	Zn	Mg	Cu	Zr	Fe	Mn	Si	Cr	Ti
7010	5.7 - 6.7	2.2 - 2.7	1.5 - 2	0.11 - 0.17	0.15	0.30	0.10	0.05	-
7050	5.7 - 6.7	1.9 - 2.6	2 - 2.6	0.08 - 0.15	0.15	0.10	0.12	0.04	0.06
7150	5.9 - 6.9	2.0 - 2.7	1.9 - 2.5	0.08 - 0.15	0.15	0.10	0.12	0.04	0.06

7xxx series alloys attain their high strength via a high temperature (circa 470°C) solution heat treatment followed by a rapid quench into water/organic quenchant and a subsequent artificial ageing treatment. The quenching part of this process sets up severe thermal gradients resulting in a complex residual stress distribution. These internal stresses can cause distortion during machining and provide the driving force for intergranular stress corrosion cracks (SCC). This is especially evident when the residual stresses act in the short transverse direction where mechanical fibring of the grain structure results in lower fracture properties in this direction.

Previous work[1,2,3,4,5,6,7] has shown that residual stress prediction in aluminium alloys using the finite element method is quite accurate. However, verification of these predictions is dependent on the inherent inaccuracies in both the determination of the stresses and in the definition of the material's plasticity behaviour. This paper aims to ensure the accuracy of the finite element model through verification obtained using two different techniques, namely the non-destructive X-ray diffraction technique and the semi-destructive hole drilling strain gage method.

EXPERIMENTAL PROCEDURE

HEAT TREATMENT

The material (block 1 – fig 1), was received from HDA Forgings Ltd, Redditch, UK, as a rectilinear 124(ST)*156(LT)*550mm(L) open die forging. It had undergone the following heat treatment: solution heat treated for 6 hours at 475°C, quenched into water at room temperature (<40°C), cold compressed (2¼±½%) and aged for 10 hours at 120°C followed by 8hrs at 172°C (T7652).

Block 1 was re-heat treated as follows: solution heat treatment (two hours at 470°C±5°C), horizontal quench (a 156*550mm or 124*550mm face entered the quenchant first) in still water at room temperature (<40°C). This solution heat treatment was repeated eight times allowing the temperature of the block to be monitored during the quenching operation through four type 'k' thermocouples using a Labview[8] data acquisition system. These thermocouples were located at a 124*156mm cross-section in the centre of the block (see fig 2). Any finite element predictions could then be compared with these measured cooling curves. The grain structure of the material proved stable due to the pinning action of the fine Al_3Zr dispersoids present, preventing the subsequent solution and ageing heat treatments from having any effect on the optical microstructure. Following solution heat treatment, the material was aged to a T6 type temper (24 hours at 120±5°C).

Four samples of block 2 (26*26*156mm) were sectioned from block 1 (see fig 1). They were subsequently solution heat treated for two hours (470°C±5°C) and quenched vertically (a 26*26mm face entered the quenchant first) in still water at room temperature (<40°C).

RESIDUAL STRESS DETERMINATION

Verification of the residual stress magnitudes predicted for block 1 was obtained through use of the hole drilling strain gage method as outlined by ASTM E837-95[9] and in the procedure provided by Measurements Group[10]. Strain gage rosettes of type CEA-13-062UL-120 were used with a three-wire quarter bridge set-up to prevent any temperature changes in the lead wires affecting the measured strain values. Data acquisition of the resultant strains was achieved through a Measurements Group P-3500 Portable Strain Indicator (±1με) connected to a Measurements Group model SB-10 switch and balance unit. A Measurements Group RS200 milling guide and assembly was used to introduce the hole for each measurement, with the air turbine assembly used to prevent introducing residual stresses during the measurement. An orbiting technique was used to introduce the hole, resulting in each hole having a diameter of ~1.88mm. This gave a hole diameter to mean gage diameter ratio of ~0.37, which is within the parameters recommended by ASTM E837-95[9]. Each hole was drilled to a depth of 2mm as dictated by ASTM E837-95[9].

In cases where residual stress values are greater than 1/2 of the yield stress, some region around the perimeter of the hole will be plastically deformed[10,11] leading to an overestimate in the residual stress magnitudes. To prevent this, the material was aged to a T6 type temper, resulting in an increased tensile strength (0.2% proof of ~500MPa[12]) with only a small reduction in residual stress magnitude[13].

