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An experimental investigation of the drawability of AA6082 sheet under different elevated temperature forming processes

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Abstract

The performed research has, for the first time, investigated and compared the drawability of AA6082 at a comparable temperature range between two elevated temperature forming processes: termed (i) Low Temperature Hot Form and Quench (LT-HFQ[®]) or pre-cooled HFQ[®], patented by Adam et al. (2015) and (ii) Direct Heating Aluminium Forming (DHAF) which represents a refinement of conventional warm forming targeting a higher temperature range. A series of uniaxial tensile and cylindrical deep drawing experiments were conducted. According to uniaxial tensile test results, the most obvious work-hardening and reasonable ductility was observed under LT-HFQ[®] at a deformation temperature of 350 °C and strain rate of 1 s⁻¹, which can enhance drawability. For deep drawing experiments, it was found that preheating conditions of each process prior to forming significantly affected forming characteristics and post-formed hardness of the alloy; both the achieved maximum Draw ratio (DR) and limit Blank Holding Force (BHF) at some specific process parameters were increased under LT-HFQ[®]. Forming speed and temperature had significant effects on alloy deformation and thus drawability for both processes. In addition, by evaluating the post-formed hardness, process drawability and microstructural evolutions under different processes were simultaneously analyzed.

Keywords: Hot Form and Quench (HFQ[®]); Direct Heating Aluminium Forming (DHAF); drawability; deep drawing; AA6082.

1. Introduction

In recent years, increasing economic and environmental concerns have driven the automotive and aerospace industries to select alloys with lighter weight, such as aluminum and magnesium, to manufacture lighter structural components. Non-heat treatable AA5xxx are already widely used in automobile body inner structures, while applications of medium and high strength heat treatable aluminium alloys, such as AA6xxx and AA7xxx, are still limited due to the poor ductility of aluminium alloys in T6 condition and severe springback under cold forming condition, as experimentally proved by Kim and Koç (2008). These alloys are potentially cold formed in T4, W or O tempers with an improved ductility. However, additional heat treatments are required to increase post-formed strength, which may induce shape distortion and loss of dimensional accuracy. Therefore, elevated temperature forming processes have been developed and investigated. Toros et al., (2008) reviewed the warm forming processes of aluminium alloys. Mohamed et al. (2012) investigated the hot stamping process of AA6082. Fan et al. (2016) preliminarily explored the hot gas forming of Al-Li alloy. The above research concluded that the ductility of alloy enables to be improved at elevated temperature. Increasing forming temperature contributes to manufacture complex-shaped high strength aluminium panel components for automotive Original Equipment Manufacturers (OEMs).

In terms of stamping using rigid dies at elevated temperatures, warm stamping refers to forming alloys normally at an elevated temperature range, between $0.33 T_m$ and $0.6 T_m$ (approx. recrystallisation temperature), where T_m represents the melting temperature of alloys in Kelvin. It should be noted the above temperature boundaries may vary with different initial alloy state and process characteristics. Conventionally, preheating condition in warm stamping is simple, which is directly heating blanks to target temperatures below recrystallisation temperature. These temperatures used are normally lower than the Solution Heat Treatment (SHT) temperature of heat treatable aluminium alloys, resulting in precipitates undissolved. Therefore, the microstructural evolution of alloy is uncontrolled during preheating and warm stamping for heat treatable aluminium alloys. The alloy

deformation and post-formed strength might be deteriorated. In addition, the limited increase of ductility may not be sufficient for forming very complex-shaped components. A further increase of temperature in warm forming might be a solution to aforementioned problems. A refined warm forming process, defined as Direct Heating Aluminium Forming (DHAf) process in this study, was proposed with the aim of dealing with a higher temperature range. However, relevant investigations on the alloy deformation and post-formed strength for this new process are lacking.

Recently, a novel hot stamping process, named Hot Form Quench (HFQ[®]), has been developed by Lin et al. (2008) for aluminium alloys. In this process, a sheet blank is heated to its SHT temperature and soaked for a specific period so that original precipitates and dissolvable inclusions are dissolved to obtain a more ductile microstructure, as concluded by Garrett et al. (2005). Then the hot blank is quickly transferred to the press, stamped and held for a certain period between cold dies. Finally, the hot stamped component needs to be quenched to a temperature normally lower than artificial ageing temperature according to the specific alloy grade. For heat-treatable alloys, the in-die quenching enables the microstructure of formed component to remain as a super saturated solid solution (SSSS), which can be artificially aged to higher strength. The HFQ[®] process enables production of complex-shaped components with high strains and satisfying post-formed strength in one press and one ageing operation, which is believed to be the forefront of processes for stamping lightweight alloys. To further exploit its potential, Adam et al. (2015) patented a refinement of HFQ[®], aiming to deal with a lower temperature range has been introduced, termed ‘Low-Temperature HFQ[®] (LT-HFQ[®]) or pre-cooled HFQ[®]. The blank heated to SHT is rapidly quenched to a lower temperature range above $0.33 T_m$, before being formed in cold dies. By a scientifically strict definition, the LT-HFQ[®] can be either warm or hot forming, which depends on the cooled temperature is lower or greater than the recrystallisation temperature. The reason of adopting fast cooling is to avoid the generation of precipitates and preserve the optimal microstructure. According to previous studies, the refinement can potentially provide the following advantages: (i) low-melting phases which have shown to cause damage observed by Wang et al. (2011) during hot forming can be hot stamped, (ii) work hardening of the alloy is increased which suppresses localised deformation, as experimentally determined by Li et al. (2016), (iii) blank/tool interface friction coefficient can be reduced, and (iv) in-die quenching efficiency can be significantly increased. Zheng et al. (2018) simulated the practical in-die quenching stages for hot stamping aluminium alloy sheets at different temperatures. The

results evidenced the contribution of improving in-die quenching efficiency of LT-HFQ[®]. However, to the date, the capabilities of this novel LT-HFQ[®] process have not been thoroughly investigated.

Drawability is a key factor to evaluate the capabilities of a specific forming process and to determine the success in manufacturing complex-shaped panel parts. This represents the extent to which a sheet may be deformed (drawn deeply in the extreme) in a real forming process, which is related to both intrinsic material properties (e.g. viscoplastic responses under different process parameters, anisotropy) and extrinsic tooling conditions (tribological and thermal properties). Although the ductility and formability of AA6082 under HFQ[®] conditions have been extensively studied, for example, Mohamed et al. (2012) evaluated the uniaxial ductility through uniaxial tensile tests and Shao et al. (2017) determined the formability under biaxial stress states. However, it is insufficient to evaluate the capability of a stamping process, as ductility and formability represent intrinsic material properties only. Therefore, to be able to predict the capability of a forming process with confidence, systematic studies on drawability are essential.

There are studies on drawability of aluminium alloys at elevated temperatures having been performed. Laurent et al. (2015) revealed enhanced drawability of AA5754-O at an elevated temperature range between 150 °C and 200 °C. Fan et al. (2016) investigated the drawability for Al-Mg-Si sheets at an extended temperature range between room temperature and 500 °C. It was found that the limit draw ratio increased with an increase of temperature to 200 °C and decreased with further temperature increase. But, the effect of the initial heat treatment temper of the alloy was not taken into account. It has to be noted that the initial temper and microstructural evolution play a critical role during forming of heat-treatable aluminium alloys, in terms of affecting both forming capability and post-form properties. For instance, preheating in elevated temperature forming might induce the deterioration of microstructure due to precipitate growth and annealing, as a result, subsequent deformation is affected and post-formed strength is unsatisfied. For automotive applications, high strength heat-treatable aluminium alloys (AA2xxx, AA6xxx, and AA7xxx) in T6 or T4/W (unstable state of T4) tempers are normally used. They can be formed through the DHAF or LT-HFQ[®] process. The drawability and post-form properties are affected by not only deformation conditions (e.g. forming speed, temperature, etc.), but also temper and preheating routes under different forming processes. Still, systematic studies on this matter are missing.

The objective of this study is, to investigate and compare the drawability of heat treatable AA6082 under LT-HFQ[®] and DHAF conditions, at a comparable temperature range, while taking post-formed hardness into account. The effects of process variables including temperature, forming speed, initial alloy temper, and blank holding force (BHF) on drawability were studied using cylindrical deep drawing tests. To assist the analysis of deep drawing results, a series of high temperature tensile tests were performed under DHAF and LT-HFQ[®] at different process variables. In addition, the post-formed hardness of LT-HFQ[®] deep drawn cup cylinders with different process variables was assessed and compared with DHAF process of AA6082 in W temper.

