Materials Research, Vol. 6, No. 2, 273-278, 2003. 

# Cracking Susceptibility of Aluminum Alloys During Laser Welding

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Received: November 19, 2002; Revised: February 23, 2003

The influence of laser parameters in welding aluminum alloys was studied in order to reduce hot cracking. The extension of cracks at the welding surface was used as a cracking susceptibility (CS) index. It has been shown that the CS changes with changing welding velocity for binary Al-Cu alloys. In general, the CS index increased until a maximum velocity and then dropped to zero, generating a typical λ-curve. This curve is due to two different mechanisms: 1) the refinement of porosities with increasing velocity and 2) the changes in the liquid fraction due to decreasing microsegregation with increasing velocities.

**Keywords:** *weld cracking, cracking susceptibility, aluminum alloys*

## 1. Introduction

Laser welding aluminum alloys offers many advantages: precise heat input, narrow weld bead, narrow heat-affected zone, minimal thermal distortion, as well as elevated welding speeds on thin sections and deep penetration on thick sections. It is because of these advantages that the application of laser welding of aluminum alloys is increasing, such as in the automobile sector for body and exterior paneling[1,2]. Although aluminum alloys are weldable by other more traditional fusion welding methods, the laser offers unique qualities that make it an ideal technology for joining aluminum alloys. Most strain-hardenable alloys (for example, AA5182 and AA5754) can be laser welded autogenously, although filler metal can be introduced during laser welding to add reinforcement or to improve the strength and ductility of the joint, if desired. Many heat-treatable alloys (for example, AA6016) are susceptible to hot cracking[3] during welding, due both to their chemical compositions and the thermal strains induced in the metal during welding. To avoid hot cracking, a filler metal is used to adjust the weld bead composition beyond the crack-sensitive range. Industrial experience has shown that hot cracking can be avoided by the addition of an eutectic alloy, such as Al-Si, to the weld. However, this leads to a shorter solidification interval. The filler wire technique has two major disadvantages. Firstly, the macroscopic properties of the joint will change in an uncontrolled way. Secondly, the small size of the laser beam leads to occasional feeding of filler wire directly into the beam, causing inconsistent penetration and weld-pool instability. For example, adding a 4043 Al-Si alloy wire to the 6061 alloy weld reduces hot cracking susceptibility[4], but both the tensile and ultimate strengths are reduced by 50%. Therefore an optimization of the process parameters in order to reduce cracking sensitivity would be highly desirable for the production of light weight high-strength components, especially for transport equipment industry.

Various studies about hot cracking phenomena in metallic alloys can be found in the literature. Clyne and Davies[5], states that the Hot Cracking Susceptibility (HCS) is directly linked to a critical ratio in the solidification period. Feurer[6] proposed that cracks will appear in the mushy zone if the rate of interdendritic liquid feeding is smaller or equal to the rate of shrinkage. In the same way, Piwonka and Flemings' approach[7] is based on the Poisseuille equation which describes the pressure gradient required to cause a

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fluid to flow along interdendritic path. All these theories are linked to the alloy solidification path and the HCS can be determined by the cooling curves in thermal calorimetry. For some binary alloys, it is verified the typical λ-shape when the Cracking Susceptibility (CS) is plotted as a function of the solute content. A more realistic approach of the CS criterion had been published by Rappaz *et al.*[8]. This criterion accounts for the tensile deformation of the solid skeleton perpendicular to the dendrite growth direction and for the feeding of the interdendritic liquid. The authors suggest that hot cracks will nucleate if the cavitation pressure reaches a critical value which can be calculated from the physicochemical properties of the alloy and the microstructural dimensions of the material. The typical λ-shape is maintained after Rappaz analyses however the strain-stress conditions in the solidification region of the melt pool readily changes both its magnitude and position.

Hot cracking susceptibility during welding is usually evaluated when the strain or stress is changed during the process. Some examples are the Sigmajig, Varestrain and Houldcroft tests[9,10]. These tests are performed in a specific mechanical device where loading or bending is applied. The major limitations of these methods include poor reproducibility due to complex mounting geometry and difficult interpretation of the results after testing.