Residual stresses were determined at three locations on the L-T face of block 2 using the X-ray diffraction technique (Philips X'Pert X-ray diffractometer) as outlined in literature[14,15,16,17]. Scan parameters were controlled using Philips PC-APDW (V4.0c) software with 2θ (2θ - angle between source and diffracted X-ray beam) values chosen to encompass the Cu-Kα doublet for the {422} planes: 135.5deg<2θ<139.5deg. Eight scans were performed at each point using different ψ values for each (ψ – angle between surface normal and bisector of source and diffracted X-ray beam). The ψ values used were 0deg, -18deg, -27deg, -33deg, -40deg, -45deg, -51deg, and -57deg. Negative ψ angles were used because limitations of the goniometer prevented measurements being taken at angles of ψ>+25deg. All of the measurements taken indicated that the Δd versus $Sin^2(\psi)$ plots were linear, indicating that texture and stress gradient did not affect the calculated residual stresses. Shearing stresses were assumed negligible when compared with normal stress magnitudes, as they could not be determined given the limitations of the goniometer. The resulting spectra were analysed using Philips PC-Stress Software (version 2.61). The $Sin^2\psi$ technique was used with S_1 and $½S_2$ values of $4.97*10^{-12}(m^2/N)$ and $19.07*10^{-12}(m^2/N)$ respectively, both of which were taken from literature[14] for the {422} planes.

INPUT DATA FOR THE FINITE ELEMENT CALCULATION

HEAT TRANSFER ANALYSIS

The analysis method used by ABAQUS[18] to predict residual stress distributions after quenching is uncoupled in that it solves the temperature and displacement problems consecutively. Results from the thermal analysis are read at the beginning of the stress/displacement analysis and provide the displacement loading through thermal contraction. This thermal contraction results in the development of elastic and plastic strains from which residual stresses can be calculated. Askel et al.[19] gives a comprehensive summary of the general algorithm used for calculating residual stresses due to quenching. An uncoupled method is considered valid for this analysis, as the temperature of the alloy is not dependent on the strains or displacements that occur during quenching but only on the heat transfer that occurs at the sample's surface. Similarly, the precipitation hardening and volume changes that occur do not induce significant temperature changes.

For all of the three-dimensional heat transfer models, 8-noded quadratic brick, heat transfer elements were used, with the number required for each model determined through mesh density experiments. These experiments involved predicting residual stress magnitudes at a location in the core and on the surface of a 70mm cubic block using different mesh densities. This then allowed the calculated stress at each location to be plotted as a function of the mesh density (Number of elements per m^3) as shown in fig 3. From this graph, a mesh density of ~6E+6 [Elements/m^3] was used for future models to ensure accurate results.

Values for the material's specific heat capacity (C_p)[20], thermal conductivity (k)[21] and density (ρ)[2] vary as a function of temperature and are all readily available in literature for aluminium alloys. The ABAQUS code allows these parameters to be input at a few discrete temperatures from which it can interpolate for all temperatures observed during quenching.

The heat transfer coefficient (h) acts as the main boundary condition on the finite element model as it determines the rate at which heat leaves the material's surface. Some researchers[4] have used the heat transfer coefficient curve originally defined by Nukiyama[22] in 1934. His experiment involved immersing a Nichrome wire in a bath of still (not agitated) water and passing an electric current through it to produce heat. Heat flux was determined by measuring the current flow and the potential drop, while the wire temperature was determined through knowledge of the manner in which resistance varies with temperature. More recent advances have seen researchers determining the heat transfer coefficient curves through inverse methods[3,7,19,23,24].

Heat transfer data available in literature was limited so it was therefore decided to attempt to calculate it experimentally as outlined in the steps below.

1) A block measuring 60*60*20mm (Block 3 - fig 4), had a hole of 1.6mm diameter drilled at the centre of one face to a depth of approximately 19.5mm so that a type 'k' thermocouple of 1.5mm diameter could be inserted.
2) This block was then heated to 470°C and monitored using a Labview[8] data acquisition system taking measurements at 6Hz. It was then possible to plot a temperature versus time cooling curve as the block was quenched in water at room temperature (20°C).

The block was quenched vertically, whereby the 60*60mm face with the thermocouple just beneath the surface entered the water first.