2. Experimental programme

2.1 Material

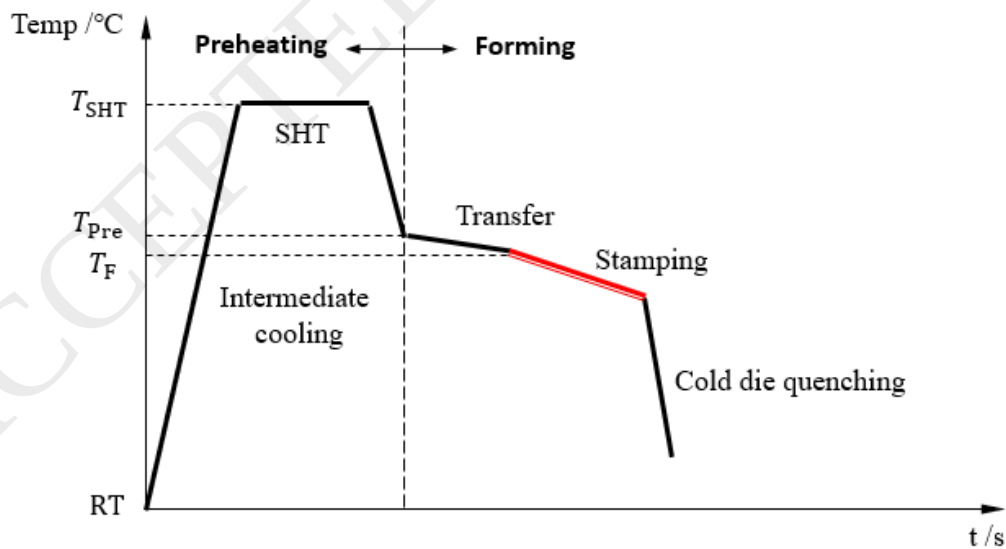
The test-piece material was commercial AA6082 sheet with a thickness of 1.5 mm provided by Smith Metal UK. The heat treatment temper of as-received raw material was T6, with an original Vickers hardness of 116 HV. The main chemical composition of AA6082 is given in Table 1.

2.2 Test temperature profiles

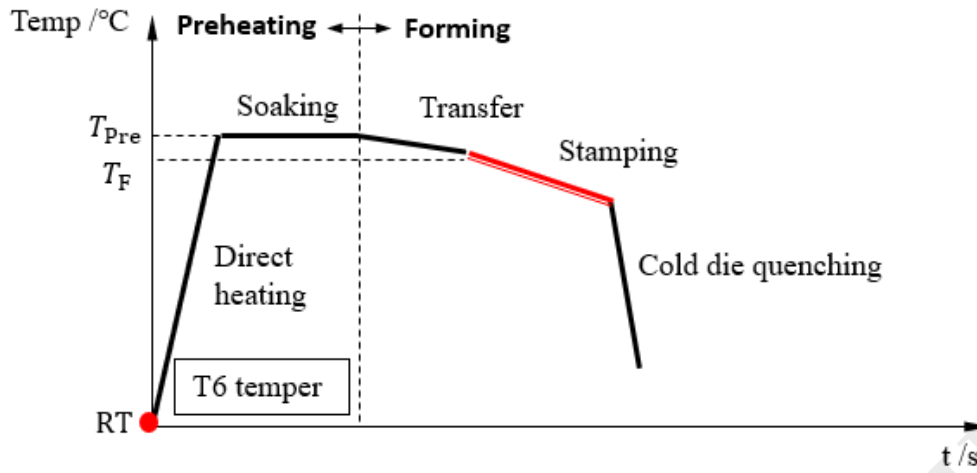
Fig. 1 shows the temperature profiles of the LT-HFQ[®] and DHAF processes used in this study. For each, the procedure is divided into preheating, shown on the left side of the dashed lines, and forming shown on the right side. The preheating refers to the pre-thermal treatment of alloy to achieve the target forming temperature T_{Pre} . Fig. 1(a) shows the temperature profile of LT-HFQ[®]. First, the material experiences solution heat treatment (SHT), which is consistent with conventional HFQ[®]; then, an intermediate cooling step is performed to quench the SHT treated material, at a sufficiently high cooling rate, to a lower temperature, T_{Pre} . The initial heat temper was T6. It is generally believed that annealing condition exhibited an increase of formability (Aginagalde et al. 2009), while during HFQ[®], the microstructure, especially precipitates can be dissolved in the experienced SHT. Therefore, the initial alloy temper is not critical for HFQ[®] process. To minimize the SHT time, T6

tempered alloy with fine and highly dispersed precipitates that can be quickly dissolved was chosen.

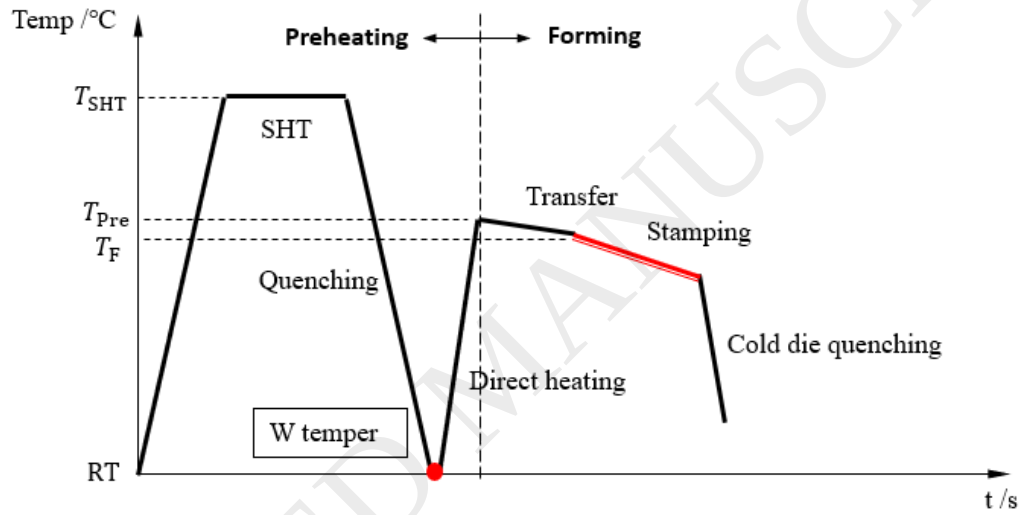
To assess drawability resulting from LT-HFQ[®], results have been compared with another elevated temperature forming process, DHAF. To investigate the effect of initial alloy temper, T6 and W tempers were used, as illustrated in Figs. 1(b) and (c), respectively. For DHAF of alloy in T6 temper (DHAF-T6), before blank transfer, the test-piece was directly heated to a high temperature, T_{Pre} , corresponding to the cooled temperature of LT-HFQ[®], and soaked for a certain period to obtain a uniform temperature distribution within test-piece. While for DHAF of W temper alloy (DHAF-W), the test-piece was first solution heat treated and then rapidly quenched to room temperature, in order to obtain the as-quenched W temper with a super saturated solid solution (SSSS). The quenched material was reheated to T_{Pre} and soaked. For both DHAF processes, the test-piece was subsequently transferred and stamped at a temperature of T_F . Compared with the LT-HFQ[®] process, test-piece temperatures of DHAF during the preheating stage (prior to stamping) are much easier to control without requiring a rapid cooling step. To unify the expression of temperature variable, T_{Pre} is used to represent the target temperature of alloy after preheating and before blank transfer in each process. For all of the three processes, T_F represents the start forming temperature; the as-formed part is quenched in closed cold dies.



(a) LT-HFQ[®]



(b) DHAF of alloy in T6 temper (DHAF-T6).



(c) DHAF of alloy in W temper (DHAF-W)

Fig. 1. Test temperature profiles of LT-HFQ[®] and DHAF processes.

2.3 High temperature tensile tests

2.3.1 Test-piece design and test equipment

High temperature uniaxial tensile tests were performed using a Gleeble 3800 materials simulator which enables an accurate temperature and strain control to be achieved. Specimens were machined from the as-received AA6082-T6 sheets. The dimensions of specimen are shown in Fig. 2. The dog-bone shape specimen had a 152 mm total length to enable a uniform temperature to be achieved along the gauge length. The total length of the parallel section was 80 mm and the width was 8 mm. The test-piece surface was sprayed black first and then with white speckles of paint to generate irregular white dots. The painted

pattern enabled the strain measurement by the Digital Image Correlation (DIC) system with a high-speed camera. The painted features were recorded using the camera, and the strain evolution can be calculated by the movements of the features during uniaxially stretching using the ARAMIS Software, which results in an improved accuracy of strain measurement. A frame rate of 500 fps with a resolution of 1280×1024 pixels was selected. The detailed description of the test set-up and strain analysis can be found in Shao et al. (2017).

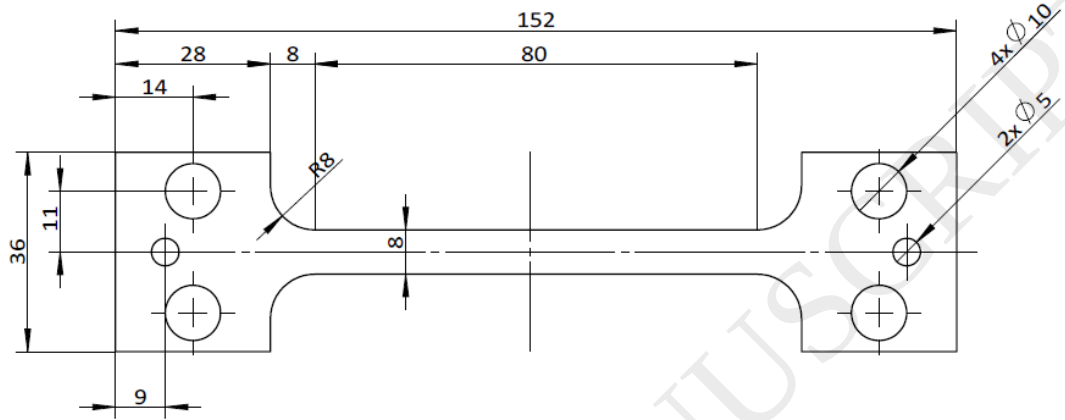


Fig. 2. Uniaxial tensile specimen design (All dimensions are in mm)

2.3.2 Test procedure

A typical deformation condition, 350 °C and strain rate 1/s, was selected to compare stress-strain behavior for the alloy under different processes. Also, a range of tensile tests at different elevated temperatures within a range, 350 – 525 °C, were performed under LT-HFQ[®] condition at a fixed strain rate, 1/s.

