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# Using evolutionary algorithms to determine the residual stress profile across welds of age-hardenable aluminum alloys

J. Ignacio Hidalgo [a,*], Ricardo Fernández [b], J. Manuel Colmenar [c], Florencia Cioffi [b], José L. Risco-Martín [a], Gaspar González-Doncel [b]

[a] Adaptive and Bioinspired System Group, Universidad Complutense de Madrid, Calle del Profesor José García Santesmases 9, 28040 Madrid, Spain
[b] Department of Physical Metallurgy, Centro Nacional de Investigaciones Metalúrgicas, CENIM, C.S.I.C., Av. de Gregorio del Amo 8, 28040 Madrid, Spain
[c] Universidad Rey Juan Carlos, Calle Tulipán, 28933 Móstoles, Spain

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ABSTRACT

This paper presents an evolutionary based method to obtain the un-stressed lattice spacing, $d_0$, required to calculate the residual stress profile across a weld of an age-hardenable aluminum alloy, *AA*2024. Due to the age-hardening nature of this alloy, the $d_0$ value depends on the heat treatment. In the case of welds, the heat treatment imposed by the welding operation differs significantly depending on the distance to the center of the joint. This implies that a variation of $d_0$ across the weld is expected, a circumstances which limits the possibilities of conventional analytical methods to determine the required $d_0$ profile. The interest of the paper is, therefore, two-fold: First, to demonstrate that the application of an evolutionary algorithm solves a problem not addressed in the literature such as the determination of the required data to calculate the residual stress state across a weld. Second, to show the robustness of the approximation used, which allows obtaining solutions for different constraints of the problem. Our results confirm the capacity of evolutionary computation to reach realistic solutions under three different scenarios of the initial conditions and the available experimental data.



## 1. Introduction

Aluminium Alloys (*AA*) are used in a plethora of industrial processes. For example, alloys of the *AA2xxx* series are commonly used in aircraft structures. In many applications, AAs need to be welded, a process which is difficult to control using common fusion welding techniques in comparison to other metals, such as steels. This is due in part to the high thermal conductivity of aluminum. As an alternative to conventional methods, Friction Stir Welding (FSW) [1,2] has proven to solve most of the difficulties associated with the joining of aluminum alloys [3–5].

However, the development of a Residual Stress (RS) is, as in other welding procedures, inherent to this joining operation. In FSW the existence of a RS is mainly due to the non-homogeneous nature of the severe plastic deformation phenomenon involved and the heat generated during the welding process. The FSW process also involves significant changes in the microstructure and the mechanical properties of the material. There are various methods for determining residual stresses in a component. We can use destructive techniques, such as "hole drilling", sectioning or indentation, and nondestructive techniques, such as those based on X-ray and neutron diffraction [6]. These diffraction techniques use the lattice spacing as an atomic probe to measure elastic deformations. By comparing the lattice spacing in the stressed and un-stressed condition, one can determine, first, the elastic deformation and, then, the associated stress using well established linear elasticity theory equations and tabulated elastic constants. One of the advantages of the diffraction techniques is that the full triaxial stress state inside a component can be investigated if one takes advantage of the high penetrating capacity of neutrons and (hard) X-rays [7,8].

As it will be shown, the crucial factor to determine the RS state by these methods is the availability of a reliable un-stressed lattice spacing value, $d_0$. When a unique constant, $d_0$ value is sought, conventional analytical procedures can be applied to determine the residual stress state. For the case of age-hardening AA2xxx series alloys, however, $d_0$ is strongly dependent on the amount of atoms in solid solution (such as copper atoms) and, in turn, on the heat treatment imposed to the component. If the heat treatment is uniform through the sample, a constant $d_0$ is expected, and the RS calculation is straightforward. But for the case of welds the heat treatment varies dramatically across the weld and, as a consequence, a variation of $d_0$ should be taken into account. For these cases, conventional analytical procedures are no longer applicable [9,10].

Recently, evolutionary algorithms (EAs) have been applied for solving optimization, modeling and identification problems [11–14]. The present investigation addresses the capacity and robustness of EAs to solve the above difficulty. Several scenarios, depending on the initial constraints imposed by the experimental data available from neutron diffraction experiment, are discussed. Specifically, the unstressed lattice spacing variation under the three scenarios and the resulting RS profile across the weld are obtained. The use of EAs, applied to current used methods to determine residual stress fields, is very appropriate when $d_0$ value is not constant. We solve this optimization problem using a Multi-Objective Evolutionary Algorithm (MOEA). The nature of the problem, where the values for a set of parameters have to be found, allows for a direct encoding of these parameters within a chromosome as in other problems [15]. In summary, the motivation of this research relies on the need of alternative methods to conventional analytical ones in order to find the required information (in this case, the $d_0$ values) to calculate the residual stress profile across welds of age-hardening Al alloys.

* Corresponding author. Tel.: +34 913947537; fax: +34 913847527.
*E-mail address:* hidalgo@ucm.es (J. Ignacio Hidalgo).

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We have found that the experimental results are metallurgically consistent, allowing the characterization of the residual stress state of the weld.

The main contributions of the present work are the following:

- We present a method for obtaining the distribution of $d_0$ parameter, based on a MOEA, under different measurement scenarios.
- We use for the first time the capacity of EAs to calculate the residual stress across a weld of 2024 alloy using neutron diffraction data of the lattice spacing variations.
- We analyze the robustness of the proposed MOEA.

The rest of the paper is organized as follows. A description of the problem is made in Section 2. Section 3 describes the metallurgical experimental details. Section 4 gives the details of the evolutionary algorithm implementation. Section5 presents the experimental results in terms of both the metallurgical solution and the convergence and robustness of the evolutionary approach. Finally, Section 6 summarizes the conclusions of this work.