3) A two-dimensional finite element model (with dimensions of 10mm by 0.1mm - fig 5) was built to simulate the quenching of the block. One-dimensional heat flow was assumed, thus mimicking the cooling of the block at the point where the thermocouple was inserted at 0.5mm beneath the surface. It was assumed that at high temperatures during quenching, the cooling at the centre of the large faces would be unidirectional. As the surface temperature approached the temperature of the quench media this would not be the case, as cooling would be influenced by heat loss from other faces. This would result in a higher estimated heat transfer coefficient at lower temperatures. The model contained twenty elements, biased toward the edge from which heat was leaving.

4) The main boundary condition defining the cooling of the model was introduced from the cooling curve obtained using the average values achieved from six quenches of the block shown in fig 4. The model took this cooling curve as the cooling curve of the end nodes from which it could calculate the surface element's heat flux at each time increment. From knowledge of the surface heat flux (q) and temperatures at the end nodes (T_{wall}) during quenching, the heat transfer coefficient could then be estimated for each time increment using eq 1. (T_{wall} refers to surface temperature and T_∞ refers to quenchant temperature).

$$q = h(T_{wall} - T_\infty) \qquad \text{Equation 1}$$

This calculated heat transfer coefficient could then be used in future three-dimensional models of simple shapes to predict time transient thermal cooling during quenching.

STRESS/DISPLACEMENT ANALYSIS

Values for the thermal expansion coefficient (α_{th})[25], elastic modulus (E)[26] and Poisson's ratio (ν)[26] of alloys similar to 7010 were taken from literature. The elastic modulus and the thermal expansion coefficient were input as a function of temperature while Poisson's ratio was assumed to remain constant. During the quenching of aluminium alloys, the material is plastically deformed at low strain rates (the finite element model of block 1 suggests a maximum of ~0.01sec^{-1}), the degree of which determines the final magnitude of residual stress. Unlike the elastic behaviour, the flow stress of 7010 is strain rate dependent. Knowledge of the deformation behaviour of 7010 at varying strain rates and temperatures up to 470°C is not widespread, and thus a compromise was reached in this model by using flow stress values obtained from torsion tests on 7150[27]. 7150 data was used because of its compositional and metallurgical similarities with 7010. Jackson[27] used the behaviour law defined by Sellars and McG. Tegart[28] (eq 2) to describe the material's flow stress (σ) behaviour as a function of absolute temperature (T) and strain rate ($\dot{\varepsilon}$). The material dependant constants used are given as: α=0.01 [m^2/MN]; Ln(A) = 29.8 – 30.7; ΔH = 160,000 [J/mol] and n = 5.5 – 5.7 while R is the Universal gas constant.

$$\dot{\varepsilon}\exp\frac{\Delta H}{RT} = A\left[Sinh(\alpha\sigma)\right]^n \qquad \text{Equation 2}$$

Measurement of the plasticity behaviour of 7150 was carried out over a range of temperatures from room temperature to the solution heat treatment temperature. One of the difficulties with this technique is that mechanical properties of 7xxx series alloys will change if held at intermediate temperatures. Therefore, when testing at these elevated temperatures, minimal delay between the sample reaching the testing temperature and completion of the test is vital to ensure reliable results. Due to these characteristics, some researchers[7] have only measured mechanical properties at high and low temperatures from which they have based their behavioural laws. Others[6] have carried out a cyclic tension-compression test during cooling from which the increasing measured mechanical properties can be used to determine the behavioural law. Both techniques have been found satisfactory for determining plasticity behaviour for computer models of quenching.

ABAQUS simulates strain rate dependence through a database type option that allows a variety of stress-strain curves to be defined at different strain rates and temperatures. ABAQUS then interpolates the yield stress at a given strain rate from these values[18] a technique that does not account for the precipitation effects that occur during quenching. The model used assumes that the material follows a perfectly plastic behaviour after yield. This assumption is true at higher temperatures but results in an oversimplification of the problem at lower temperatures where the material exhibits substantial hardening after yield. This will result in a maximum residual stress magnitude that will not exceed the material's yield strength at room temperature.