A precise control of the solidification process reduces the cracking sensitivity. Using a pulsed Nd:YAG laser, Katayama and collaborators[11] developed a pulse shaping method in which the laser power is continuously decreased with time to control the rate of solidification. These authors showed that pulse shaping could effectively reduce hot cracking in Al-Cu alloys[12]. The major limitation of the method is the small processing velocity, as the pulse length increases from 5 to 20 ms, the welding speed is reduced by a factor of four.

Continuous laser processing is a desirable welding technique due to processing speed, weld quality and mechanical joining properties. The Al-Cu system is the base of commercial alloys which are particularly affected by hot cracking. The present work intends to determine the effects of the laser processing parameters such as power and scanning speed on the cracking susceptibility of Al-Cu alloys containing 0.5, 1, 2, 3, and 5 wt% Cu. Additionally, a major contribution is given to the study of a methodology for rapid Cracking Susceptibility evaluation of these alloys.

## 2. Experimental

Five different Al-Cu alloys with nominal compositions of 0.5, 1, 2, 3 and 5 wt% Cu were used to study the cracking susceptibility by laser remelting. Each alloy was prepared from high purity raw materials by vacuum melting and casting in a copper mold. The alloys were cast in rectangular shaped molds with 8 × 8 × 40 mm dimensions except for the 1 wt% Cu alloy which was cast in a 10 × 10 × 45 mm mold. The surface of the sample to be laser treated was polished using SIC 1000 mechanical grinding to remove the oxide coating and to ensure reproducible surface roughness. The polished surface was then processed using a 1.7 kW CW-$CO_2$ laser consisting of the laser source, a beam delivery system, and a CNC worktable. The distance between the mirror and the sample was fixed at 152 mm and the incident laser beam was at an angle of 8 degrees. The diameter of the beam was 0.26 mm. Helium was blown over the sample at a rate of $3 \times 10^{-3}$ $m^3$/min to prevent massive oxidation. Laser welding was conducted at ambient temperature along the width of each sample, Fig. 1. The laser power and speed were varied for each pass, with power settings of 1200, 1300, and 1400 W and scanning speeds of 1, 2, 5, 10, and 50 mm/s.

The surfaces of the samples after laser processing were initially polished with 1000 SiC paper to eliminate the surface roughness. The polishing then proceeded to 6 mm diamond spray on a cloth wheel at 400 rpm until the traces from the 1000 SiC paper polishing were eliminated. Finally, 1 µm diamond polishing was used at 200 rpm to obtain a mirror-like surface. Ethanol was used throughout the entire polishing process to keep the sample-wheel interface lubricated. In between each level of polishing, the sample was ultrasonically cleaned for 10 min in an ethanol bath to prevent contamination from the prior polishing step. Once the sample attained a mirror-like surface, deep etching was conducted using Keller's Etch, consisting of 1% HF (48%), 1.5% HCl, 2.5% $HNO_3$, and 95% $H_2O$. The time for sufficient deep etching varied with each alloy somewhat, but 15 s was found to be adequate for most samples. With a

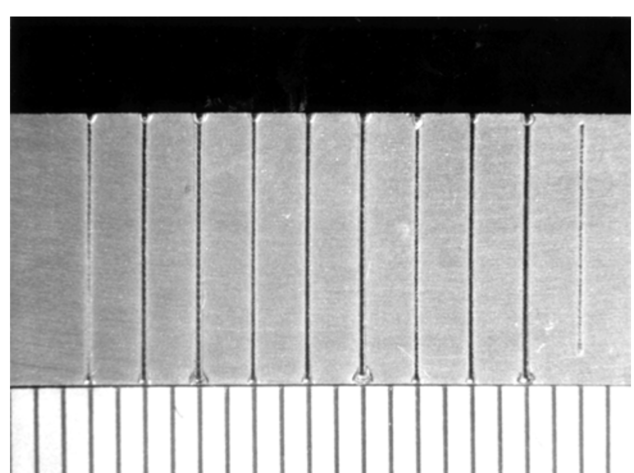

**Figure 1.** Laser welding on the aluminum surface (milimetric divisions).

proper deep etch, the laser traces, fissures, and the grain boundaries were well defined. Then the sample was repolished using the 1 μm diamond polishing compound to remove etched grain boundaries, leaving only the fissures and the faint appearance of the liquid trace. As discussed later, the repolishing process was found to be necessary to eliminate any doubt of whether a microstructural feature is a grain boundary or a fissure.