For LT-HFQ[®], the test-piece was heated to 535 °C and soaked for 2 mins to perform the SHT to dissolve existing precipitates. Two-step heating was utilized to reach the SHT temperature, which is heating up to 500 °C at a rate of 50 °C/s, and then 5 °C/s to 535 °C to avoid overheating. After SHT, the test-piece was quenched at a rate of 50 °C/s to different forming temperatures and stretched.

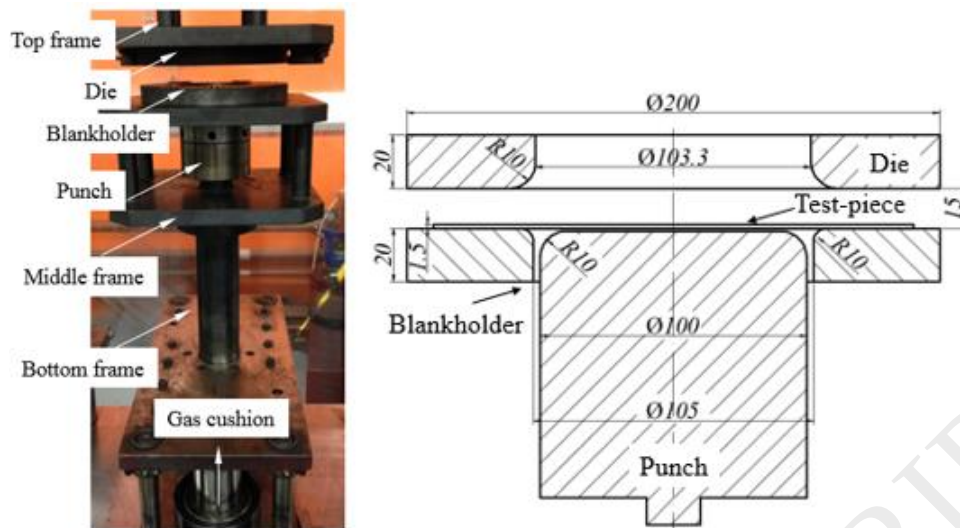
For DHAF-T6, the test-piece was heated to 350 °C at a rate of 50 °C /s and then stretched. For DHAF-W, the test-piece was solution heat treated first and quenched to room temperature at a rate of 50 °C/s to obtain the W temper. Then the W-tempered specimen was reheated to 350 °C and stretched.

2.4 Drawability tests

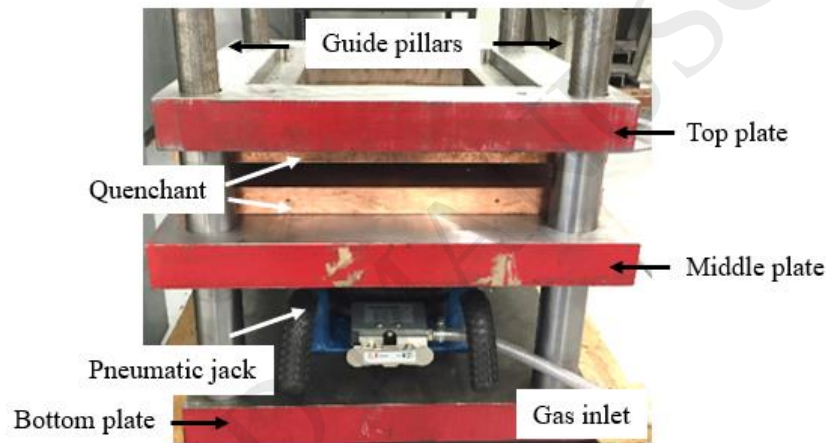
2.4.1 Experimental set-up

A Lenton radiant furnace was used for heating test-pieces before forming. The dimensions of the cylindrical deep drawing tool set are shown in Fig. 3(a). The rig was mounted on a 250 kN ESH hydraulic press. The maximum ram speed of this press was 500 mm/s and total stroke was 150 mm. Detailed dimensions and relative positions of the die, punch and blankholder are shown in Fig. 3(a). Test-pieces were 170, 180 and 190 mm circular discs, enabling three different draw ratios ($DR = D_{\text{test-piece}}/D_{\text{punch}} = 1.7, 1.8 \text{ and } 1.9$) to be obtained.

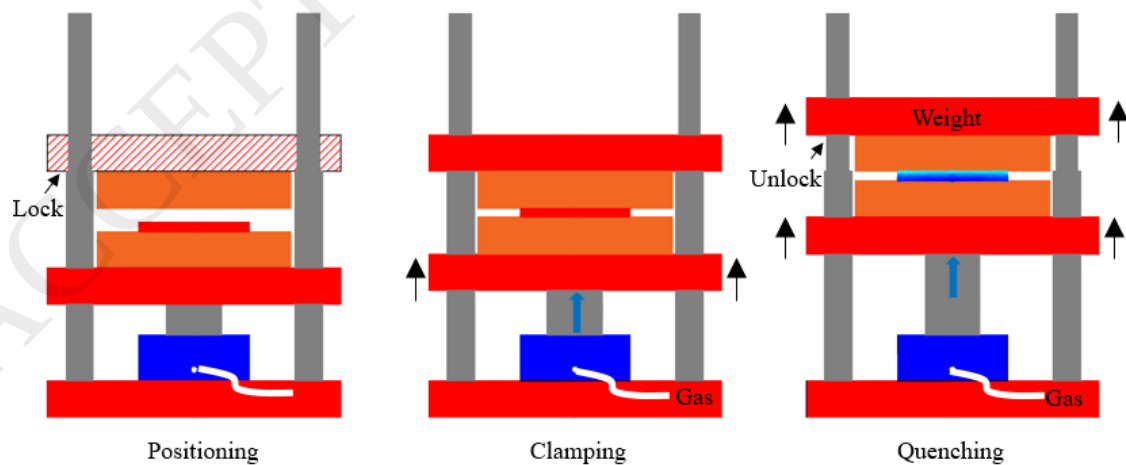
Intermediate quenching stage in LT-HFQ[®] was obtained by clamping test-pieces between cold metal slabs. Fig. 3(b) shows the rig used for this purpose. Different quenching rates were obtained by using different quenching medium, either copper or steel slabs, and different contact forces achieved by the weight of upper slabs. A reasonable quenching rate of the intermediate cooling stage of LT-HFQ[®] to avoid precipitation was 50 °C/s, which was achievable using this quenching rig. Fig. 3(c) shows the schematic of operation principal of this intermediate quenching rig. Initially, the upper quenchant was positioned on the top plate and separated with the lower quenchant using a lock on the guide pillars. After the SHT, the test-piece was quickly transferred from the furnace and positioned on the lower quenchant. Then the pneumatic jack below was activated to lift the lower quenchant and closed the gap between upper and lower quenchants. Once the gap was closed, the lock was released. The two quenchants and test-piece became a solid to be further lifted up together by the pneumatic jack. During the lifting, the test-piece was clamped and quenched between the two quenchants under a contact force exerted by the self-weight of upper quenchant. The test-piece temperature was assumed to be uniform after furnace heating and measured by attaching a thermocouple into the thickness section. The instant temperature was recorded using a Pico TC-08 data logger. When the temperature reached a target value, it was quickly transferred to the tool set for stamping and quenching.



(a) Deep drawing tool set



(b) Intermediate cooling rig



(c) Operating principal of the intermediate cooling rig

Fig. 3. Equipment for hot drawability tests (Dimensions are in mm): (a) deep drawing tool set, same tool set with the one in Zheng et al. (2016), (b) intermediate cooling rig and (c) operating principal of intermediate cooling rig.

2.4.2 Test programme

Test-pieces were heat treated following temperature profiles of different forming processes, presented in Fig. 1. The solution heat treatment (SHT) was performed at 535 °C and soaked for 5 mins to guarantee full dissolution of precipitates. For LT-HFQ[®] processes, the preheating temperature T_{Pre} , which represents the temperature at the completion of intermediate cooling here, was difficult to target for exact values by manual operations of the cooling tool, although it could be measured and recorded accurately through the attached thermocouples and data logger. So, a series of tests were conducted targeting varied values of T_{Pre} in the range between 300 °C and 525 °C and the exact values of T_{Pre} were recorded. Two forming speeds, 75 and 300 mm/s, were designed. For DHAF processes, the preheating temperature T_{Pre} , which represents the temperature at the completion of furnace heating here, could be controlled accurately. Therefore, a systematic test matrix with T_{Pre} of 350 °C, 400 °C and 450 °C and forming speeds of 75, 150 and 300 mm/s was defined, for both T6 and W tempered alloys. For all the forming processes, before each test, graphite-based lubricant was brushed to form even coatings on punch, blankholder and die surfaces. The Blank Holding Force (BHF) was set to a specified value.

After preheating and transfer, test-pieces were stamped at an elevated temperature, T_F , and cold die quenched to lower temperatures. It should be noted that, theoretically the forming process can be either warm or hot, which is determined by the temperature T_F ; also, as the transfer in this study was prompt, T_F can be taken as approximately equal to T_{Pre} (thus, only the effect of T_{Pre} is discussed).