Plastic strains are assumed to dominate the deformation while elastic strains are small. This is true for aluminium alloys such as 7075 where yield stress to Young's modulus ratios are typically 0.7% after the application of T6 ageing treatments[26]. Given that ageing to T6 conditions results in the 7xxx aluminium alloys' yield strength more than doubling[29], yield strength to Young's modulus ratios are small after solution heat treating. These low ratios also hold true at higher temperatures (see Jeanmart and Bouvaist[2] for data). All of the mechanical property data used assumes the material to be isotropic and the plasticity data assumes the material to be volume invariant. Variation in mechanical properties in alloys such as 7010, after over-ageing treatments, have been found to be less than 10% for different mechanical working directions[30].

The elements types used for the displacement model were 8-noded quadratic brick stress/displacement elements with blocks 1 and 2 consisting of 7943 elements and 240 elements respectively.

RESULTS - HEAT TRANSFER

Fig 6 shows a graph of the calculated heat transfer coefficient and the curve used by Jeanmart and Bouvaist[2] (which replicates the curve used by Yoshihara et al.[4]) both plotted as a function of excess temperature. Other researchers have generally not plotted heat transfer curves making it difficult to compare the heat transfer coefficient data used. Fig 6 shows that the heat transfer coefficient magnitudes in Jeanmart and Bouvaist's curve[2] are somewhat lower than those calculated in this paper, resulting in a reduced cooling rate. These lower magnitudes result from possible differences between the two experiments used and may be summarised as follows:

1. The surface temperature during quenching is decreasing rapidly – not held at a constant temperature as with Nukiyama's experiment. This may account for some of the differences observed between the two curves. However, it is difficult to determine if this will affect Jeanmart and Bouvaist's curve[2], as full details of determination of this curve are not given in their paper.
2. During quenching, the initial excess temperature (Delta T) is not high enough for the film boiling regime to occur[24]. It has also been found when quenching steels[31] that the film boiling stage is short or non-existent during quenching resulting in an extension in the nucleate boiling regime. If the vapour blanket does form it is generally unstable and susceptible to changes with surface conditions[31].
3. The film boiling curve can be adversely affected by pressure, surface finish, ageing and surface coatings, dissolved gasses in the quenchants, size and orientation of the

heating surface and agitation of the quenchant[32]. Therefore, quenching different samples in water will result in marginally different heat transfer curves.

4. The quenched sample is plunged into a bath of water. The effect of plunging the sample into water would be similar to agitating the water over the first few seconds of quenching, resulting in increased surface cooling rates. The rate at which the sample enters the water will therefore also affect the rate of heat transfer.

A three-dimensional finite element model of block 3, using the calculated heat transfer coefficient values as the main boundary condition, showed good agreement with cooling curves measured experimentally (fig 7). The experimental values shown are the average values over the six quenches with the error bars showing the respective standard deviation in temperature at each time increment. The good agreement achieved is surprising given the nature of the experiment. The use of a cooling curve measured at 0.5mm beneath the surface should have given a heat transfer coefficient that was lower than the actual case. However, given the high thermal conductivity of aluminium alloys, the cooling rates measured are not much different from those that occur on the surface of the material. The use of a one-dimensional finite element model should also have produced a lower heat transfer coefficient at lower temperatures, as cooling through conduction becomes more evident. This reduced cooling rate is only evident at much lower temperatures and therefore does not affect residual stress development.

The predicted heat transfer coefficient values were input as a function of surface temperature for a finite element model of block 1. Cooling curves measured at four locations (at a cross-section half way along the block in the L-direction - fig 2) with the use of deeply buried thermocouples were compared with the model and good agreement was found. Rapid cooling between temperatures of 400°C and 250°C is critical to the development of mechanical properties[33] in 7xxx series alloys. Comparison of finite element and experimental cooling rates taken between these temperatures showed good agreement for thermocouples 1, 2 and 3 (Table 2). Overall, the results for thermocouple 4 had a high standard deviation and compared poorly to the predicted values. However, some of the measured cooling curves for this thermocouple compared well with the predicted values indicating that a poor connection existed between the thermocouple tip and the aluminium during the measurement.