The as-polished surface of the samples was observed using an optical microscope at a magnification of 200×. The regions with the highest density of fissures were identified and photographed. To quantitatively define the cracking susceptibility (CS) of each trace, the photographs were enlarged and the total length of the fissures was determined. For curved fissures, the length of the fissure was broken into small linear segments in order to obtain more accurate length determination. The summation of all the fissures in the region was divided by the surface area of the remelting trace to obtain the cracking susceptibility for the given laser power and speed combination for each alloy.

## 3. Results and Discussion

The general surface feature of a laser remelted sample before being polished is shown in Fig. 1. The figure shows traces of 10 laser passes. The width of the remelted zone was found to depend on the power for any given laser speed. In general, the width of the remelting zone increased with increasing the laser power.

The laser-melted zones were found to contain different degrees of fissures and to depend on the alloy composition and laser processing parameters. An example of a fissure from the alloy with 1 wt% Cu can be seen in the SEM image given in Fig. 2. The fissure shown in the figure is relatively short and lies diagonal to the laser travel direction. Some samples showed more extensive fissures as exemplified with the optical micrograph of Fig. 3.

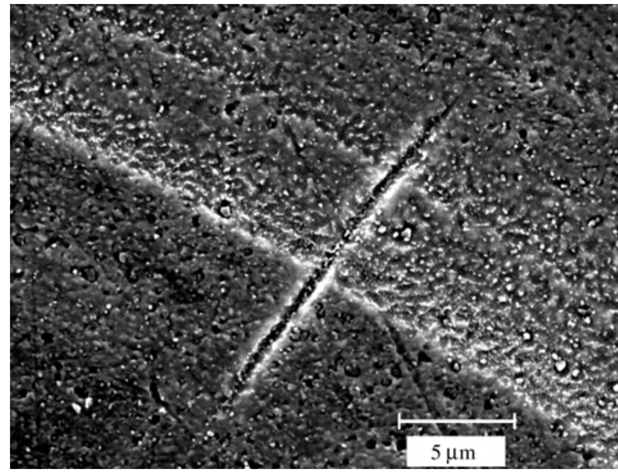


**Figure 2.** Scanning Electron Microscope (SEM) image of a fissure in a remelted trace

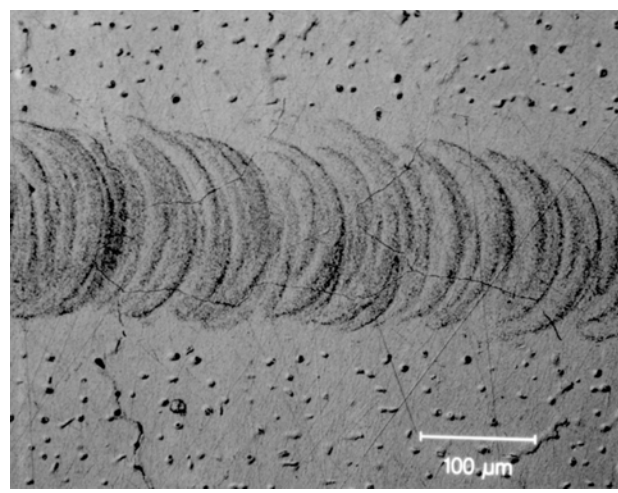


**Figure 3.** Al – 2 wt% Cu sample remelted with a laser scanning speed of 5 mm/s and power of 1300 W. The region shown has a high density of fissures.