Both LT-HFQ[®] and DHAF-W processes are designed to apply subsequent artificial ageing for age hardening of the formed components. In this study, the LT-HFQ[®] and DHAF-W formed parts were refrigerated at a temperature of -20 °C right after forming to preserve the microstructure, and then they were artificially aged by following a standard procedure for AA6082, at 190 °C for 9 hours. Two T_{Pre} values, 360 °C and 415 °C, were selected from LT-HFQ[®] processes to compare with the DHAF-W process for post-formed hardness after artificial ageing.

2.5 Hardness tests

To evaluate post-formed hardness, Vickers hardness tests were performed. Fig. 4 shows the measuring points along the cross-section of a half deep drawn cylinder and the indentation direction was perpendicular to the section. Considering the axisymmetric feature, hardness was only measured on half of a part. Six positions were chosen on half of this cross-section including the flat bottom, punch radius and side-wall regions, as marked in the figure. The samples were prepared using abrasive paper # 1200 to guarantee the cleanliness and flatness of the surface. Vickers hardness with a 5 N/s loading speed was used. For each position, 5 indentations were measured and the average value was used.

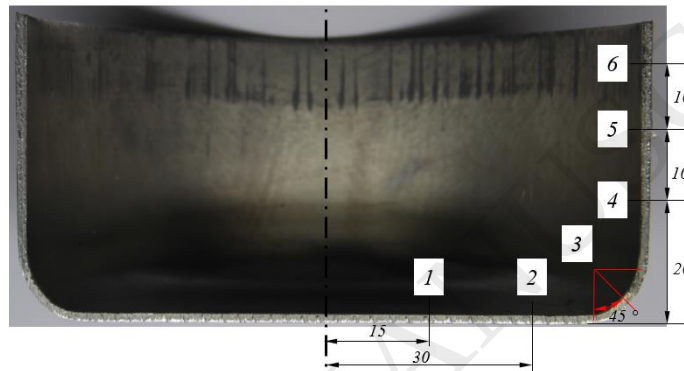


Fig. 4. Hardness measuring points on a sectioned deep drawn cylinder (unit of the dimensions in mm).

3. Results and discussion

3.1 Uniaxial tensile tests

Fig. 5(a) is a comparison of uniaxial tensile results for different processes at a temperature of 350 °C and strain rate of 1/s. In addition, the strain hardening exponent, n , determined by the true stress-strain curves in Fig. 5(a) using the Hollomon equation, as given in Eq. (1),

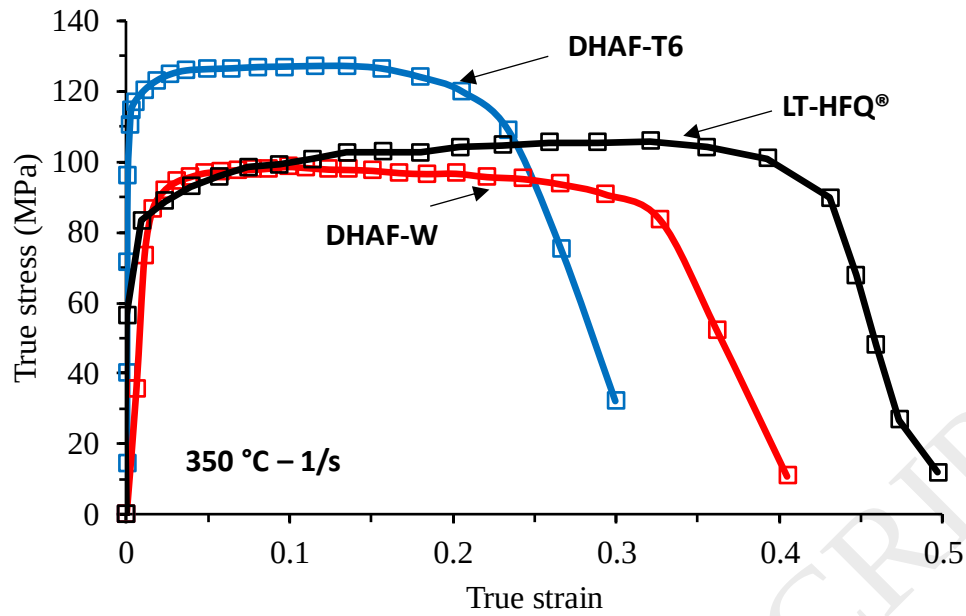
$$\sigma = K\varepsilon^n \quad (1)$$

where σ is the flow stress, ε is the plastic strain, n is the strain hardening exponent and K is the strength coefficient. Then, by applying the natural logarithm on the two sides of Eq. (1), the strain hardening exponent, n , can be determined from the plot of $(\ln\sigma \sim \ln\varepsilon)$, where n is the slope.

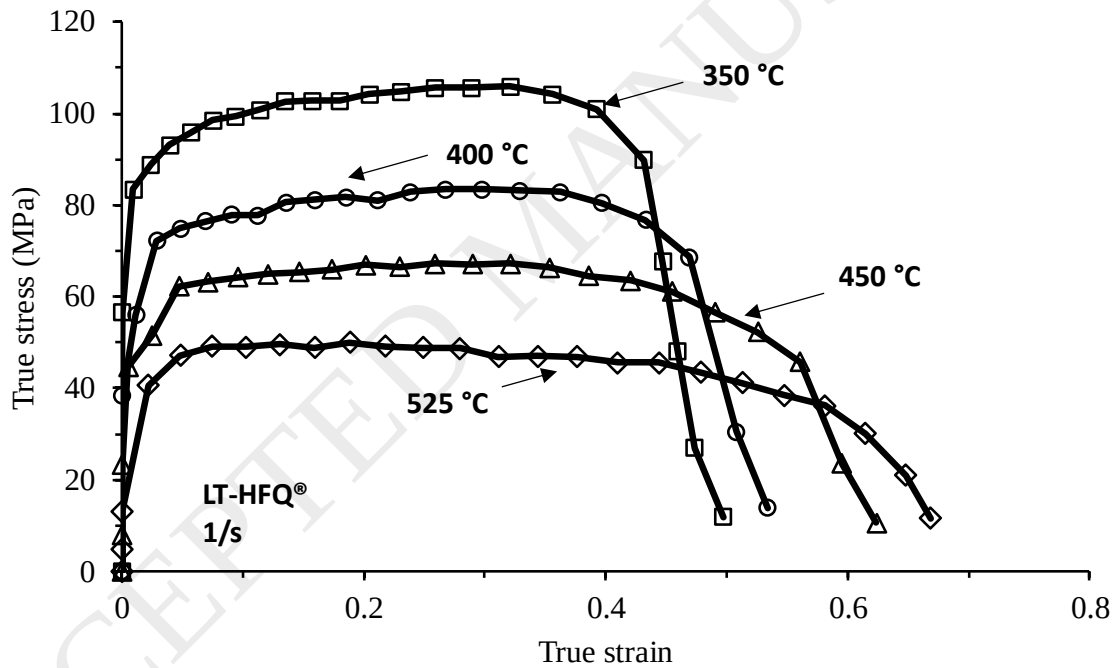
$$\ln \sigma = n \ln \epsilon + K \quad (2)$$

The values of n for different process conditions were summarised in Table 2. As can be seen, the alloy processed under LT-HFQ[®] process has the greatest ductility and strain hardening. The reason could be that, during LT-HFQ[®], precipitates were fully dissolved during SHT, and these dissolved precipitates remained within α – Al matrix during intermediate cooling as the rate was sufficiently high. In comparison, for DHAF-T6, test-pieces in T6 temper contained small precipitates distributed within the matrix and grain boundaries, as observed by Wang et al. (2011). When the test-pieces is heated to 350 °C, the initial precipitates tended to grow rather than dissolve. During uniaxial stretching, the grown precipitates impeded dislocation movement and contributed to hardening. Voids readily generated around these precipitates and macro-scale cracks emerged through void linking. Thus, the test of DHAF-T6 exhibited the highest stress level but poorest ductility. Compared with DHAF-T6, the test-piece in W temper processed under DHAF had a higher ductility but lower flow stress. This was because the initial alloy was heat treated to W temper with most precipitates dissolved. It is noted that a small portion of precipitates may still be generated due to the energy provided during reheating. Subsequently, these precipitates may grow quickly at 350 °C since it is near the nose temperature of AA6082 according to the Temperature-Time-Property (TTP) curve that is constructed by Shang et al. (2011). Nevertheless, the number of precipitates was less than that of DHAF-T6., thus the flow stress would be decreased and ductility would be increased.

Fig. 5(b) shows the temperature effect on deformation characteristics under LT-HFQ[®]. 525 °C (representing the conventional HFQ[®] process) and three lower temperatures were chosen to test the thermo-mechanical properties of AA6082. As can be seen, with decreasing temperature, flow stress increases and strain hardening becomes more obvious. The increase of flow stress, representing a harder material, which contributes the material to overcoming tool friction and being drawn into the die instead of stretching. In addition, the increase in strain hardening implies improved drawability due to a tendency for suppression of localized thinning and a uniform deformation. The strain hardening exponent, n , were summarised in Table 2 for different temperatures under the LT-HFQ[®] process. The temperature effect on stress-strain behaviours can be used to assist the analysis of deformation features of deep drawing tests using LT-HFQ[®] in the following section.



(a) Different processes at 350 °C and strain rate 1/s



(b) LT-HFQ® process at different temperatures

Fig. 5. Elevated temperature uniaxial tensile tests of AA6082: (a) different processes at 350 °C and strain rate 1/s and (b) LT-HFQ® at different temperatures and strain rate 1/s.

