Some Aspects of the Metallurgy of Automotive Al Alloys

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Abstract

This paper considers some of the basic metallurgy associated with predicting and optimizing the behaviour of automotive sheet alloys. The strength of 5000 alloys can be predicted from a knowledge of their Mg content and grain size, and while grain size is not a strengthening factor in 6000 alloys, the T4 and pre-aged strengths, as well as their paint bake strength can also be estimated from their composition. Forming performance is a more subtle phenomenon, involving a range of factors associated with ductility and fracture, such as work hardening, surface roughening, and fracture processes. The paper discusses some recent work in these areas.

1. Introduction

Al alloys are used in many automotive applications, examples are 1000 and 3000 alloys for thin gauge, heat exchanger type applications, 4000 series alloys for castings, 6000 alloys for extruded and hydro formed sections, and 5000 and 6000 series alloys for sheet applications, such as inner structural parts and outer skins. While for many of these applications conventional Al alloys are used, the requirements for sheet are often quite demanding in terms of strength and formability, and this has led to special alloys being developed. The rationale behind the development of these alloys, and the different performance of the alloys has been covered in previous publications [1-3], and will not be repeated here. The present paper will consider some of the key issues associated with automotive alloys, such as their strength, aging response, and formability taking a somewhat generic view in regards to predicting and optimizing material behavior.

2. Strengthening Mechanisms

It is desirable that the initial sheet, prior to forming, has a relatively low strength so that its formability is high. However, the final formed part requires a high strength to enhance such properties as dentability and rigidity. For 5000 series alloys in the fully annealed, O-temper, the strength is controlled by solution hardening and grain size [4].

While solution strengthening theories predict a yield strength, $\sigma_Y \propto (c)^{1/2}$, where c is the solute concentration, the strength in commercial Al-Mg alloys can be well represented by a linear dependence on the Mg level, $\sigma_Y \propto B(wt\% Mg)$, where $B \cong 15$ MPa. The yield strength, σ_Y , of 5000 series alloys obeys the Hall – Petch relationship, $\sigma_Y = \sigma_0 + kd^{-1/2}$, where σ_0 is the frictional stress, which in this case includes any solution hardening, d is the grain size and k is the Petch slope. Combining solution and grain size hardening linearly gives the O-temper strength for 5000 alloys as:

$$\sigma_Y = \sigma_o + B(wt\%Mg) + kd^{-\frac{1}{2}}$$
⁽¹⁾

and while some variation in σ_o , B and C can be expected, depending on composition and processing conditions, typical values are $\sigma_o = 5-10$ MPa, B = 15MPa and where k is in the range 0.1 – 0.3MNm^{-3/2} depending on Mg level. Figure 1 shows a comparison of equation (1) with experiment for the commercial alloys AA5052(2.5wt%Mg), AA5754(3wt%Mg) and AA5182(4.5wt%Mg), and while it has to be appreciated that this range of grain size is achieved by variation in processing, that will influence crystallographic texture, and there are compositional variations in terms of Fe, Si etc, the agreement is quite good.



Figure 1: Comparison of a model for grain size and solution strengthening and experiment in 5000 alloys.

The 5000 alloys are non-heat treatable, so in the final part the only additional strengthening factor is dislocation strengthening from the straining during forming, $\sigma_F \propto \alpha Gb\rho^{1/2}$. The dislocation accumulation rate in the AI-Mg alloys is very high compared to Mg free alloys, because Mg atoms interact with dislocations, inhibiting dynamic recovery processes and giving rise to jerky flow. The work hardening rate is not strongly affected by grain size, although the situation is complicated by the fact that the alloys exhibit Lüdering, the extent of which is inversely dependent on the grain size, but also increases with increasing Mg level at equivalent grain size. The grain size is controlled commercially so that a low degree of tension or roller leveling can remove the yield point effect.

The strengthening due to dislocations is somewhat reduced by recovery occurring during the paint bake cycle, so the strength of the final part will depend on the level of strain during forming and the thermal history associated with the paint bake, though combining work hardening and recovery models predicts these strengths quite well [2].