The total fissure length in the most heavily-fissured region of each laser melting trace was measured visually from the photographs. As indicated earlier, the cracking susceptibility (CS) was then obtained by dividing this length by the surface area of the laser melted zone. It should be noted that because of the fineness of the fissures and the lack of sharp contrast between the fissures and the remelted liquid traces, the computer image analysis technique could not be used for quantifying the cracking susceptibility. Using a fluorescent dye penetrant was also found to be inadequate, as the dye did not penetrate the fissures.

Other restrictions presented by the present method for measuring CS include omission of fissures too fine to be visible with 200× magnification and the difficulty in differentiating between deeply-etched grain boundaries and fissures. Porosities before laser treatment due to casting were observed with a diameter of 10 μm. For this study it was assumed that fissures shorter than this diameter would not further deteriorate tensile property of the remelted zone over that of the base metal. Thus fissures shorter than this critical length were not considered in the determination of CS. The only fissures then that might have been omitted would have to be extremely fine to escape observation. For the differentiation between a fissure and deeply etched grain boundary, the repolishing of the etched surface was found to be adequate in removing the latter features. Moreover, a critical evaluation of a potential fissure was conducted using polarized light approaching at an angle. By adjusting the angle of the incident light, it was then possible to view the depth of the fissure and differentiate it from grain boundaries.

The above mentioned fissure density measurement does not account for fissure width. Notwithstanding, the measurement is believed to yield a more applicable quantification as it is the presence of fissures longer than the critical length which determine the failure of the laser welded region.

The cracking susceptibility for the alloys containing 0.5, 1, 2, and 3 wt% Cu is plotted against the laser scanning speed in Fig. 4a-d, respectively. No cracking was observed in the alloy containing 5 wt% Cu at any of the laser power and speeds utilized. The maximum error of the cracking susceptibility data points plotted in the figures was determined to be less than 4.6% for a 95% confidence level. This error was found to be dependent on the geometry of the fissures: being as low as 0.5% for small and straight fissures, and as large as 4.6% for heavily curved and extended cracks.

For the sample containing 0.5 wt% Cu, as shown in Fig. 4a, the cracking susceptibility at 1200 W followed the so-called "λ" shape with increasing scanning speed. The susceptibility increased with increasing the laser scanning speed from 1 to 2 mm/s. At higher speeds of 5, 10, and 50 mm/s, no cracks were observed in the laser melted traces. The cracking susceptibility decreased with increasing the laser power from 1200 to 1300 W, and totally diminished as the power was increased to 1400 W.

The "λ" trend was also observed for the cracking susceptibility of the sample containing 1 wt% Cu, as shown in Fig. 4b. However, unlike the sample containing 0.5 wt% Cu, cracking was observed at all power settings when the velocity was less than 10 mm/s. Further, the influence of the laser power on the susceptibility seemed to be less pronounced, as cracking was observed at all the power levels utilized.