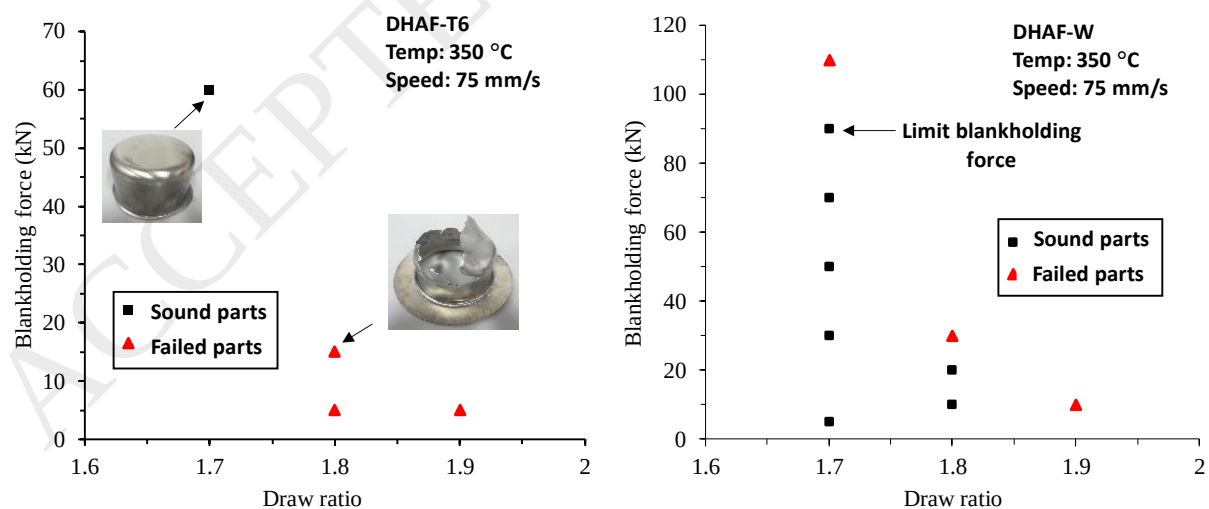
3.2 Drawability at elevated temperatures

To reflect the drawability, the conventional method is to compare the drawing depth of a cylinder or explore the limit draw ratio. In this study, similar to the latter method, the achieved maximum draw ratio among the three defined draw ratio values was used. In addition, for the same draw ratio, the drawability can be also indirectly reflected by the experimental limit BHF. Limit BHF represents the maximum value of BHF which guarantees a successful forming of a complete cylinder under certain process parameters. So, for a specific set of process parameters, the value of limit BHF could be obtained from conducting a series of deep drawing tests with varied BHF's. Theoretically, the drawing depth increases with the increasing BHF if tearing does not occurred, as the BHF can suppress thickening in the flange and enables more material to be drawn in. However, due to the earing induced by anisotropic feature and stroke error arising from ram fluctuations of used hydraulic press, the difference of drawing depth is difficult to compare accurately if the draw ratio is not large; therefore, it is reasonable to use the limit BHF concept to indirectly compare drawability for relatively smaller draw ratios in the study. This section mainly focused on the forming process parameters, i.e. temperature and forming speed, on the hot drawability. Garrett et al. (2005) has concluded that the anisotropy for AA 6xxx alloys is not significant after SHT. In addition, the use of lubricant kept a consistent interface friction condition. Therefore, the rest of influencing factors, like interface friction and normal anisotropy of material, were not taken into account in this study.

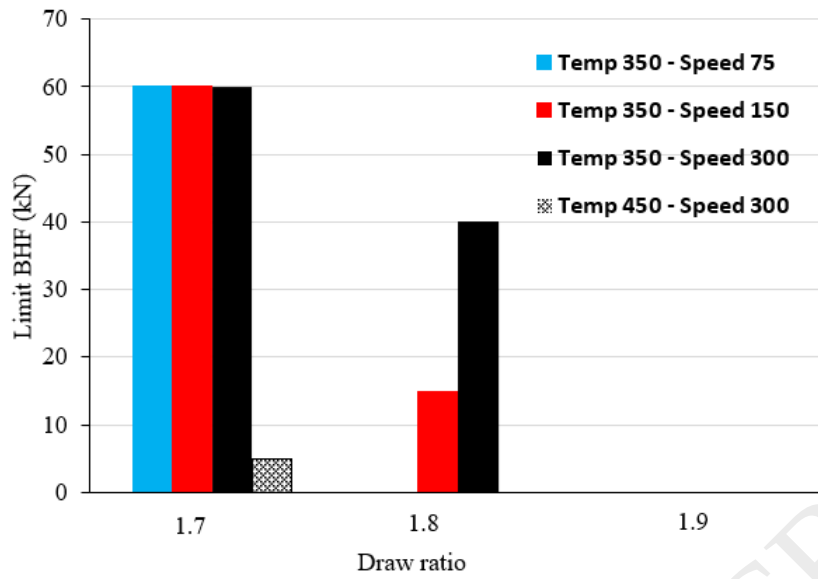
Fig. 6 shows the investigation of the drawability of AA6082, in both T6 temper and W temper formed under DHAF processes, reflected by a comparison of limit blankholding forces (BHF's) for a range of draw ratios (1.7, 1.8, and 1.9) under different temperatures and forming speeds. Normally, the maximum depth of a cylinder which can be drawn is related to BHF, when other process variables are fixed. A low BHF may result in wrinkling; a greater BHF enables to constrain the thickening in the flange thus results in more flange material drawn in and a greater drawing depth; however, a severely high BHF may cause tearing at the die entrance. Therefore, for a particular draw ratio, there exists a limit BHF (maximum value) beyond which tearing would definitely occur. In this study, for a specific combination of process variables, a series of tests were performed using different BHF's. The highest value of BHF obtained experimentally was taken as limit BHF under the certain draw ratio and forming condition. As shown in Fig. 6(a), the black square symbols represent complete cylinders that were formed successfully with the initial circular discs fully drawn in, while

red triangle symbols represent failed parts. For example, for the material formed at 350 °C and 75 mm/s, under DHAF-W, the limit BHF was determined as 90kN with a draw ratio of 1.7. In general, under a specific set of process variables, there exists a limit draw ratio beyond which the test-piece cannot be completely drawn into cylinder no matter how BHF is adjusted. As shown in the figure, no sound parts were obtained when the draw ratio was 1.9 for both example conditions.

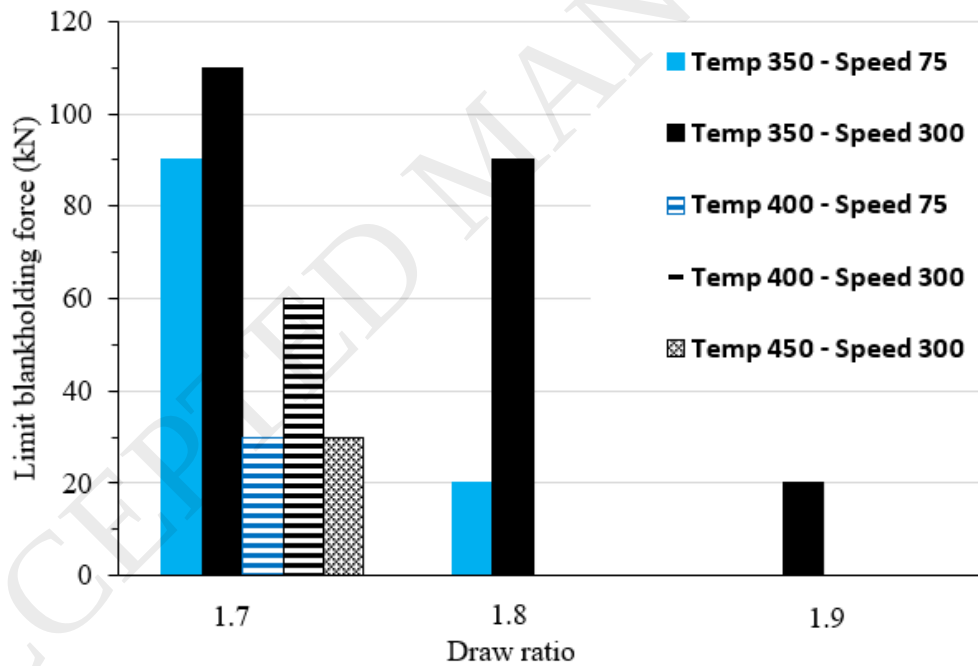
Figs. 6 (b) and (c) show the summarised limit BHF experimentally determined for different draw ratios at different process variables under DHAF-T6 and DHAF-W processes, respectively. In the figures, a higher limit BHF means greater drawing depth to be potentially obtained, thus indicating better drawability; the absence of a bar for a certain draw ratio indicates that the limit draw ratio was exceeded, under the specific forming condition, thus indicating poorer drawability. As can be seen, the process variables played an important role in general. For both tempers, the drawability increased with decreasing temperature and increasing forming speed. It was believed to be caused by enhanced hardening features, including strain hardening (Li et al., 2016) and strain rate hardening (El Fakir et al., 2014) under hot stamping conditions. Compared between the two tempers, better drawability is shown by DHAF-W. For example, a draw ratio of 1.9 was achieved at 350 °C and 300 mm/s for DHAF-W, while unachievable for DHAF-T6. In addition, for smaller draw ratios 1.7 and 1.8, the limit BHF were greater under DHAF-W. The trend is consistent with the result and analysis derived from the tensile test in the earlier section.