The 6000 Al-Mg-Si-(Cu) alloys are heat treatable, but prior to forming they are in the unaged T4 temper. The strength of these alloys is not dependent on the grain size, but rather by solute/cluster strengthening, with the strength increasing with Mg and Si levels, and also with time at room temperature as clusters form during natural aging. For the Cu free alloys the natural aging kinetics are independent of alloy composition, and Figure 2 shows the natural aging of alloys with different Mg (0.5-0.8wt%) and Si (0.6-1.25wt%) levels. After about the first hour at room temperature the yield stress, σ_t after t hours at room temperature, is given by:

$$\sigma_t = \sigma_s + m \log(t) \tag{2}$$

where $m \cong 15$. A Cu addition increases the natural aging kinetics, with $m \cong 30$ for AA6111 (nominally AI-0.7wtMg-0.7wt%Si-0.7wt%Cu), and this can be explained on the basis of differences in cluster formation during quenching and natural aging [5]. So knowing the initial, quenched strength of the alloy, and the natural aging time, the yield strength can be estimated from equation (2).



Figure 2: Natural aging of 6000 alloys.

For a given natural aging time, the strength is dependent on the alloy composition and, as in the case of Mg, the solute and cluster hardening is linearly dependent on composition. Figure 3 shows the hardening associated with Mg content for different Si levels, and a similar graph for Si hardening at constant Mg level. Mg provides the greater hardening, but these relationships mean that the T4 strength of any of the automotive Al-Mg-Si alloy can be obtained from:

$$\sigma_Y = \sigma_A + H_{Mg}(wt\%Mg) + H_{Si}(wt\%Si)$$
(3)

where σ_A reflects the strength of the solute free matrix and any additional strength factors, such as dispersoid and texture hardening, are very small in comparison with the other two terms.



Figure 3: Cluster hardening in 6000 alloys.

A comparison of equation (3) and experimental results for a range of alloys and 2 weeks natural aging is shown in Figure 4: in this case $H_{Mg} = 105$ and $H_{Si} = 55$.

The situation is more complicated in Cu containing alloys, because while Cu also provides linear solution hardening, the functional relationship is dependent on the base alloy composition. To some extent this probably reflects changes in the phase diagram, with Cu atoms being incorporated in to the clusters, and not only metastable forms of Mg₂Si, such as β'' and β' , but also metastable forms of Q phase, Al₄Cu₂Mg₈Si₇.



Figure 4: Comparison of model with experiment for 6000 model and commercial alloys in T4.

Other solute elements, as well as dispersoid particles, can contribute to the strength of the 6000 alloys, primarily Fe and Mn, but the levels of these elements are low, and the dispersoids relatively coarse, typically 0.2 to 0.5μ m in the case of Mn, so these contributions are quite small compared to those in equation (3).

Strain hardening during forming and precipitation hardening strengthens the final formed part during the paint bake.

The work hardening rate of 6000 series alloys is not very sensitive to composition, and while strain hardening is a major contribution to the strength, the major difference between alloys is in the age hardening component. Considerable progress has been made in the last few years in predicting the precipitation kinetics and age hardening response of the 6000 alloys [6-8].

The conventional age hardened strength for the different alloys and heat treatments can now be predicted quite well. However, the temperature and times associated with the paint bake process are very low, and within the temperature - time region where the models are least successful. There is the additional complexity that the precipitation kinetics are a function of the extent of natural aging, with precipitation kinetics decreasing with increasing time at room temperature prior to aging. An age hardening model that accounts for natural aging has been recently developed for AA6111 [8], but it is also applicable to other 6000 alloys, and Figure 5 shows some results for AA6111.



Figure 5: Aging model for AA6111.

The model predicts the reduced aging response in naturally aged sheet, but the details of the model also suggest that for a given natural aging time and a specific paint bake treatment, the paint bake strength should scale with the alloy composition, and hence the T4 strength. This is demonstrated in Figure 6 for a 2 week natural age and a paint bake simulation using a 2% stretch followed by 30 mins at 177°C.

It is also apparent from Figure 5 that the aging response during the paint bake is quite small, but it can be increased by pre-aging the sheet after the solution treatment and prior to forming [6-9]. This treatment creates solute clusters that are stable at the paint

bake temperature, increasing the density of fine precipitates, and hence the paint bake hardening [10]. The pre-age has the additional effect of decreasing the rate of natural aging, and without increasing the strength significantly over T4 sheet, provided the pre-aging treatment is chosen appropriately.