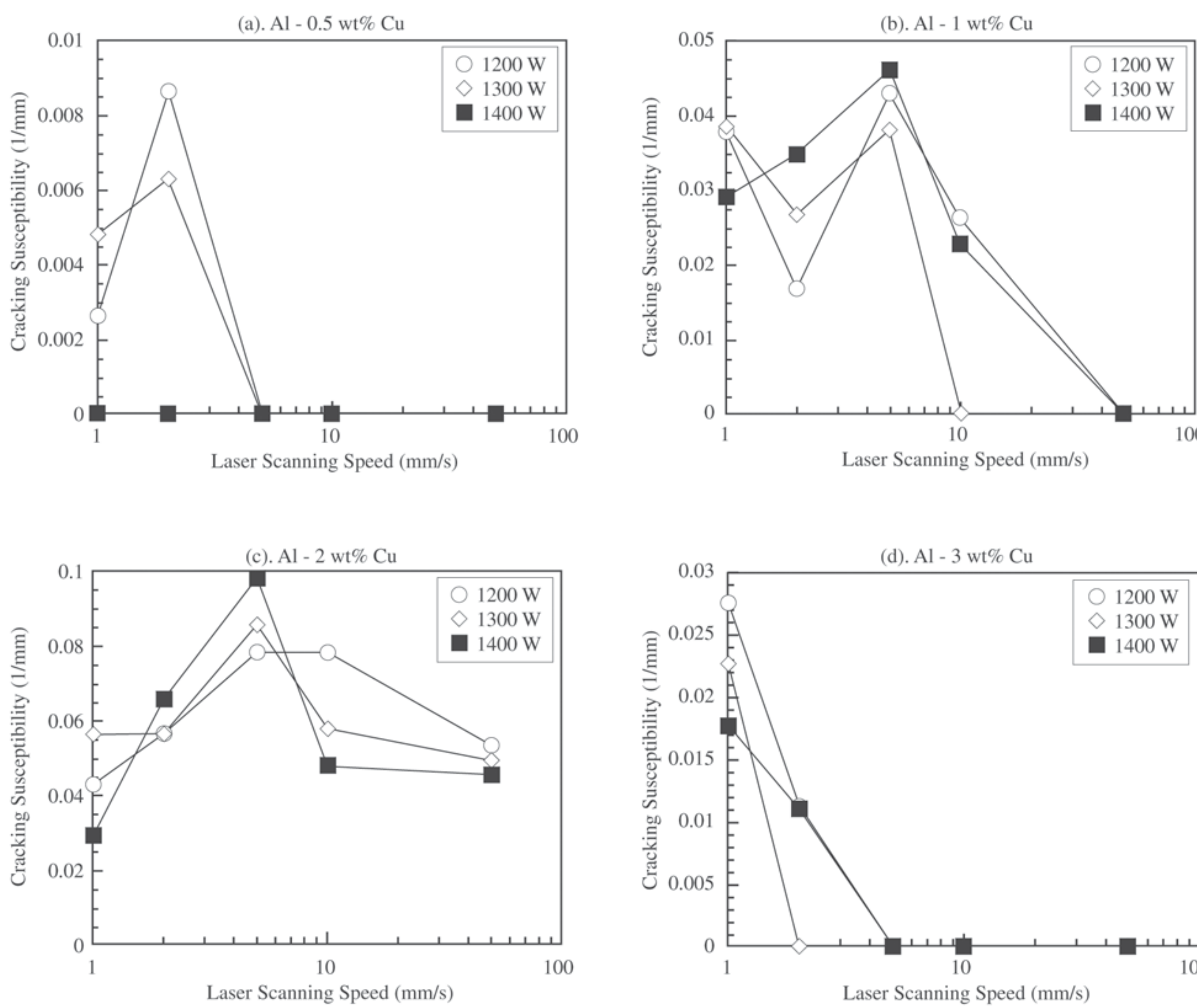


**Figure 4.** Cracking susceptibility of each alloy as a function of laser scanning speed.

For the sample containing 2 wt% Cu (Fig. 4c), the trend for the cracking susceptibility followed the same general trend as the samples containing 0.5 and 1 wt% Cu. However, the cracking susceptibility did not diminish completely at any of the speeds utilized. On the other hand, for the sample containing 3 wt% Cu, the susceptibility decreased with increasing the scanning speed form 1 to 2 and 5 mm/s. No cracks were observed at speeds 10 and 50 mm/s. It is conceivable that the susceptibility data for this alloy represents the right side of the "λ" shape.

At the scanning speeds utilized, it was found that changing the laser power between 1200, 1300, and 1400 W produced no noticeable effects on the cracking susceptibility. The only noticeable effect of augmented power was the slightly increase in the diameter of the remelted trace.

As indicated earlier, no cracking was observed for 5 wt% Cu samples with the laser scanning speeds utilized. In fact, at low speeds, development of planar front growth was observed, indicating that constitutional undercooling was almost obtained. It is conceivable that had the scanning been performed at higher speeds, the data would have covered the "λ" shape.