(a) Determination of limit BHF corresponding to draw ratios (1.7, 1.8, and 1.9), under two example forming conditions. (black square symbols represent sound parts and red triangle symbols represent failed parts)



(b) Summarised limit BHF for different process parameters under DHAF-T6 (Each color represents a specific combination of temperature and speed; the absence of a bar indicates an excess of limit draw ratio.)



(c) Summarised limit BHF for different process parameters under DHAF-W (Each color represents a specific combination of temperature and speed; the absence of a bar indicates an excess of limit draw ratio)

Fig. 6. Drawability of AA6082 under DHAF processes

Fig. 7 shows the obtained BHF at different temperatures with a forming speed of 300 mm/s under LT-HFQ[®] processes, for a range of draw ratios (DR: 1.7, 1.8, and 1.9). The black square symbols indicate sound parts while the red triangle symbols represent ruptured parts. For the draw ratio of 1.7, a successful deep drawing was achieved at a preheating temperature of 415 °C and 300 mm/s of forming speed, with BHF of 30 kN. For the draw ratio 1.8, it is very difficult to succeed if the temperature was higher than 410 °C. The draw ratio of 1.9 was achievable when the BHF was as low as 5 kN. All the obtained successful results were summarised in Fig. 7(d), which can be used to compare with the determined limit BHF under DHAF processes in Fig. 6.

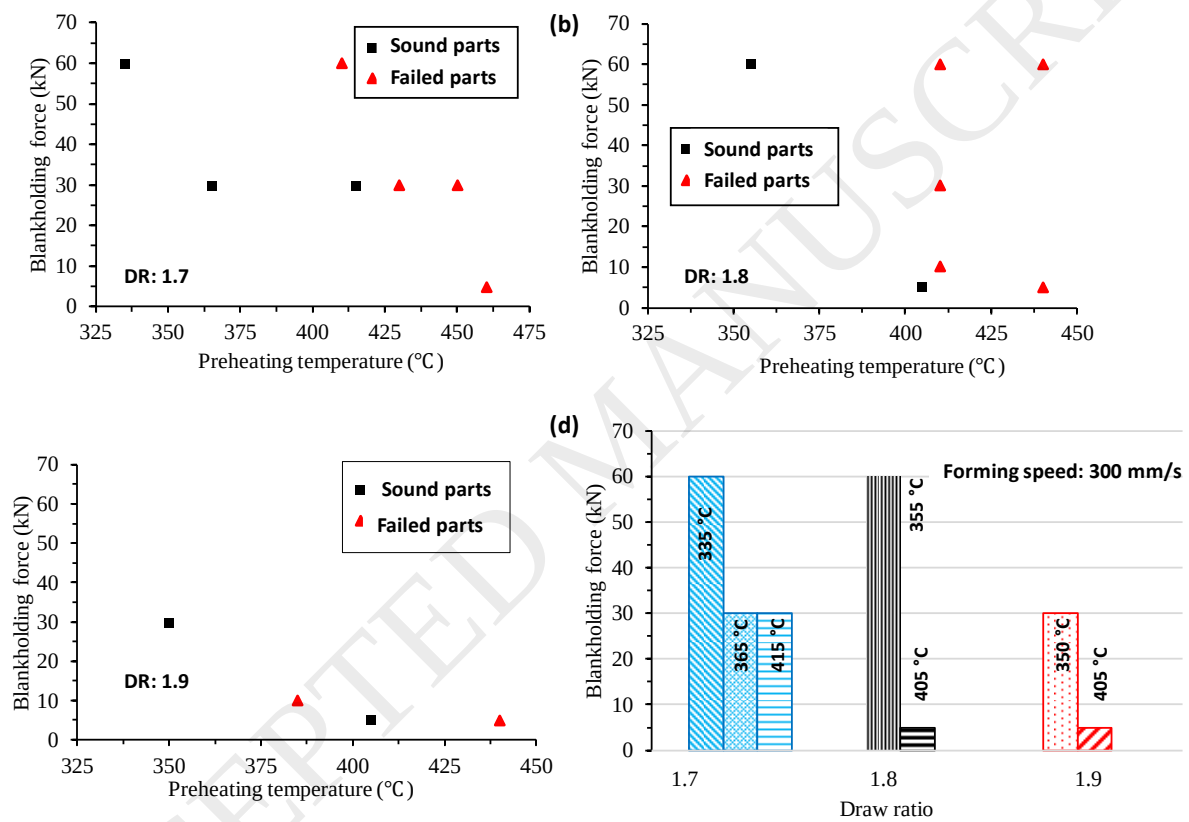


Fig. 7. Drawability of AA6082 at different temperatures with a forming speed of 300 mm/s, under LT-HFQ[®] processes, for draw ratios of (a) 1.7, (b) 1.8, and (c) 1.9. The solid square symbols indicate sound parts and the triangle solid symbols indicate failed parts with tearing; (d) summarised successful conditions. (The conditions of failed results are not shown).

Compared DHAF-W with LT-HFQ[®] processes, as can be seen in Fig. 6(c) and Fig. 7(d), a draw ratio of 1.9 was obtainable at a temperature of 350 °C and speed of 300 mm/s under both LT-HFQ[®] and DHAF-W. The limit BHF for DHAF-W was 20 kN; while for LT-HFQ[®],

it was successful at a higher BHF 30 kN, which indicates an improved drawability. In addition, LT-HFQ[®] process enabled to successfully deep draw a test-piece with a draw ratio of 1.9 at a temperature of 405 °C and speed of 300 mm/s; while for DHAF-W, even a draw ratio of 1.8 was unachievable for 400 °C. The above comparisons demonstrate the improvement of drawability of LT-HFQ[®] compared with DHAF. It was believed that such improvement was mainly due to two reasons. First, a more significant strain hardening behavior of material was observed under LT-HFQ[®] with the proceeding of deformation. As quantified in Table 2, the strain hardening exponent under LT-HFQ[®] at the condition of 350 °C × 1/s is 0.066, which is greater than that of DHAF-W, 0.018. The strengthening of material contributed to a more uniform deformation and eased the test-piece to be drawn into the die. While under DHAF, the severe softening phenomenon of material may induce localized deformation leading to cracks. Second, the ductility of material in LT-HFQ[®] condition was greater, which means the test-piece could sustain a greater deformation before tearing so as to obtain a greater draw ratio.

Fig. 8 shows the effect of preheating temperature on the material deformation and tearing location in LT-HFQ[®] processes. In order to reflect the degree of material drawn-in, a normalized diameter reduction of test-piece, Δ , was used and calculated as: $\Delta = (D_i - D_f) / (D_i - D_p)$, where D_i is the initial test-piece diameter, D_f is the flange diameter of formed part, and D_p represents the punch diameter. The test-pieces shown in the figure were deformed at a speed of 300 mm/s at, with the draw ratio 1.7 and BHF 30 kN. The black square symbols represent sound parts which were fully drawn in, while the red triangle symbols represent failed parts. As shown in this figure, temperature played an important effect on the failure locations. At the higher temperature, such as 450 °C, failure occurred at the die entrance, which was mainly because the material at the die entrance had a high temperature initially due to the absence of contact with the tools; correspondingly, the material strength was low as indicated by the maximum flow stress in Fig. 5(b) and strength coefficient, K , and was not bolstered by strain hardening and insufficient to overcome flange friction. As a result, localised deformation was easy to occur here. Draw-in was virtually absent and failure occurred. At a temperature of 430 °C, failure occurred at the punch nose, and some draw-in of the flange was evident. As shown in Fig. 5(b), material flow stress and strain hardening behaviour were increased at a lower temperature. The strain hardening exponents under LT-HFQ[®] decreased with increasing temperature as shown in Table 2. Therefore, the material at flange became strong enough to overcome friction force and

enabled material drawn in initially. In addition, the decrease of interface friction due to decreased test-piece temperature also contributed to the flange material drawn in. During drawing, the temperature difference between the punch nose area and flange area was greater. Hence, deformation began to concentrate on the punch nose area where the material had the highest temperature and lowest strength, resulting in tearing experienced here. When further decreasing the forming temperature to 420 °C or even lower, a fully drawn cylinder was achieved. This was because the whole test-piece was stronger and the temperature difference within test-piece was relatively less, which contributed to a more uniform deformation.