While some details of the above modeling need to be modified, for a particular set of pre-aging and paint bake conditions, the pre-aged strength and paint bake strength can be predicted using the same approach as for the T4 condition.



Pre-aging gives a significant increase in the paint bake strength, as shown for different alloys in Figure 7.

3. Formability

The term "formability" is somewhat ubiquitous but it can be considered in terms of two limiting conditions: (1) the strain at which localization is initiated, and, (2) the strain at which fracture occurs.

The strain at which localization is initiated is the uniform strain in the tensile test, and the strain to neck nucleation typically identified in the Forming Limit Diagram or FLD. It is the strain that can be fully utilized in most forming operations, such as the stretching associated with the sheet forming of hoods. While tensile elongation is often quoted, which is the total elongation obtained in a tensile test and the sum of the uniform strain and the strain associated with the growth of a neck to final failure, it is the uniform strain that is of practical use. The total elongation is also very sensitive to specimen configuration, since neck growth is very sensitive to the development of triaxiality, while the uniform strain is dominated by material properties.

In Al alloys the uniform strain is usually in good agreement with the Considère criterion, $(d\sigma/d\epsilon) = \sigma$, which is a purely geometric criterion stating that when the work hardening rate, $(d\sigma/d\epsilon)$, decreases to the level of the flow stress, σ , a neck can nucleate and

subsequently grow to result in final fracture. Since this is a geometrical criterion it should be an upper limit for the uniform strain, and microstructural factors, such as coarse particles nucleating voids, could nucleate localization earlier than the Considère strain. However, voids are not extensively nucleated prior to neck nucleation.

The rate of neck growth is influenced by the strain rate sensitive, m=(dln σ /dln ϵ), but m is about zero or negative at room temperature in the alloys of interest. Features of the alloy microstructure such as particle size and distribution control void nucleation, growth and coalescence, and the post-necking strain to final fracture. For the Considère criterion the strain or stress dependence of the work hardening rate is the key parameter. Many expressions have been developed to represent work hardening in materials, but for convenience in the present context the Voce expression, as developed in differential form by Kocks and Mecking [11] is most convenient. In this case the work hardening rate, (d σ /d ϵ) = θ , is seen as the difference between the work hardening rate in the absence of recovery, θ_0 , and the extent of dynamic recovery due to dislocation annihilation and rearrangement during straining, (d θ /d σ), that results in the following expression for work hardening:

$$\theta = \theta_0 \left(1 - \frac{\sigma}{\sigma_s} \right) \tag{4}$$

where σ_s is the saturation stress at which $\theta = 0$. Equation (4) indicates that θ should decrease linearly with σ , but at large strains or stresses the work hardening deviates from equation (5), and does not decrease to zero as rapidly as expected. However, over the stress and strain ranges typically of interest in forming, equation (4) is a good representation of the stress – strain behavior, as shown in Figure 8. In this figure σ is expressed as $(\sigma - \sigma_Y)$ to normalize for the different yield strengths, σ_Y , and to only consider plastic deformation. The alloys investigated fall in to two groups. The first group, consisting of AA1100 and AA3003, which are low solute alloys, have the lowest work hardening rates, and $\theta_0 \cong$ 1200MPa, which is comparable to what would be expected from the Stage II work hardening rate, $\theta_{II} \cong G/20MPa$ for high purity AI. The second group, consisting of the 5000 series alloys and AA6111, are high solute containing alloys, and they have higher work hardening rates and a $\theta_0 \cong 2500$ MPa, about double that of the lower solute alloys. The rate of dynamic recovery decreases with increasing solute level, as shown by the high Mg AA5182 having a larger θ at comparable (σ - $\sigma_{\rm Y}$) than AA5052, which has a lower Mg content. This would be expected on the basis of Mg solute drag and the pinning of dislocations, thus hindering the ability of dislocations to rearrange and annihilate. Unlike the Mg alloys, AA6111- T4 does not exhibit serrated flow at conventional temperatures and strain rates, but the results indicate that the solute clusters present in naturally aged sheet inhibit dynamic recovery in a similar fashion to Mg in 5000 alloys.



Figure 8: Work hardening plots for Al alloys.

From the formability and Considère perspectives it is not only the level of work hardening rate but also the retention of work hardening at high stresses, and hence strains, that is important, and this is reflected in practice when AA5182 is used in preference to AA5052 or AA5754 in difficult to form parts.