The curves obtained for changing the scanning speed at a given power (Fig. 4a-4d) can be seen to demonstrate the presence of two different effects: (1) pore refining and (2) decreasing solute rejection with increasing processing velocity.

As the scanning speed is augmented, the microstructure of the remelted volume becomes more refined. With this refinement, the radius of interdendritic pores left after solidification decreases. Decreasing pore radius decreases the probability that pores will link together through fissures, thus this effect would indicate that augmenting scanning speed decreases cracking susceptibility. This effect leads to a continuous decrease of CS with increasing velocity.

The second effect can be explained in terms of changes in microsegregation that occur with changes in laser scanning speed. Given an alloy composition, the microsegregation will be large at low scanning speeds and then go through a minimum as the velocity is increased[13]. Figure 5 presents the effect of growth velocity in the partition coefficient of the studied Al-Cu alloys using Aziz formalism[14]. The residual liquid due to segregation decreases as solidification velocity increases. In a first stage, the reduction in the liquid fraction between dendrite increases the cracking susceptibility (Fig. 4a-4c). This is because the rate of interdendritic liquid feeding is smaller than the rate of shrinkage due to viscosity and surface tension[6]. In second stage for higher velocity, the liquid volume is even smaller however the critical path for cracking networking is surpassed. The mushy zone can accommodate the welding stresses which are distributed to the bulk material[5]. The "λ" shape seen in the figures for each alloy seems to follow this trend. At low solute compositions, cracking susceptibility is also low as the alloy behaves more like a pure metal.

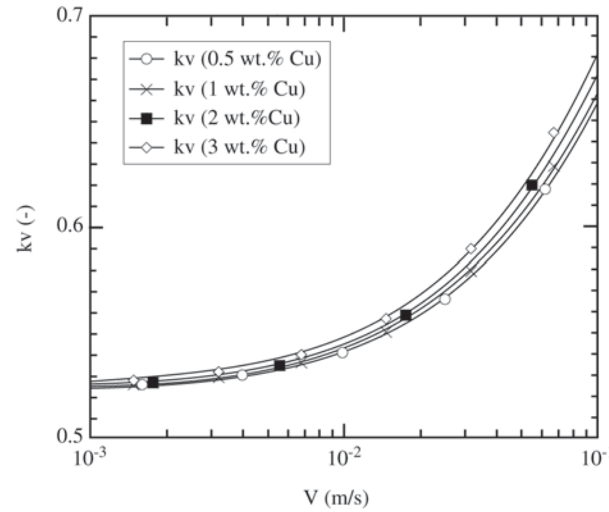


**Figure 5.** Solute trapping effects in Al-Cu alloys in the selected range of growth velocities.

It must be pointed out that the solidification velocity can be different from the welding speed because the welding speed represents the isotherms speed at the centerline of the fusion zone, whereas dendrites grow in some specific crystallographic directions leading to an angle between growth and thermal gradient orientations. Consequently the velocities reported in Fig. 5 must be seen only in qualitative manner.

Finally, the present method of laser welding and observation procedure is proposed as an alternative to the traditional Cracking Susceptibility methods. The limitations of the method as pointed out above can guide for effective utilization of the procedure.

## 4. Conclusions

Laser welding of binary Al-Cu alloys was studied from the cracking susceptibility point of view. This study has yielded a methodology for quantifying cracking susceptibility of the remelted traces. Using this methodology, scanning speed versus cracking susceptibility was plotted for each alloy and each power. Two factors were determined to affect cracking susceptibility. First, increasing scanning speed decreases cracking susceptibility due to pore refinement. Second, microsegregational effects yield a λ-shape with increasing speed. Cracking susceptibility is linked to a given liquid fraction. Due to solute trapping effects, this critical limit is reached at some growth velocity linked to a welding speed beyond which crack formation ceases. As alloy composition is changed, the two effects maintain the same shapes, but shift according to scanning speed.

## Acknowledgements

The authors would like to thank J.-D. Wagnière for the laser experiments as well as Professor W. Kurz for financial support and the use of laboratory facilities.

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