During this experiment, block 1 was rotated about its longitudinal axis (through increments of 90deg) between quenches to observe if heat flow rates varied as a function of surface orientation during quenching. Cooling curves measured offered no indication that one surface had undergone a substantially more rapid heat loss than any other surface had. The finite element model therefore assumes that cooling is independent of orientation even though it is generally understood that heat transfer rates vary as a function of surface orientation. For example, during quenching heat will find it more difficult to escape from a downward facing surface than an upward facing one. During the quenching of aluminium alloys in water at room temperature (<40°C) most of the surface cooling occurs through a nucleate boiling regime as the material's surface temperature is not high enough to allow a vapour blanket to form[24,31]. Therefore, the problem of breaking down the vapour blanket to improve cooling rates does not occur when quenching aluminium alloys. Any agitation provided is used to carry bubbles of hot vapour away from the material's surface, ensuring that the lower surface will cool at a similar rate to the upper surfaces. More precise measurements taken on opposing surfaces of block 1 would be required to determine the effect of orientation on the heat transfer rate. However, for this study, the heat transfer coefficient is assumed the same on all surfaces.

RESULTS - RESIDUAL STRESS

Predicted residual stresses were compressive at the surface and tensile in the core for both large and small models. Stress magnitudes in the core and at the surface of block 1 (Table 3) approached the input yield strength of the material in its solution heat treated condition (~270MPa). The assumption that material behaviour is perfectly plastic after yield results in a maximum residual stress magnitude (compressive or tensile) which can only be equivalent to the material's yield strength at room temperature. This is a limiting factor in the model as the material may undergo some work hardening as the temperature reaches room temperature, resulting in residual stresses that are higher than this. However, other researchers of 7xxx series aluminium alloys have observed similar residual stress magnitudes[2,4,23,33]. The principal surface stresses were always found to be most compressive in the L direction as dictated by the dimensions of the block.

The maximum and minimum principal stresses and the angle α determined from the results of six hole drilling residual stress measurements taken on block 1 (fig 8) can be found in Table 3. Measurements taken using this technique have been found to be within an accuracy of ±10%[9] and the results shown in Table 3 are therefore also assumed to be within this 10% standard deviation.

The residual stresses that exist in a quenched sample are compressive on the surface and tensile in the core. Therefore, it cannot be guaranteed that the stresses in the first 2mm of the surface will vary uniformly with hole depth, as they may become decreasingly compressive with depth. However, the stress values calculated with the finite element model were uniform over the measured hole depth and were thus compared with the hole drilling results as uniform stresses. Calculated stresses were generally equal biaxial in nature which made verification of uniformity with hole depth difficult using the ASTM E837-95 standard technique[10].

The magnitude of the determined residual stress was, in some cases, greater than 50% of the materials yield strength after the T6 treatment. It is acknowledged that the determined stresses may have been marginally higher than the actual case and they should be compared to the predicted values as such.

The results of stresses calculated using the X-ray diffraction technique were taken at three locations on block 2 (fig 9) on both L-T faces, with the results shown in Table 4. Measurements were taken in the LT direction only, and averaged between measurements taken on opposing surfaces of the four samples. Stress magnitudes for block 2 were predicted to be approximately 50% of those predicted for block 1. The stresses calculated exceeded 70% of the materials solution heat treated yield strength indicating that most quenched die forgings of both small and large cross-sections would contain significant residual stress distributions.

Given the inherent inaccuracies in both the prediction and determination of residual stresses, the directions and magnitudes obtained for block 1 (Table 3) compare well.

However, predicted stress magnitudes for block 2 (Table 4) were less in all cases than those determined, revealing the limitations in the flow stress data used. The strain levels which occurred during quenching of the small block only approached the strain required to cause yield, resulting in an increase in the model's sensitivity to errors in the flow stress data. However, the level of strain in the large model greatly exceeded the material's flow stress, resulting in most of the predicted strain occurring in the plastic region, therefore leading to the appearance of a more accurate residual stress prediction.

CONCLUSION

The results show that the finite element method can be used successfully to predict the heat transfer coefficient from knowledge of surface cooling. Resultant time transient thermal cooling during quenching of simple shapes can then be predicted with reasonable accuracy using these calculated heat transfer coefficient values.

The stresses produced during the cold water quenching of aluminium alloys approach the yield strength of the material in its solution heat treated state, even for small cross-sections.

The prediction of residual stress magnitudes using verified cooling rates remains difficult given the complex plasticity behaviour which occurs during quenching. This suggests that the reasonably accurate prediction of residual stresses in block 1 may only be a coincidence and adds weight to the claim that residual stress prediction is extremely sensitive to material property definition[5]. A better definition of the plasticity behaviour is required and may be achieved through future work involving torsion testing at varying strain rates and temperatures.

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