## 2. Problem description

Any forming process used in the industry (rolling, extrusion, welding, deep drawing, etc.) for the fabrication of components leads to the development of a RS which extends over the resulting component. This is because, generally, these procedures are non-homogeneous in nature and the development of a RS is inherent to non-homogeneous dimensional changes of bodies (due to deformation, thermal effects, or both). Because this RS varies throughout the whole piece, it is usually referred to as a macroscopic stress, M-RS, to distinguish it from the microscopic stresses, m-RS, inherent to the material microstructure [16]. Understanding the development of M-RSs and their magnitude is of great importance in structural engineering because it is a key factor in the design of structural components. Disregarding this stress may be critical and the cause of component rupture at stress levels below that at which failure would be expected from the applied stresses alone. As mentioned above, several techniques have been developed in the past decades aimed at determining this stress.

Probably, the most successful ones among them are those which make use of the diffraction phenomenon. Given the crystallographic nature of most engineering materials (metals), diffraction allows determining with high precision, lattice spacing values, $d_{hkl}$. In this way, an expansion or contraction of the given lattice spacing is identified with an elastic strain and, associated to it through the linear elasticity theory, the corresponding internal stress. Determining this expansion/contraction of the lattice spacing with respect to a relaxed value, $d^0_{hkl}$, is, hence, the core of the diffraction methods to calculate internal (and residual) stresses in structural materials and components. The use of diffraction methods implies the use of Bragg's law,

$$2 \cdot d_{hkl} \cdot \sin(\theta) = \lambda \tag{1}$$

where $\theta$ is the Bragg's angle and $\lambda$ the wavelength of the incident beam. Bragg's law allows determining $d_{hkl}$ for different spatial directions. Then, by using the formula,

$$\epsilon = \frac{d_{hkl} - d^0_{hkl}}{d^0_{hkl}} \tag{2}$$

the elastic strain, $\epsilon$, can be calculated in the same way as macroscopic strains are obtained in conventional mechanical testing with the help of a suitable extensometer. Then, direct use of the generalized Hookes law of elasticity, namely,

$$\sigma_i = \frac{E}{(1+\nu)\cdot(1-2\nu)} \cdot [(1-\nu)\epsilon_i + \nu(\epsilon_j + \epsilon_k)] \tag{3}$$

where $\sigma_i$ is the $i$ component of the stress tensor, related to the three strain components, $\epsilon_i$, $\epsilon_j$ and$\epsilon_k$, $E$ the elastic modulus, and $\nu$ the Poisson's ratio provides the stress state of the component under study.

Formerly, when only standard laboratory X-rays were available, determining the residual stress in components was limited to the near surface region where a bi-axial stress state should be assumed. Under this approximation, the extra calculation of an un-stressed spacing value is not needed since it can be obtained from the fact that the normal stress component is null.

More recently, however, the availability of high penetration capacity of neutrons and synchrotron X-rays has allowed calculating the full tri-axial residual stress state inside materials. In these cases, the extra calculation of an un-stressed lattice spacing value is required. In fact, it is very likely that the most critical issue and a key factor for the determination of this M-RS is, precisely, the availability of an accurate and reliable value of $d^0_{hkl}$. This is because small fluctuations or experimental errors in this parameter (and of $d_{hkl}$ itself) can lead to huge variations of the calculated strain and the resulting stress. For example, a variation of some few $10^{-14}$ m (or $10^{-3}$ Å) in the value of $d^0_{hkl}$, say $2 \times 10^{-14}$ m, can lead to a variation of the stress in the order of 10 MPa for the case of aluminum alloys. This variation increases with an increase of materials stiffness, e.g., titanium or steels. As a consequence, the development of reliable methods to determine $d^0_{hkl}$ has been the subject of many efforts. A detailed description of these methods can be found in [17,18]. Here, a brief description of the main ones namely, the powder method, the sample sectioning method, and the use of the equilibrium conditions is given.

### 2.1. Powder method

Powder of the same material under study guarantees a macroscopic un-stressed reference since powder is always free of long range stresses. Furthermore, powder is susceptible to undergo a heat treatment if necessary: it is known that the lattice spacing can be strongly affected by the amount of solid solution elements and, hence, of the precipitation state. This is the case of some age-hardening aluminum alloys. These aluminum alloys, such as those of the 2*xxx* and 7*xxx* series, widely employed in the aeronautical an aerospace industries, have the particularity that their mechanical properties are strongly dependent on the thermal history, i.e., their precipitation state. Hence, if powder is to be used as reference for $d^0_{hkl}$ determination in components fabricated from one of these alloys, one must be sure that the heat treatment undergone by the alloy powder is the same as that of the component under study. This task is, however, virtually impossible in some cases, for example in welds, where the specific thermal cycle imposed during this process is not known. Furthermore, it may change dramatically across the weld. This implies that the powder method cannot be recommended as a way to determine $d^0_{hkl}$ in complex cases such as in welds of age hardening alloys.

### 2.2. Sample sectioning in small pieces or combs

In the same manner as powder particles do not hold a M-RS, a sample that is cut into small pieces relieves its M-RS. In this way, one can section the sample (or a twin replica if it is not possible to section the sample) such that the size of the pieces is smaller than the expected residual stress variation: The larger the stress gradient or the shorter its range, the smaller the piece dimension to ensure stress relaxation. The use of these small pieces, cut from different parts of the sample, allows determining the specific $d^0_{hkl}$ value at the corresponding sample position and avoids errors associated with compositional changes or precipitation variations resulting from possible heat treatments. When large number of pieces are required, as in the case of welds, they can be organized by using "comb" samples. These are samples where the different stress relaxed pieces are maintained together and, hence, "recall" their specific coordinates with respect to those in the original component