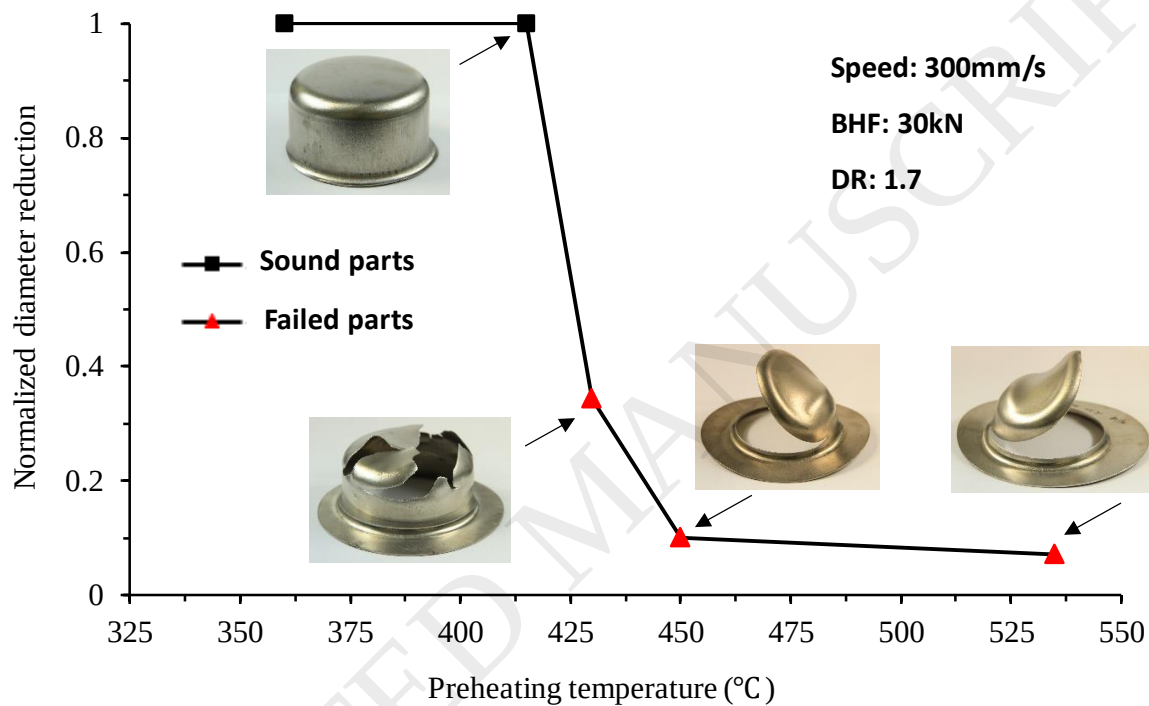


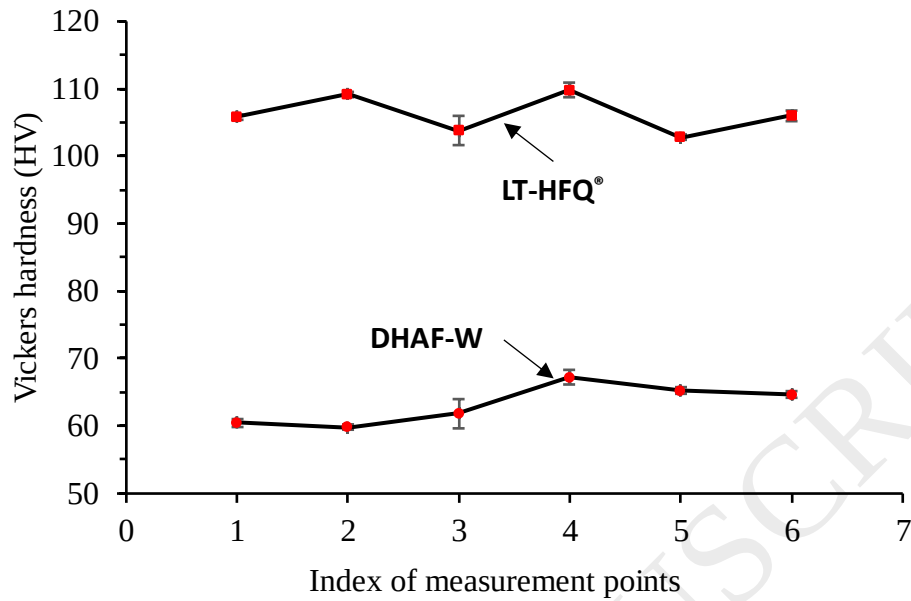
Fig. 8. Effect of preheating temperature on the material drawn in and tearing location for AA6082 deformed at a forming speed of 300 mm/s under LT-HFQ[®], with a draw ratio of 1.7 and BHF of 30 kN.

3.3 Post-formed hardness

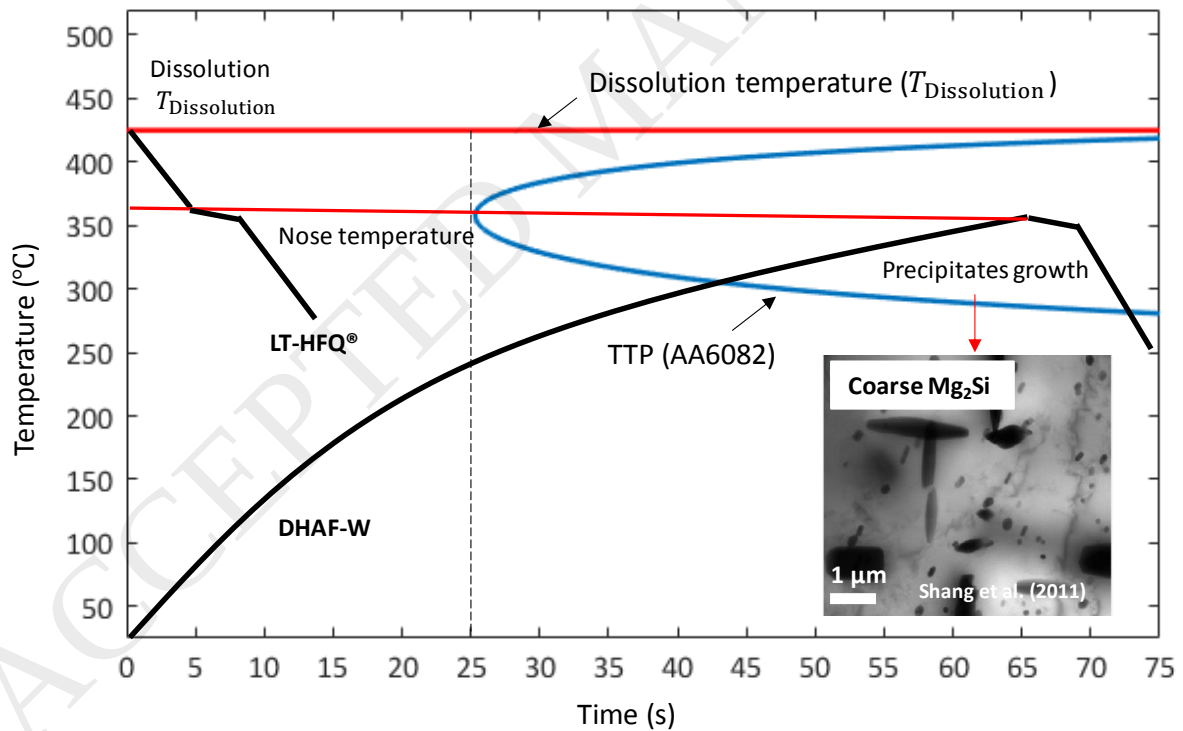
Fig. 9(a) shows the comparison of the hardness of parts formed through LT-HFQ[®] and DHAF-W processes, followed by artificial ageing under the same condition of 190 °C for 9 hours. Both work-pieces were deformed under the same process variables of 360 °C and 300 mm/s, with a draw ratio of 1.7 and BHF 30 kN. The hardness of parts formed through LT-HFQ[®] was significantly higher than the parts formed through DHAF-W. The post-formed hardness is believed to be caused by the underlying microstructural evolution. For heat-

treatable AA6082 in this study, the strength of alloy is dominated by the precipitation hardening (Polmear, 2005). It is generally believed that the stacking fault energy is high for aluminium alloys. It is assumed that dynamic recrystallization is difficult to occur during hot stamping as the deformation time is very short (as the stamping speed is very fast) and deformation level is low (as the material is more drawn-in than stretched). The assumption has been successfully used in constitutive modelling work (Mohamed et al. 2012) (Lin et al., 2014). Therefore, effect of grain size evolution during different processes on hardening and post-formed strength can also be neglected compared to the effect of precipitation during thermal-mechanical processing for aluminium alloys with the dominant precipitation hardening mechanism. For the W temper material in DHAF-W process, the material was in an as-quenched state (super saturated solid solution, or SSSS) with an unstable microstructure. During heating to T_{Pre} , the energy input could trigger pre-dissolved precipitates to regenerate. In addition, considering the insufficient heating rate of the furnace and soaking at 360 °C which is the nose temperature of AA6082 in TTP diagram with the fastest rate of precipitation, as shown in Fig. 9(b), both the high phase transformation tendency and the high diffusion rate of solute atoms could lead to the primary precipitation of Mg_2Si particles coarsely, according to the study by Shang et al. (2011). These coarse precipitates are not strengthening phases like those in T6 condition. β'' is the dominant strengthening phase of AA6082 (Li et al., 2015); however, the precipitation of coarse precipitates could occupy the required elements of β'' , i.e. Mg and Si, which would result in less elements to generate β'' in the subsequent artificial ageing process. Therefore, compared to ageing from SSSS state, the full strength could not be achieved. While in the LT-HFQ[®] process, the tiny and finely distributed, β'' , phases in the raw blank (T6 condition) could be completely dissolved during the SHT of LT-HFQ[®]. Because the SHT treated material was rapidly quenched from the temperature above the dissolution temperature of coarse Mg_2Si phase (Milkereit and Starink, 2015) during transfer and then was stamped at a high rate, precipitation of Mg_2Si particles could be avoided during the short time period prior to and during forming, as schematically illustrated in Fig. 9(b). As a result, the material could achieve a SSSS after cold-die quenching. Such a microstructure could be artificially aged to obtain fine and dispersed precipitates, which is optimum for higher hardness. As can be seen in this figure, hardness under LT- HFQ[®] was nearly the same as that of the as-received alloy in the T6 condition. It should be noted that, the DHAF-T6 process is not considered here since no artificial ageing

was applied after forming. For heat treatable alloys in T6 or T4 condition, the direct heating can cause initial precipitates to grow, which impairs subsequent age-hardenability.