However, for any particular alloy the uniform strain can appear to be influenced by other factors, such as the Mn and Fe levels, and variations in processing, but these can often be linked back to the effect on work hardening behavior. In highly alloyed material damage nucleation at particles can have an influence, but in the automotive alloys void nucleation only occurs at the point of incipient necking i.e. the uniform strain in a tensile test. The uniform strain is also sensitive to the test temperature, generally increasing at both low and high temperatures. As the temperature decreases both the work hardening rate and the strain rate sensitivity increase, tending to increase the strain to neck nucleation and decrease the rate of neck growth. At high temperatures the strain rate sensitivity increases, increasing the strain for damage nucleation.

The second limiting condition for formability is fracture. This is important in hemming, which is really bending, and fracture is also the limiting condition under crash conditions, so fracture processes are also important. The fracture strain, as measured by the reduction in area in a tensile test, is dependent on the alloy chemistry and microstructure, but it also scales with the yield strength of the alloy, provided the yield strength change is accomplished without any change in microstructure that modifies the mode of fracture [12]. This has been demonstrated in the heat treatable alloy AA6111 when natural or artificial aging increases the strength, both conditions exhibiting a linear decrease in fracture strain with increasing strength, but at different rates. In broad terms this behavior can be extended across different sheet alloys, provided their microstructures are comparable in terms of fracture behavior. This is shown for O-temper non-heat treatable alloys and T4 temper 6000 alloys in Figure 9, and the data indicate that the fracture strain for O-temper material would go to zero at a yield strength of about 350MPa.



Figure 9: The variation in fracture strain with alloy yield strength.

At the surface of a bend the deformation is plain strain tension, and Datsko and Yang [13] suggested that the failure in bending occurs when the true surface strain under bending is equal to the true strain at fracture in a tensile test, and developed an expression for the minimum bend radius, r_{min} for a sheet, thickness t, of

$$r_{\min}/t = (C/RA) - 1 \tag{5}$$

where RA is the fracture strain in tension and C is a constant. Figure 10 shows that the bendability for several different alloys are in general agreement with equation (5), and the constant C is in the range 65 - 75% when the fracture strain is expressed as a percentage.



Figure 10: Variation of bendability with fracture strain for (a) several alloys and (b) for aged materials.

Since the fracture strain also scales with the yield strength, the bendability deteriorates (i.e. r_{min}/t increases) with strain in the sheet. Microstructural features, such as grain boundary precipitates, crystallographic texture, and coarse particle content also affect the bendability. Figure 11 shows the affect of quench rate from the solutionizing

temperature on the bendability after different levels of prestrain; the higher quench rate produces a lower grain boundary precipitate level and hence improved bendability. Crystallographic texture and its spatial distribution affect the development of surface topography, while particle content influences damage development.



Figure 11: Influence of cooling rate after solutionising on bendability of a 6000 alloy.

4. Summary

The mechanisms controlling the strength of the 5000 and 6000 series automotive alloys are generally understood, and can be used to broadly define the alloy composition for a particular strength level. In the 5000 alloys the predominant mechanisms are grain size strengthening, solution strengthening and work hardening. Developing the appropriate grain size will be a function of the fabrication history, and alloy composition also plays a roll in terms of recrystallization nucleating particles, dispersoid particles to provide grain size control, and the level of cold work prior to the final anneal. The 6000 series are in some ways more complicated, since natural aging or pre-aging have a strong influence on aging kinetics, but functional models are available that provide guite a good guide. Designing alloys for optimum formability is more of a challenge since various features of the microstructure effect different formability parameters. While instability and necking is generally controlled by work hardening behavior, it may also be influenced by surface roughening, microstructural uniformity and the like that are not fully understood. Fracture behavior scales with the features of the microstructure that influence damage nucleation and growth, and understanding this requires a spatial quantification of the microstructure to accurately reflect microstructural inhomogeneity in fracture models. It should also be appreciated that there are other aspects of automotive sheet that are critically important, such as surface aesthetics, paint adhesion, joinability and corrosion resistance that have not been considered in this paper. However, Al automotive sheet is now at the stage where it can be considered as a reasonably mature product in the technical sense. Future developments are likely to focus on process development issues around appropriate alloy compositions.

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