(a) Comparison of post-formed hardness between two forming processes

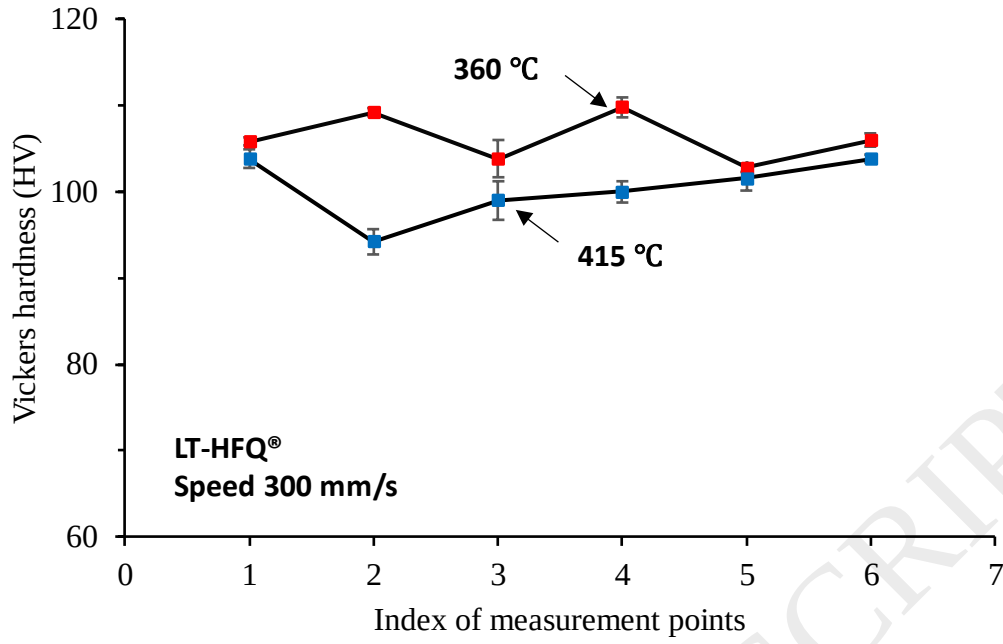


(b) Schematic of TTP curve of AA6082; the data for replotting the curves and the TEM image were from Shang et al. (2011).

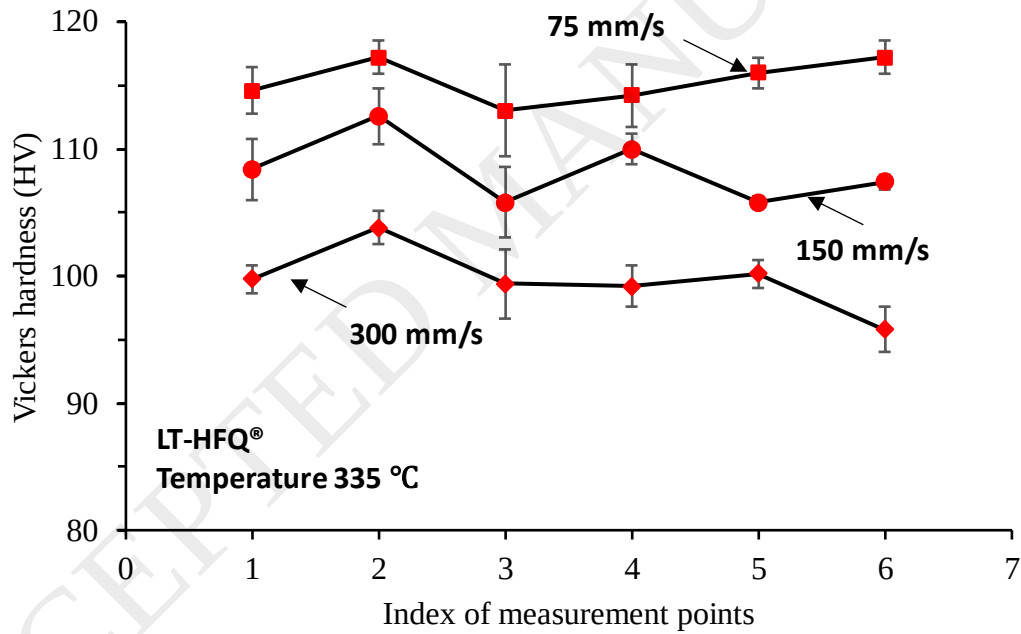
Fig. 9. Effects of forming processes on the hardness of post-formed parts after artificial ageing at 190 °C for 9 hours. (Preheating temperature 360 °C, forming speed 300 mm/s, draw ratio 1.7, BHF 30 kN)

Fig. 10 shows the effects of (a) temperature and (b) forming speed on the age hardening performance of LT-HFQ[®] formed parts. Fig. 10(a) shows that, the average hardness of parts with $T_{\text{Pre}} = 360$ °C was greater than that of parts with $T_{\text{Pre}} = 415$ °C, where T_{Pre} is the preheating temperature after intermediate cooling. Both cases had the same draw ratio of 1.7 and BHF of 30 kN. The reason could be that, in the LT-HFQ[®] process, forming starting from the higher temperature made test-piece material stay within the sensitive temperature region for a longer time compared with forming starting from the lower temperature. As discussed earlier, a longer holding time within sensitive temperature range around nose temperature can induce the generation of large β and β' phases. As a result, insufficient β'' phase would be generated during subsequent artificial ageing.

Fig. 10(b) shows the forming speed effect on the post-formed hardness of LT-HFQ[®] formed parts at a forming temperature of 335 °C. For all the cases, the draw ratio was 1.7 and the used BHF was 60 kN. Hardness increased with decreasing forming speed. The possible causes of such a trend were both the cylindrical deep drawing tool set-up and the deformation characteristics of test-piece. First, according to the tool set-up, the contact pressures in flange area under high BHFs could be higher than those in the punch nose, punch corner and sidewall areas, and ensure the sensitive precipitation zone to be skipped during quenching. Thus, for a lower forming speed, there might be a longer time for the work-piece to be sufficiently cooled by blank holders before being drawn-in. While for a higher forming speed, the forming was completed in a shorter time at a higher temperature; so the material might go through the sensitive precipitation zone due to both a relatively higher starting temperature for in-die quenching and insufficient quenching rate because of lower contact pressures in the punch nose, punch corner and sidewall areas. Second, another possible reason for the speed effect could be that referring the viscoplastic deformation mechanism, recovery was less for higher strain rates (Lin and Liu, 2003), which caused a higher level of dislocation density during forming. A higher level of dislocations would facilitate precipitation during ageing by providing more activation energy. Thus, a part formed at a higher speed was prone to be over-aged under the standard ageing condition.



(a) Preheating temperature effect. (Draw ratio 1.7, BHF 30 kN)



(a) Forming speed effect. (Draw ratio 1.7, BHF 60 kN)

Fig. 10. Effects of process variables (a) temperature and (b) forming speed, on the hardness of LT- HFQ® formed AA6082 after artificial ageing at 190 °C for 9 hours.

5. Conclusions

This paper reports experimental investigations of the drawability of AA6082 under two novel elevated temperature forming processes, LT-HFQ[®] and DHAF. It enabled the drawability of the alloy to be, for the first time, investigated and compared at a comparable temperature range between HFQ[®] and conventional warm forming. Based on the results of deep drawing tests assisted by hot tensile tests and hardness testing, the following conclusions have been drawn:

(1) The thermo-mechanical response of the alloy is dependent on the manner of its preheating conditions. For hot tensile tests, under the same deformation condition of 350 °C and 1 /s, an obvious strain hardening and improved ductility was observed for LT-HFQ[®] compared with DHAF processes.

(2) For deep drawing tests, the alloy processed through LT-HFQ[®] had a higher drawability than that processed under DHAF conditions. For a temperature around 400 °C and forming speed at 300 mm/s, the maximum draw ratio experimentally determined was 1.9 for LT-HFQ[®], which was unachievable for DHAF processes. The achievable draw ratios for DHAF-T6 and DHAF-W were 1.7 only.

(3) Effects of process variables on drawability have been analyzed systematically through DHAF processes. A lower forming temperature and higher forming speed contributed to improved drawability.

(4) The LT-HFQ[®] deep-drawn part could be age hardened to a higher level than a DHAF-W deep-drawn part, which was influenced by the preheating conditions. In addition, process variables of temperature and forming speed significantly affected the post-formed hardness of LT-HFQ[®] deep-drawn parts. Thorough understanding of the forming mechanism, mechanisms of microstructural evolution and viscoplastic deformation under dynamic non-isothermal conditions, as well as their interactions, are essential for the process designs.

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Table 1 Chemical composition of AA6082

Element (wt%)	Mg	Fe	Mg	Si	Cu	Zn	Ti	Cr	Al
min	0.4	-	0.6	0.7	-	-	-	-	Remaining
max	1.0	0.5	1.2	0.3	0.1	0.2	0.1	0.25	

Table 2 Strain hardening exponents and strength coefficients calculated from the uniaxial tensile data

Process	Condition	K (MPa)	n
DHAF-T6	350 °C $\times$ 1/s	137.366	0.030
DHAF-W	350 °C $\times$ 1/s	101.393	0.018
LT-HFQ [®]	350 °C $\times$ 1/s	114.434	0.066
	400 °C $\times$ 1/s	87.639	0.049
	450 °C $\times$ 1/s	70.295	0.038
	525 °C $\times$ 1/s	49.156	0.020